THE CREEP OF METALS AND ALLOYS

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A survey of this kind is essentially a description of the results obtained by other investigators; the author has drawn freely from a large number of original papers and he hopes that full acknowledgment of the work of the investigators concerned, and of the source from which information has been taken, is provided in the text and by reference to the bibliography.

Finally the author wishes to thank Professor N. F. Mott, Dr. E. Orowan, Dr. H. J. Tapsell, Professor F. Seitz, Mr. T. A. Read, and the following publishers and Societies, for their kindness in so readily granting permission to reproduce tables and diagrams: The Oxford University Press (for Figures 6, 7, 9, and 16); The McGraw-Hill Book Company (for Figures 3, 20, and 23, and for the table on page 25); The Reinhold Publishing Corporation (for Table I); The American Institute of Mining and Metallurgical Engineers (for Figures 2a, 2b, 10, 11, 13, 14, 15, 17, 18, 21, and 22); The American Institute of Physics (for Figure 24); The American Society for Metals (for Figure 12).

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FOREWORD

by

Professor N. F. MOTT, M.A., D.Sc., F.R.S.

CREEP OF METALS is so important a subject for engineers and designers that an up-to-date account of any aspect of the subject is to be welcomed. It would be beyond the scope of a book of this size, or perhaps of any book, to summarize all that is known about the creep of metals and alloys and the factors that affect resistance to creep ; as the author says, information concerning many of them is rather indefinite at the moment, and he is wise to limit himself to a broad indication of trends. It is particularly interesting that, for the first time in a book on creep, an account has been given of the present state of the theory of the subject. Admittedly the theory is as yet in a primitive state ; it has not yet advanced to the point reached in other branches of applied science, such as electronics, rectifiers and so on, where it is of direct use to the man who develops materials. It is probable, however, that this time will soon come, and it is all to the good that a book meant for the practical scientist should give some attention to the developments which may well affect the future progress of the subject.

PREFACE

IT IS THE purpose of this book to discuss the technique adopted in the measurement of creep and to review the information which has been gained in the study of the creep of metals and alloys. In building up the theme, it is hoped that sufficient detail has been collected to emphasize the gaps in existing knowledge, and to indicate the lines along which future research work might profitably be pursued, with the object of establishing a more ordered system of knowledge upon which engineering design may be based, and to which metallurgical theory may appeal for confirmation and guidance.

The text is divided into five chapters. In Chapter I the experimental procedure adopted in creep testing is surveyed in the light of the author's own experience in connection with the measurement of the creep of aluminium alloys. Considerable space is allotted to this chapter since, in any investigation of the creep characteristics of metals and alloys, it is important that extreme care should be taken to ensure: (1) sufficient accuracy of control of testing conditions, (2) reproducibility of testing conditions and (3) repeatability of test results.

Chapter II is devoted to a discussion of the form of the creep curve and of the physical interpretations which are associated with each of the stages of the curve.

Chapter III is concerned with the factors which influence the creep characteristics of metals and alloys. This subject is treated under three main headings, namely: (1) grain size, (2) heat-treatment and structural stability and (3) chemical composition. Reference is made to other factors which influence creep, but information concerning them is so much in doubt at the moment that it is not deemed advisable to treat each of them separately; any relevant details of their effects are, therefore, included under the main headings.

In Chapter IV an account is given of the presentation of creep results and of the methods which are adopted in order to obtain data upon which engineering design may be based. Creep results are often presented in a form which so lacks essential detail that it is very difficult, if not impossible, to establish a satisfactory interpretation of their significance. It is important that the results submitted by different laboratories should conform to a standard of detail sufficient to render them capable of comparison and of interpretation in terms of knowledge gained from engineering experience.

Chapter V deals with the physical study of the mechanism of creep. It is not possible in the space of one chapter to give a detailed account of the application of the theories of plastic flow in solids to this particular phenomenon, but an attempt is made to outline the stages in the development of the fundamental treatment of the subject. Progress in this field of fundamental research has been comparatively slow but, during the past few years, the attention of the mathematical physicist has been drawn more and more to the mechanical behaviour of metals under varying conditions of stress and temperature, and it is hoped that the near future will see rapid strides towards the establishment of a theory which will lead to a more comprehensive understanding of what is happening during the progress of creep, and to a more exact knowledge of the mechanisms which govern the flow of metals.

INTRODUCTION

THE IMPORTANCE of the creep of metals as a factor to be considered in the design of equipment for service under stress at elevated temperatures first began to gain recognition among engineers after the publication of work carried out, between 1919 and 1926, by Chevenard(1), French(2) and Dickenson(3). This work emphasized the significance of the fact that the mechanical properties of metals, as determined by short-time tests at normal temperatures, are changed by exposure to elevated temperatures. In particular no definite limit of elasticity is observed and the application of stress causes continued extension to fracture in a period of time dependent upon the stress and temperature in question. The publication of Dickenson's work, in 1922, was extremely valuable in this respect since he carried out the first tests of long duration and, in doing so, he demonstrated the inadequacy of short-time tests as a guide to the elevated temperature strengths of metals.

Most of the early work, undertaken within the period 1920 to 1930, on the properties of metals at elevated temperatures, was concerned with the time to failure under a given tensile stress at a particular temperature. Design, however, is usually as much concerned with strain as with stress, since the amount of deformation which can be tolerated during service is limited in most applications, and this fact led to the study of the deformation, with time, of metals subjected to externally applied stresses at elevated temperatures. Dickenson described the deformation which takes place under such conditions as " Flow "; on the Continent it is often referred to as the " Viscosity—Time Effect," but in England and America the phenomenon has become known as " Creep."

Creep may be defined as the deformation of a metal occurring with time under, and due to, an applied stress, whether the deformation be of the nature of plastic flow or of viscous type flow. As

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[1] The Numbers in Parentheses refer to the list of references given in Appendix I on pages 128 to 151.

defined, creep is not meant to include dimensional changes with time due to structural or constitutional changes.

Little information is available concerning the nature of the flow which takes place during deformation and, as yet, it is impossible to differentiate between the relative amounts of plastic flow and viscous type flow which occur during creep. When plastic flow occurs in solids it is usually considered that a definite finite force is required to produce the shearing action which accompanies the flow, and that such flow will stop after a time. Flow, whose rate is proportional to the applied stress, and is continous with time for any stress however small, is referred to as viscous flow. If creep were truly viscous in nature, ultimate failure of a metal would result however small the initial creep rate. Experience has shown that this is not the case, and the form of the creep curve for metals at service temperatures does not represent viscous flow as defined above.

In the experimental work on this subject attention has been largely devoted to the study of the behaviour of metals subjected to static tensile loads, and it is proposed to confine this survey to the consideration of this aspect of the subject alone.

Although in initial studies of the flow of metals(4) attention was devoted to non-ferrous metals and alloys, the bulk of the information which is available at present is concerned with steels. Steels have found greater application in service at elevated temperatures than have non-ferrous metals and alloys, and the need for knowledge of their behaviour under such service conditions has thus been more urgent. In recent years, however, the development in design of internal combustion engines, for use in aircraft in particular, has provided application for non-ferrous metals and the shortage of information concerning their properties at elevated temperatures has been a source of acute handicap to the design engineer.

In the past twenty-five years considerable information has been gained, and much has been written, on the subject of the creep metals, but as Gillett(5) has pointed out, a major part of the published work contains plausible, but untrue, generalizations about creep which are based on results obtained from a very limited field of experimental work. In addition, the results of a great deal of the work are of doubtful value owing to the fact that insufficient attention has been devoted to the control of testing conditions. There is still a distinct shortage of reliable information, and, while for the time being, legitimate use may be made of empirical laws for guessing the answers to problems which arise in connection with engineering design, the presentation of hypothetical theories based on limited experimental evidence serves no other purpose than add to the confusion, which already exists, of the conception of creep.

The results of the work of investigators who have been engaged on the study of creep have led to the belief, which is now generally accepted, that this phenomenon is one of the most sensitive of the physical properties of metals, for it is influenced by many factors which have little or no effect on the more well-known properties. Owing to the complexity of the subject, it cannot be said that all factors which modify creep results have yet been identified. Although information concerning those factors which are known is very limited, certain of them have been studied in some detail and those most commonly investigated and discussed are: (1) chemical composition; (2) grain size; (3) heat-treatment; (4) previous deformation; (5) structural stability; (6) structural uniformity; (7) temperature of recrystallization and (8) method of manufacture. These factors are, to some extent, overlapping; for example, heat-treatment influences the actual grain size and often the structural stability; likewise chemical composition may influence the structural stability and recrystallization phenomena. Thus the study of the effect of any one factor is a difficult matter, nevertheless the results of investigations in which the effects of variations in the separate factors have been observed are of the utmost importance from both the engineering and physical points of view.

In addition to the variation of results caused by differences in chemical composition and in physical condition of metals and alloys, it is known that observed creep characteristics are influenced by variations in testing procedure such as, for example, the duration of test, the method of loading or heating and the sensitivity of the extensometer system. It is, therefore, important that testing conditions should be standardized so that precise and repeatable measurements may be made and the results submitted by different investigators may be comparable.

INTRODUCTION

In this survey attention is mainly devoted to the treatment of the subject from an engineering point of view. In making this choice the author has been guided by the fact that by far the greater part of the published work on the creep of metals is concerned with the engineering aspect. In addition, the author's own work, in connection with the creep of aluminium alloys, is directly concerned with the application of these alloys to service under conditions in which creep is a factor to be considered, and it was felt that, in a survey of this nature, the subject should be reviewed, as far as possible, in the light of the author's own experience. An outline of the progress which has been made in the physical study of the mechanism of creep has been included, however, since work of this kind is of primary interest to all physicists whose object is to study the flow of metals for whatever reason.

CHAPTER I

CONDITIONS REQUIRED FOR THE MEASUREMENT OF CREEP

Introduction

IN STUDYING the creep characteristics of a metal, or alloy, attention is ultimately devoted to the establishment of a relationship between the stress to which the metal, or alloy, is subjected, the temperature to which it is exposed, the time for which the stress is applied and the deformation which takes place. The standard test is carried out by applying a fixed tensile load to a test piece kept at a fixed temperature; both the load and the temperature are maintained at a constant value for a long time, during which observations are made of the deformation of the test piece and of the rate of extension.

Considerable information is available on the various methods which have been used for the measurement of creep[1] and on the results which have been obtained by these methods. Much of the published work, however, describes investigations so lacking in control of experimental conditions that the results quoted must be regarded as unreliable for the purpose of both metallurgical theory and engineering design. This applies particularly to the work which has been carried out in Germany where the creep characteristics of metals, and alloys, are often assessed from tests of short duration and high rates of creep.

Investigations which are undertaken in order to obtain data for the purpose of engineering design involve the quantitative determination of creep rates whose order of magnitude is very low (0.1% strain per 100,000 hours is not an uncommon requirement). Such determinations are difficult and costly and, as a result of this, relatively few laboratories have succeeded in establishing a system of measurement which may be regarded as

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[1] For a collection of references to work describing methods used for the measurement of creep, see Appendix II on page 152.

satisfactory. Attempts have been made to set up alternative tests which are less costly and which involve less time. From the results of these "accelerated" methods, creep characteristics of metals and alloys have been estimated which are of doubtful significance when studied in the light of results obtained by reliable long-time methods (6), (7), (8), (9), (10), (11), (12), (13).

The difficulties which confront the student of creep become evident from a consideration of the complexity of the phenomenon, and of the experimental technique which must be acquired for its determination.

The method to be adopted in the measurement of such very small deformations presents an initial problem. In addition the tests are usually carried out at elevated temperatures, thereby involving the establishment of an accurate means of temperature control, since creep rate varies rapidly with change in temperature, and, furthermore, the thermal expansion of metals and alloys, which find application in service at elevated temperatures, is large in comparison with the creep rates which are acceptable.

It is mentioned on page xv that the creep test is perhaps the most sensitive of all mechanical tests, since the results obtained appear to be related to so many variables. In addition to the effects of such variables as have already been mentioned on page xv, many alloys suitable for high temperature applications are not always stable, either structurally or dimensionally, at the temperatures at which they are used (14), (15), (16), (17). It is obviously difficult to study the creep behaviour of metals and alloys if structural and dimensional changes, independent of stress, accompany long exposure to the temperatures in question. Experience has shown, particularly in connection with steels(5), that deformation, during the progress of a creep test, often does not settle down to a reasonably definite behaviour for hundreds of hours. Tapsell(18) has mentioned the fact that cases have been known in connection with the testing of steels where shrinkage has taken place initially in heated and loaded specimens; this shrinkage has lasted for several hours before elongation has set in as a result of creep rate exceeding contraction rate. Dimensional changes might also be expected in non-ferrous alloys. Kempf and Hopkins(17) examined the dimensional changes of a number of aluminium alloys on re-heating, in the temperature range 175° C.—230° C., following solution heat-treatment and found growth ranging from 0—0.16%, depending on the chemical composition. So that, even if the temperature can be controlled accurately and the deformation can be measured accurately, the complete creep behaviour cannot be studied unless the test is continued for a period of such length of time as to ensure metallurgical stability. The testing period most commonly adopted is 1,000 hours, this being, in most cases, approximately 1.0% of the expected service life. Some laboratories have adopted 3,000 to 5,000 hours as their standard testing time(6).

In the study of the creep characteristics of aluminium alloys the author has not observed such a marked degree of instability as is apparent in steels, and there is every indication that a testing period of 500 hours would yield reliable results. In view of this evidence, and of the fact that the expected life of aluminium alloy components in service at elevated temperatures is usually many times less than that of steel components, it is reasonable to attempt to standardize on a testing period of shorter duration than that adopted for steels.

Variations in creep results are often observed in specimens taken from the same batch of material and tested under identical conditions. Evidence exists(19), (20) which suggests that these variations may probably be attributed to submicroscopic differences in structure and to the presence of oxygen, nitrogen or other components not usually shown by chemical analysis. If this be the case, no amount of refinement of equipment and procedure can produce uniformly consistent results.

It is evident, then, that in carrying out an investigation of the creep properties of metals and alloys, considerable care must be devoted to the choice of test specimens. The choice will depend on the ultimate object of the investigation. If the purpose is that of verifying the validity of certain metallurgical hypotheses, then the material for test should be produced under very carefully controlled conditions, so that complete details are known of each stage of its manufacture; positive evidence should be available as to the absolute uniformity of the final product and of the test specimens taken from it. If, on the other hand, the purpose is that of obtaining data upon which engineering design may be based, then the test material must be taken from carefully selected samples which are representative of normal production; it should always be remembered, however, that the results obtained will not necessarily be typical, in the sense of being average, since no two batches or normally produced material are usually similar.

Since production specifications must permit variations within certain limits, the information gained from creep tests must provide a knowledge of the effects of these variations in order that a system of production control may be established on a basis of a balance between desirable properties and resulting costs. Furthermore, the testing programme should be such that a sufficient number of tests are carried out to make certain that the properties of the weakest material, acceptable under the specification, are known. This leads to a statistical study of the creep properties of materials in which the scatter of results is observed from a relatively large number of similar tests, for, unless this is done, it is necessary to assume a degree of uniformity among test specimens, and details of testing procedure, which it is impossible to realize.

An investigation which aims at providing sufficient information for a statistical analysis of creep results is an undertaking of considerable magnitude, for real problems are introduced by: (1) the length of time which must be devoted to each of the tests, (2) the production and calibration of furnaces and testing equipment, and (3) the maintenance of large numbers of control instruments. It is not surprising, then, that very little work along these lines has been possible, although the matter has received active consideration(21), (22), (23), (24). Under present conditions the solution to this difficulty can best be provided by as large a number of tests as possible being carried out in a number of laboratories, with parallel metallographic investigation of structure and careful analyses of the results obtained.

This raises the question of whether different laboratories can produce comparable results. The Joint Research Committee, on the Effect of Temperature on the Properties of Metals, of the American Society for Testing Materials and the American Society of Mechanical Engineers has studied this question and has laid down a code(25) as a tentative standard to be adopted in testing procedure. In order to form an idea of the adequacy of the code, samples, taken from a batch of uniform and stable steel, were supplied to several laboratories for creep measurements to be made. From the results returned by those laboratories which took the precautions laid down in the code(5), (26) a determination of the stress which would produce a creep rate of 10^{-6} inches/inch/hour at the end of a 1,000 hours shows that an agreement was reached within 3.5%.

General Requirements for Creep Testing

The fundamental requirements for a tension creep test (to which attention is limited in this survey) are as follows(23):—

(1) The provision of a means for applying an accurately measured stress, uniformly distributed over the cross-sectional area of the test specimen, which stress must be maintained constant.

(2) The provision of a means of accurately measuring the dimensional changes which take place in the length of the test section of the test specimen while it is maintained at constant stress and temperature.

(3) The provision of a means of heating the test specimen to the desired temperature with negligible variations throughout its mass. This temperature must be maintained with negligible fluctuations in the heating and cooling cycle and from day to day over long periods of time.

If these three requirements are met it is then possible to study the stress-strain-time-temperature relationship of a material, which should be independent of the details of testing equipment and procedure used to satisfy the requirements.

The amount of attention devoted to each separate requirement depends on the significance of the final results and the application to which they are to be put. Sufficient information may often be obtained by sacrificing, within reasonable limits, the complete fulfilment of one of the requirements; compromises of this nature are frequently necessary and justifiable as long as their limitations are not overlooked, and the interpretations of the results obtained are supported by results of tests carried out under conditions of the most rigid control possible.

The Application of Stress

The design of equipment for the application of stress to a creep test specimen is the easiest part of the testing technique. Little difficulty arises from the requirement that the load shall be applied axially in order to obtain uniform stress distribution over the cross-section of the specimen; the simplest, and most usual, methods of doing this use weights applied directly to spherical-seated specimen adaptors, or to a lever system acting on the adaptors. If, as is often the case, the spherical seatings are situated at the ends (away from the specimen) of long adaptors, it is difficult to eliminate bending of the specimen; this may be overcome by using a crossed knife-edge system in place of the spherical seating or, better still, by arranging for the spherical seatings to be situated between the ends of the specimen and its adaptors.

For tests which involve small deformations it is customary to neglect the small increase of stress which accompanies a decrease of cross-sectional area of the specimen as creep proceeds; for larger deformations, however, automatic compensation for this change of stress may be provided(4), (27), (28).

From an engineering point of view, whether, or not, it is desirable to confine attention to conditions of static loading is controversial, since, in service, the application of stress to components varies from static stress to cyclic impact with (to make matters worse) variable temperature. Information obtained from tests carried out under variable load conditions is obviously of use, and several attempts have been made to study this kind of stressing(29), (30); but while tests of this nature are useful, other factors, such as fatigue and damping, enter the field and make it difficult to access the role played by creep.

The Test Specimen

The chief concern in the design of the creep test specimen is that of ensuring that the stress distribution over the cross-sectional area of the gauge length is uniform. If the fillets of the test specimen are of small radius, and they are too near the ends of the gauge length, then variation in creep rates may occur due to stress concentration at the ends of the gauge length. This is overcome by making the radius of the fillets as large as possible and by increasing the length of the specimen so that the fillets are relatively far away from the ends of the gauge length.

Limitations in length of test specimen are imposed by the fact that it must be heated to a uniform temperature along its length; thus the relationship between specimen length and furnace length must be borne in mind. Furthermore, the gauge length should be such that, with the strain measuring apparatus available, the desired accuracy of measurement of deformation may be achieved.

There is no reason for choosing any one particular design of test specimen; it is advantageous for different laboratories to use the same type, and dimensions, of specimen in order to reduce possible variables to a minimum and thereby achieve more accurate comparison of results. Failing this the standard relationship(31) between surface area and cross-sectional area should be adhered to. The possibility of specimen size effect should not be neglected, and a study of this matter would be of considerable practical interest.

Lack of homogeneity of the material of the test specimen should naturally be avoided since this will give rise to stress concentration. The homogeneity of the material may be ascertained by careful metallographic and chemical investigation, and by careful examination for the presence of defects(32). Cast specimens intended to give typical results will, however, show minor discontinuities which have to be tolerated.

The machining operation in the production of the test specimen should be carefully controlled so as to reduce variations, due to surface finish, to a minimum. In addition, care should be exercised to avoid bending of the specimen, due to the readjustment of internal stress, during the machining operation; this is particularly important in the production of long specimens, and may be avoided by arranging for the specimen to be turned between centres on a lathe.

Extensometers

It is common practice among investigators, engaged on work which involves the measurement of strain, to design an extensometer which will satisfy their own particular requirements. Since the measurement of the variation of a number of physical phenomena may be employed for the purpose of strain measurement it is not surprising to find that many different types of extensometer have been used for the measurement of creep. It is proposed, here, to review the essential features of creep extensometers by referring, in some detail, to two types in common use, rather than to enumerate the various designs which have been employed and the principles upon which these designs have been based.

It has been pointed out on page 1 that the total allowable creep in engineering components is often of the order of 0.01% per year. Thus if daily readings are to be taken during the progress of creep at this rate the extensometer used must be capable of recording an increment of strain of the order of 3×10^{-7} inches per inch. This figure is based on the assumption of constant creep rate throughout the test; since, however, the stage of constant creep rate is preceded by a stage of relatively rapid, but decreasing creep rate, it is desirable that the extensometer should be capable of measuring strains considerably less than 3×10^{-7}.

The most direct means of introducing a magnification factor into the deformation increment is that of using as long a test specimen as is consistent with the fulfilment of uniform temperature along its length, and with the limitations imposed by the dimensions of the testing machine.

The measurement of small increments in length comparable with that mentioned above requires high magnification and is an application to which the optical lever principle is well suited. This principle forms the basis of the Martens' extensometer which has been modified (8), (33), (34), (35), (36) for work at elevated temperatures in a manner shown in Figures 1a and 1b.

The optical arrangement of this extensometer consists of a small, adjustable, stainless steel mirror, M (Figure 1a), which reflects an

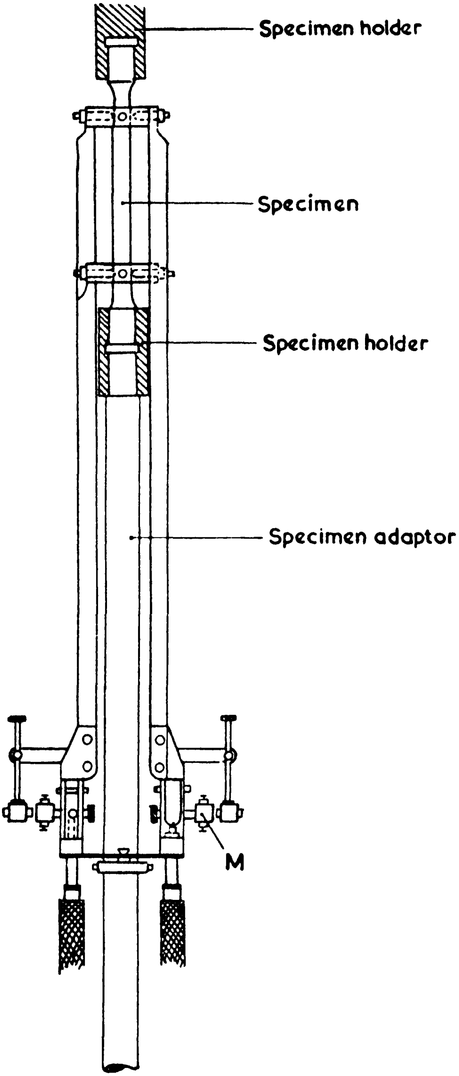


Figure 1a.
The modified design of the Martens' Extensometer for work at elevated temperatures (front elevation).

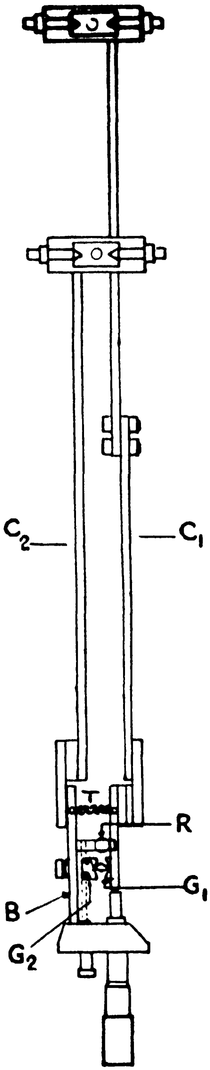


Figure 1b.
The modified design of the Martens' Extensometer for work at elevated temperatures(side elevation).

image of an illuminated scale into a telescope. The scale and telescope are mounted side by side on an optical bench which is situated at a distance of the order of 12 feet in front of the mirror. The mirror is carried by a hardened steel rhomb which is seated in the Vee-grooves G_1 and G_2 (Figure 1b). As extension of the test piece proceeds these grooves move relative to one another in a vertical direction, thus causing rotation of the rhomb and mirror.

The optical system must be outside the furnace, and to achieve this the comparison strips C_1 and C_2 (Figure 1b) are made long enough to extend through the bottom of the furnace; they pass down so that their adjacent surfaces are radial to the axis of the test piece adaptor.

At the upper ends, the comparison strips may be brazed either to a collar through which pointed locating screws pass and provide a means of anchoring to the gauge length of the specimen (in some designs(33) pointed locating pins are used, instead of screws, and the pins are sprung on the gauge length), or to split nuts which are clamped on to the screw ends of the test piece(36). The strip C_1 is anchored to the top, and the strip C_2 to the bottom, of the gauge length.

At the lower end, the comparison strip C_2 is adjustably fixed to a steel block B which is flexibly supported by spring steel strips which, in turn, are clamped to the adaptor by means of a heat insulating bush (Figure 1a). This mounting permits freedom for relative vertical movement without allowing lateral movement. The block B carries a micrometer screw gauge and the two Vee-grooves, G_2, which are formed by the crossed knife-edges of overlapping plates. (Only one groove G_2 is shown in Figure 1b, the other is behind, and in line with, this groove.)

The lower end of the strip C_1 is unconstrained except for the spring T (Figure 1b) which presses it against the roller R (Figure 1b). It carries the Vee-groove G_1 which faces the grooves G_2 and together these grooves form a three-point seating for the mirror rhomb. The grooves G_2 are attached to a slide, and adjustment of their vertical position is effected by traversing the slide by means of a screw, thus the rhomb may be reset when the amount of strain exceeds the available range.

In order to detect non-axial loading of the test specimen, the extensometer is duplicated on either side of it (Figure 1a). The comparison strips are made of heat-resisting steel, or an 80/20 nickel-chrome alloy, to reduce heat flow to a minimum, also to provide permanency of structure and to avoid scaling. In addition, mica spacers are fitted between the comparison strips C_2 and the steel blocks B, in order to reduce the heating of the blocks.

It is essential to ensure that the material from which the extensometer is made is in a stable condition, otherwise trouble is experienced due to dimensional changes with time. This may be overcome by subjecting the extensometer to a suitable stabilizing heat-treatment.

As a precaution against draughts the portion of the extensometer which extends below the furnace is enclosed in an asbestos box fitted with shutters, so that when readings are taken the shutters may be opened to expose the mirrors. In addition, it is advisable to lag the test piece adaptors in the vicinity of the furnace.

With this type of extensometer, Tapsell(33) claims that a strain of 8×10^{-7} inches per inch may be measured using a gauge length of 5 inches, that is, the extensometer is capable of measuring an increment of length equal to 4×10^{-6}. Thus, with the slight staggering of the plotted points, it is found in practice(36) that constant creep rates of 10^{-8} strain/hour can be determined with reasonable certainty.

The major criticism which may be levelled against this extensometer is that its weight is excessive. If the comparison strips are anchored to the enlarged screw ends of the test piece, then this criticism is relatively unimportant, but in this case the measurement of elongation includes deformation which may take place in the fillets and shoulders of the specimen where the stress distribution is not uniform. If, on the other hand, the comparison strips are anchored to the gauge-length by screws, or pins, the weight is important, for, in order to avoid slipping, the locating screws must grip the test piece very firmly; this usually involves (particularly in the case of non-ferrous metals) considerable penetration into the surface, and, after the care which is devoted to the production of a good surface, it is not good practice to allow the existence of such centres of stress concentration.

The author has been able to overcome this objection by the use of circular knife-edges instead of locating pins, and by the use of light tubular comparison strips. The two adjacent semi-circular knife-edges (forming the complete circle) are of greater diameter

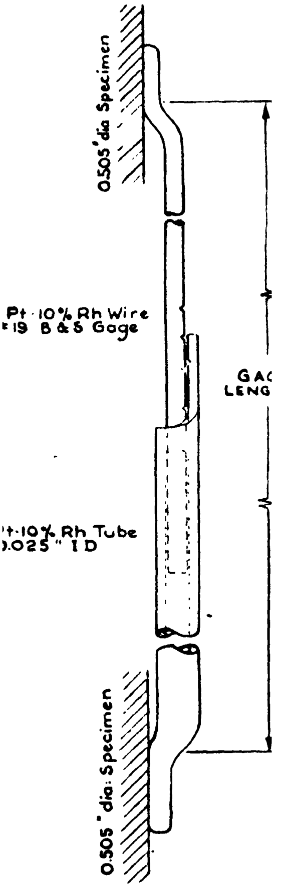


Figure 2a.
Schematic representation of Strip Extensometer. (Fellows, Cook and Avery, ref. 39.)

Figure 2b.
Extensometer attached to creep test-piece. (Fellows, Cook and Avery, ref. 39.)

than the test piece and are clamped on to the gauge length by means of screw clamps. There is a danger of rotation about the knife-edge contacts with the specimen, which may cause errors in readings, but experience has shown that, if care is taken in assembling the extensometer, errors due to this cause do not arise. (In any case the source of error may be eliminated by fitting horizontal locking strips.) The comparison tubes are attached to the knife-edge collars by means of flexible spring steel strips, and the remainder of the extensometer is similar to that which has been described, except that the blocks which carry the Vee-grooves are of light construction and the micrometer gauges have been eliminated.

Some investigators(37), (38), (39) prefer to use an extensometer which is wholly enclosed within the furnace. This type of extensometer is generally constructed by welding platinum comparison strips (or tubes) to the gauge length of the specimen in a manner shown in Figures 2a and 2b. The Vee-notch cut in the outer tube is arranged to be opposite a silica window in the furnace wall and is sighted by means of a microscope or telescope. The telescope is fitted with a filar micrometer eye-piece, one drum division movement of which is equivalent to an extension of the order 5×10^{-5} inches. The eye-piece is calibrated before use against a stage micrometer, which is used as a sub-standard.

In cases where it is not possible to weld the comparison strips, or tubes, to the gauge length, a comparison strip is wired securely to one shoulder of the test piece and the free end extends to the shoulder at the opposite end of the gauge length(38). Impressions are made with a diamond pyramid at the free end and on the shoulder of the test piece near the free end. The window in the furnace is arranged to be opposite the free end, and the microscope, or telescope, is focused on the impressions.

There are quite serious objections to the use of this type of extensometer. Apart from the difficulty involved in obtaining a satisfactory mounting, it does not provide a means of detecting bending of the specimen. Its use always involves the provision of a window in the furnace wall, the presence of which introduces a cold zone in the furnace in the vicinity of the gauge length of the specimen, a position where the most uniform temperature conditions should prevail. Compensation for this latter objection can be made by suitable spacing of the furnace windings, but this only adds to the difficulties of furnace construction.

The Temperature of the Test Specimen

The measurement of the temperature of the test specimen is perhaps the most important single measurement in connection with creep testing, because small errors, which are not easy to avoid, in the measurement of temperature can lead to large variations in results.

The thermocouple is the standard instrument used for the measurement of specimen temperature, even at temperatures below its range on the International Temperature Scale(40), (41), since, in order to detect gradients, observation of temperature at various points along the length of the specimen is required, and the thermocouple provides the most convenient means of fulfilling this requirement.

To achieve accuracy in temperature measurement by means of thermocouples, extreme care must be devoted to the calibrations of the couples, to the maintenance of the couples and their associated instruments, and to the attachment of the couples to the specimen.

The Calibration of the Thermocouples

The International Temperature Scale is almost universally accepted as the basis of reference for all temperature measurements, and the calibration of thermocouples should enable measurements to be made which agree with those obtained from the instruments defined in the appropriate ranges of the Scale. The most direct method of achieving this would be that of comparing the thermocouples with the primary instruments, but such a procedure is long, tedious and requires more apparatus than is justifiable; so other methods are adopted whereby the thermocouples may be compared indirectly with the primary instruments. The practice which is commonly adopted is that of determining the e.m.f. of the couples at a number of fixed points, either those used in defining the International Temperature Scale, or others whose values have been determined with primary instruments. A more satisfactory arrangement is for a few laboratories to possess the equipment for calibrating couples as working standards to yield the temperature of the International Temperature Scale, these working standards may then be used in the creep testing laboratories to calibrate the couples in use by a comparison method. This latter method is to be particularly recommended, since it is simpler to compare two couples than it is to compare two different instruments.

The platinum/platinum-rhodium thermocouple is the one which is chosen as a working standard for use outside the thermocouple range of the International Temperature Scale. Other couples may be calibrated by comparison with it almost as accurately as the calibration of the standard is known(42).

In creep testing the order of accuracy of temperature measurement required is 0.1° C.; thus agreement with the International Temperature Scale and methods of interpolation between the calibration points become matters of prime importance. For accurate work of this nature it is essential to follow, rigidly, certain prescribed methods and to take special precautions in the standard practice adopted in the calibration and testing of couples(42). This standard practice is discussed below.

THE HOMOGENEITY OF THERMOCOUPLES. In the first place it is important to establish the homogeneity of the thermocouple elements, since an inhomogeneous couple develops an e.m.f., characteristic of the hot junction, only when the inhomogeneous section of it is outside the regions of temperature gradient. Although no satisfactory method has been devised for the quantitative determination of homogeneity, certain well-known tests(43), (44) may be applied for detecting any variations in the properties of the thermocouple elements prior to use. These tests are rarely necessary on new wire, since manufacturers supply it to a high standard of homogeneity, but they are important when wire is being used again after a period in service; in view of this fact, it is advisable to discard base-metal couples, if these are used, after each creep test.

THE STABILIZATION OF THERMOCOUPLES. Before a thermocouple is calibrated it should be given a stabilizing, or annealing, treatment if this has not already been done by the manufacturer of the wire. In the case of base metal thermocouples the advice of the manufacturer should be sought as to the most suitable temperature for annealing and the length of time required at that temperature. In the case of platinum/platinum-rhodium couples the annealing treatment is carried out by heating the wires electrically, in air, whilst observations of their temperature are made with the aid of an optical pyrometer. There appears to be some doubt as to the optimum temperature for most successful annealing and as to the soaking period necessary, but the procedure has been standardized at one hour at 1,450° C.(42).

CALIBRATION AT FIXED POINTS. Table I (page 17) summarizes the fixed points which may be conveniently used for calibration. The values of these fixed points have been assigned, or determined, accurately(42) and from the measurements of the e.m.f.'s developed by thermocouples at the various points, a calibration relating temperature to e.m.f. may be established.

Whenever possible it is preferable, from the point of view of simplicity, to choose freezing points for calibration purposes, because a molten metal may be brought to a uniform temperature simply by stirring just prior to freezing, whereas, in the case of melting points, it is a matter of some difficulty to produce uniformity of temperature in a solid sample just prior to melting and, in the case of boiling points, the additional observation of pressure is necessary, the apparatus required is more complicated and considerably more skill is required to determine the points accurately.

The e.m.f. developed by a homogeneous thermocouple at the freezing point of a metal is constant and reproducible if all of the following conditions are fulfilled:—

1. The thermocouple is protected from contamination.
2. The depth of immersion of the thermocouple in the freezing point sample is sufficient to eliminate heating, or cooling, of

TABLE I

FIXED POINTS AVAILABLE FOR CALIBRATING THERMOCOUPLES FOR CREEP TESTING (TAKEN FROM ROESER AND WENSEL, REF. 42)

Thermometric Fixed Point	Values in the International Temperature Scale				Temperature of equilibrium (t_p) in °C. as a function of the pressure (p) between 680 and 780 mm. of Hg.
	Assigned (Primary Points)		Determined (Secondary Points)		
	°C.	°F.	°C.	°F.	
M.P. of ice	0.000	32.000	—	—	—
B.P. of water	100.000	212.000	—	—	$t_p = t_{760} + 0.0367(p-760) - 0.000023(p-760)^2$
B.P. of naphthalene	—	—	217.96	424.33	$t_p = t_{760} + 0.208(t_p+273.2) \log \left(\frac{p}{760}\right)$
F.P. of tin	—	—	231.9	449.4	—
B.P. of benzophenone	—	—	305.9	582.6	$t_p = t_{760} + 0.194(t_p+273.2) \log \left(\frac{p}{760}\right)$
F.P. of cadmium	—	—	320.9	609.6	—
F.P. of lead	—	—	327.3	621.2	—
F.P. of zinc	—	—	419.5	787.1	—
B.P. of sulphur	444.60	832.28	—	—	$t_p = t_{760} + 0.0909(p-760) - 0.000048(p-760)^2$
F.P. of antimony	—	—	630.5	1,166.9	—
F.P. of aluminium	—	—	660.1	1,220.2	—
F.P. of Cu–Ag eutectic alloy ..	—	—	778.8	1,433.8	—
F.P. of silver	960.5	1,760.9	—	—	—
F.P. of gold	1,063.0	1,945.4	—	—	—
F.P. of copper	—	—	1,083.0	1,981.4	—

M.P.—Melting point ; B.P.—boiling point ; F.P.—freezing point.

the hot junction by heat flow along the wires and protection tube.

3. The reference junction of the thermocouple is maintained at a constant and reproducible temperature.

4. The freezing point sample is pure.

5. The freezing point sample is maintained at essentially a uniform temperature during freezing, i.e., a very slow rate of cooling is used.

Contamination of thermocouples occurs when they come into contact with molten metals, or metallic vapours, or from the action of reducing gases at high temperatures. Porcelain, or " Pyrex " tubes, closed at one end, are generally used for protection, and the thermocouples should fit snugly into them. The elements of the couples should be insulated from one another by two-holed porcelain insulating tubes. Silica is not a good protection material, particularly at high temperatures (above 600° C.) because, in the presence of reducing gases, it is reduced to silicon which will alloy with thermocouple wire.

The depth of immersion in the freezing point sample can only be fixed by trial. It depends upon the material and size of the thermocouple wires, the dimensions of the insulating and protection tubes and on the difference in temperature between the freezing point sample and the furnace atmosphere surrounding it. The depth should be such that, during the freezing period, the couple can be raised or lowered at least one centimetre(42) from its normal position without altering the indicated e.m.f. by as much as the allowable uncertainty in the calibration.

The most convenient reference temperature is that of melting ice. An ice bath is usually made by filling a wide-mouthed thermos flask with ice shavings saturated with distilled water. The wires leading to the measuring instruments are connected at the reference junction so that two legs are formed which are insulated from one another by narrow glass tubes, closed at one end, which are immersed in the melting ice. Here again depth of immersion is important since it is essential that the reference junction should be exactly at the temperature of the melting ice. To make certain of this it is sometimes worth while to use wider glass tubes which also contain a mixture of ice and distilled water. The various parts of the reference junction equipment should be kept clean, particularly the interior of the glass tubes, since dirt collecting at the bottom of these tubes tends to set up galvanic action which may cause erroneous readings. The thermocouple wires and the lead wires should be insulated from one another right to the junction and the actual junction should be, as near as possible, a point connection.

PURITY OF THE FREEZING-POINT SAMPLES. The freezing point of a metal is dependent upon its state of purity. The presence of 0.1% impurities will, in some cases, lower the freezing point by 0.1° C.(42) and if alloying has occurred the freezing may take place over a range of temperature. Metals purchased for freezing point determinations should be 99.99% pure or, alternatively, a certificate stating the freezing point, should be issued by the manufacturer. It should be noted that tin and antimony have a marked tendency to supercooling during freezing and care must be taken to keep these metals well stirred as the freezing point is approached.

CRUCIBLE MATERIALS. Acheson graphite is most suitable for use as a crucible material. It is very pure, it can be machined into any desired shape and it can be used in contact with any of the freezing point samples listed in Table I without detectable contamination of the metals. At high temperatures the gases formed by its oxidation provide the reducing atmosphere usually necessary for the protection of the freezing point sample. Porcelain, or any material containing silica, should not be used for crucibles since silica is readily attacked by molten metals, particularly by aluminium and lead.

MELTING POINT DETERMINATIONS. The same equipment may be used for melting point determinations as is used for freezing points, but the method is rarely used, except in cases when the quantity of the fixed point sample available is small, as often occurs in the case of the gold point. In such instances the particular method(45), (46), (47) in which the sample is made to form the junction of the thermocouple is useful. The e.m.f. of the thermocouple is read at the instant of melting, i.e., just at the instant when the circuit breaks.

BOILING POINT DETERMINATIONS. Although boiling-points play an important part in the definition of the International Temperature Scale, they are seldom used in the calibration of thermocouples, except in the case of water. The steam point is realized experimentally by the use of a hypsometer so constructed as to avoid superheating of the vapour around the thermocouple and to prevent contamination of the thermocouple with air and other impurities. Details of hypsometers used for precision measurements may be found in the literature on the subject(48), (49) and, if proper conditions are attained, the observed e.m.f. of the couple will be independent of the rate of heat supply to the boiler, the length of time the hypsometer has been in operation and the depth of immersion of the couple. Care must be taken to shield the couple, for some distance from the junction, from radiation from the inner surface of the hypsometer. Observations of pressure must be made and corrections applied to give the boiling point at standard pressure (760 mm. mercury); the relationship between the temperature "T" and the vapour pressure "p" for the range 680 to 780 mm. of mercury is given by(42):—

$$T = 100.000 + 0.0367\ (p - 760) - 0.000023\ (p - 760)^2$$

The steam point, as realized by utilizing the condensing vapour in a hypsometer, is accurate to 0.01° C.(42).

CALIBRATION BY COMPARISON METHODS. The calibration of thermocouples by comparison with a platinum/platinum-rhodium thermocouple, used as a working standard, has as much claim to accuracy as any of the methods described above and it has the advantage that the time taken for the comparison is far less than that required for observation at the fixed points. The success of the method is dependent upon the hot junction of the couple, under test, being exactly at the same temperature as that of the standard, and this may be achieved by any of the following methods:—

1. Welding the junctions together.
2. Binding the junctions together with platinum wire or foil.
3. Inserting the junction of the standard in a hole drilled in the junction of the couple under test.
4. Inserting the couples, properly protected, in a bath of molten metal (e.g., tin).

5. Inserting the couples into holes drilled in a large block of metal so that the junctions are in good contact with the block and are as close together as possible. The block is heated in a muffle furnace.

When the junctions are in contact with one another they may be inserted in a furnace designed so that the temperature may be varied at a convenient rate and, if two potentiometers are available, simultaneous observations of the e.m.f.s of the two couples may be made by observing the two light spots from the respective galvanometers which are focused on a single scale. If readings can be taken on only one potentiometer, or if the junctions are not in contact, then the region surrounding the hot junctions must be brought to uniform temperature before each measurement of potential. As in the case of other methods of calibration, the reference junctions are maintained at a fixed temperature by immersion in melting ice.

In all comparison methods, particularly those in which the junctions are not in direct contact, it is important that the depth of immersion of the hot junction should be sufficient to eliminate heating, or cooling, of the junction by heat flow. When direct contact is made between the junctions, precautions must be taken against contamination of the standard; this is usually done by sealing it in a protection tube to within a few millimetres of the junction, then any contamination which takes place must be within a region of uniform temperature (the region around the junction must of necessity be uniform if the calibration is to be carried out with precision) and no error is introduced. After calibration is complete, the contaminated section of the standard can be cut off, more wire drawn through the seal and a new junction made.

INTERPOLATION BETWEEN CALIBRATION POINTS. After a thermocouple has been calibrated at a number of fixed points the problem of interpolation between these points arises.

If a curve of e.m.f. against temperature is plotted on a scale large enough to read the small intervals of temperature which are of interest in creep testing, then, in order to achieve accuracy in plotting, a large number of calibration points (at smaller and smaller intervals of temperature as the scale increases) must be taken; such a curve involves a considerable amount of work and is cumbersome when completed.

In the case of platinum/platinum-rhodium couples working in the temperature range 660° C. to 1,063° C. the International Temperature Scale prescribes that temperatures shall be derived from the relationship $e = a + bt + ct^2$ where " e " represents e.m.f., " t " temperature and " a," " b " and " c " are constants determined by calibration at the freezing points of gold, silver and antimony. If the freezing points of copper and aluminium are substituted for gold and antimony a scale of temperature is obtained which agrees with the International Temperature Scale to within 0.1° C.(42).

For temperatures outside the range 660° C. to 1,063° C. the method of drawing a graph has as much claim to accuracy as that of using an empirical equation. In order to reduce the number of uncertainties involved in interpolation on such a graph, and also to reduce the amount of work required for its construction, an empirical equation which fits the calibration points most closely can be used to plot a " different curve." From the empirical equation a table of temperature/e.m.f. values may be computed; if the differences between observed readings of the couple (taken, say, every 100° C.) and e.m.f. values in the table for the appropriate temperatures, are plotted against observed readings, the maximum differences to be plotted will be only a few degrees and the scale of plotting can then be conveniently opened out to give an accuracy more comparable with the accuracy obtained at the calibration points.

STABILITY OF THERMOCOUPLES. Employing the methods described above, and taking readings at intervals of about 100° C., the author has verified that in the calibration of base metal thermocouples (chromel-alumel and iron-constantan) the observed points in the temperature range 0° C.-400° C. may be determined with an accuracy of 0.1° C., in terms of the International Temperature Scale. For temperature ranges which extend above 400° C., however, the accuracy at observed points is of the order of 1° C. Interpolation between observed points introduces uncertainties which, using a difference curve, do not exceed ½° C. in the temperature range 0° C.-400° C. but which may amount to as much as 2° C. for large temperature ranges (e.g., 0° C.-1,000° C.).

During the progress of a creep test the calibration of base-metal thermocouples changes, the amount of change taking place being dependent upon the temperature of the test, the length of time devoted to the test and the gauge of the thermocouple wires. Results published on the stability of chromel-alumel and iron-constantan couples(50), (51) after exposure to a temperature of 600° C. for 1,000 hours show that the greatest change in calibration in the range 0-600° C. is about 0.5° C. At temperatures between 600° C. and 1,000° C. changes in calibration of the order of 1° C. may be expected.

In view of this evidence it is advisable that a calibration of the couples used in creep tests should be carried out before and after each test, and that new couples should be used for each test.

From experience gained in working at temperatures up to 300° C. the author has found that, as compared with the reading which would be obtained with a platinum/platinum-rhodium sub-standard, the overall accuracy which may be obtained in the measurement of the temperature of the test specimen with chromel-alumel couples is within ¼° C.

The Measurement of the E.M.F. of Thermocouples

The accurate measurement of the e.m.f. of thermocouples, used for creep tests, should be carried out by means of a potentiometric method. In view of the smallness of thermo-electric e.m.f.s, care must be exercised in the choice of potentiometer, since appreciable errors may be introduced by voltage drop in the variable resistance of switches and by thermo-electric (parasitic) disturbances due to:—(1) contact of dissimilar materials, (2) inequalities of temperature within the potentiometer itself and (3) the manipulation of sliding contacts and switches. The potentiometer circuit, the design of the switches and the metal employed in the construction of coils and terminals should be chosen so as to reduce these sources of error to a minimum.

The development of a suitable potentiometer for the accurate measurement of thermo-electric e.m.f.s was begun and completed between the years 1905 and 1908. In 1908 Diesselhörst(52) constructed a potentiometer which incorporated features of design suggested by Hausrath(53), White(54) and himself(55) and which was so successful that it is still made for use in laboratories where the accurate measurement of low potentials is required. The Diesselhörst "Thermokraftfrei" potentiometer has been followed by more recent devices(56), (57), (58) but it is doubtful whether these devices can lay claim to any greater accuracy.

A circuit diagram of the Diesselhörst potentiometer is shown in Figure 3. The apparatus consists of two units: the potentiometer unit itself, and the current regulator unit. The latter unit serves the purpose of standardizing the current, which is fed through the potentiometer circuit via the switch contacts T_1 and T_1', at 0.001 ampere. This is done by balancing the voltage drop across the resistance g g^1 against the e.m.f. of a standard cell.

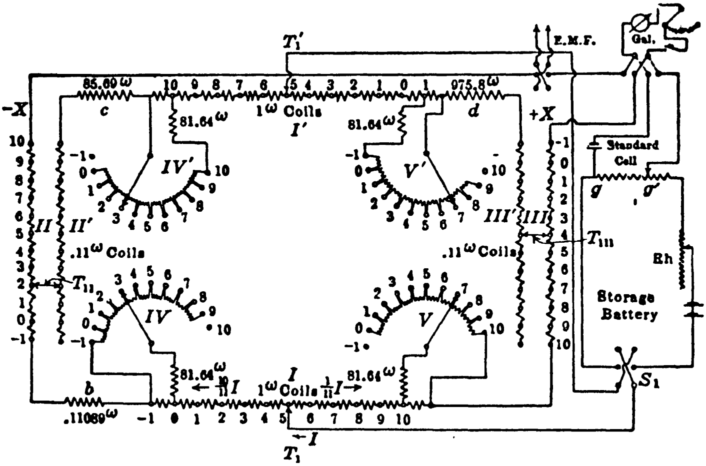


Figure 3.

Circuit diagram of the Diesselhörst Potentiometer. (Laws, ref. 59.)

The potentiometer circuit consists of a divided network in which the difference in the potential drops along the two branches, measured from the point at which the current enters (T_1) is used to balance the unknown potential to be measured.

There are ten decades of coils (marked I I′, II II′, III III′, and so on). The ends of each of the coils in each decade are connected to studs which form contacts for five rotary switches. Thus each of the pairs of decades: I I′, II II′, III III′, and so on, are operated by a single two-pole switch rotor such that contact is always made with studs of the same number (see Figure 3) in each of the pairs of decades.

The values of the resistance of the coils are given in the following table(59):—

Decades	Coil Nos.	Resistance
I and I′	All coils	1 ohm each
II and II′ III and III′	All coils	0.11 ohms each
IV and V′	Between −1 and 0	8.264 ohms
	Between 0 and 1	10.101 ohms
	Between 1 and 2	12.626 ohms
	Between 2 and 3	16.234 ohms
	Between 3 and 4	21.645 ohms
	Between 4 and 5	30.30 ohms
	Between 5 and 6	45.41 ohms
	Between 6 and 7	75.80 ohms
	Between 7 and 8	151.51 ohms
	Between 8 and 9	454.54 ohms
	Between 9 and 10	Infinity
IV′ and V	Combination of coil resistances the same as in IV and V′, but the numbers are reversed	

The decades IV, V and IV′, V′ with their series resistances, each of value 81.64 ohms, are shunted across 1 ohm coils in the decades I and I′ respectively, as shown in Figure 3.

A study of the circuit reveals the following characteristics:—

1. The fixed resistances: b, c, d, are chosen such that the currents in the left-hand and right-hand branches of the network are in ratio of 10 : 1.

2. The construction of the switches and the values of the coil resistances are such that the resistances of the two paths between the contacts T_I and T_I' are independent of the positions of T_I T_I', T_{II} T_{II}', and T_{III} T_{III}'.

3. When the sliding contacts in decades IV, IV′ and V, V′ are on studs No. 0, then the effective resistances of each of these decades, on the circuit is as follows:—

Decade IV	0.9890 ohms.
Decade IV′	0.9989 ohms.
Decade V	0.9989 ohms.
Decade V′	0.9890 ohms.

4. A movement of one stud position (in the forward direction) of the switch operating decades IV and IV′ increases the resistance between contacts T_1 and T_{11} by 0.0011 ohms and decreases the resistance between contacts T_{11} and T_1' by the same amount. A movement of one stud position (in the forward direction) of the switch operating decades V and V′ decreases the resistance between contacts T_1 and T_{111} by 0.0011 ohms and increases the resistance between contacts T_{111} and T_1' by the same amount.

5. Since the potential drop between the points + X and − X is equal to the difference between the P.D.s from T_1 to + X and T_1 to − X, then, if n_1, n_2, n_3, and so on, are the settings of the graduated dials, fitted to the five rotary switches, required to balance the unknown potential, the measure of the unknown potential is given by:

$$0.001 \left\{ \frac{10}{11} [n_1 \times 1 + 0.9890 + 0.0011n_4 + 0.11089 + 0.11(n_2 + 1)] \right.$$
$$\left. - \frac{1}{11} [(10 - n_1)1 + 0.9989 - 0.0011n_5 + 0.11\,(10 - n_3)] \right\}$$
$$= 0.001 \left\{ n_1 + \frac{n_2}{10} + \frac{n_3}{100} + \frac{n_4}{1000} + \frac{n_5}{10000} \right\} \text{ volt.}$$

6. The maximum variation of the galvanometer circuit resistance is about 0.3% thus the potentiometer may be used as a deflection potentiometer and, from observations of the galvanometer deflection, estimation may be made of the voltage down to the next place of decimals beyond the reading of dial No. 5. A further advantage is that the galvanometer damping remains constant.

Parasitic Disturbances in the Diesselhörst Potentiometer

CONTACT RESISTANCE IN SWITCHES. In a well made switch the order of variation in resistance, due to manipulation, is about 0.0002 ohm(59).

The contacts T_1 and T_1' are in the battery circuit which has a resistance of approximately 2,000 ohms. (This may be seen from a study of the network.)

The double contacts T_{11} and T_{111} are in circuits of resistance greater than 85 and 975 ohms respectively.

The sliding contacts of decades IV, IV′, V and V′ are in circuits of at least 90 ohms (approximately) which circuits are in parallel with coils of 1 ohm resistance. Thus a variation of resistance of the order of 0.0002 ohm in each sliding contact would produce a variation in the circuit resistance of the order of $0.0002\left(\frac{1}{1+90}\right)^2$ ohm.

It is seen from these figures that variations in resistance due to the manipulation of switches have a negligible effect on the readings of the potentiometer.

THERMO-ELECTRIC DISTURBANCES. In a well made switch the order of magnitude of parasitic e.m.f.s is 0.000001 volt(59).

Parasitic e.m.f.s arising from the manipulation of contacts T_1 and T_1' are in series with the battery (2 volts).

If the parasitic e.m.f.s arising in contacts T_{11} and T_{111} have a maximum value represented by " e " then the maximum variation in potentiometer reading will arise when these e.m.f.s are in opposition and the change in potential between the points $+X$ and $-X$, due to this cause will be:—

$$\frac{\pm e(A+B)}{A+B+C+D}$$

where A = resistance between T_1 and T_{11}; B = resistance between T_1 and T_{111}; C = resistance between T_{11} and T_1'; D = resistance between T_{111} and T_1'.

The maximum value of (A+B) is 14.42 ohms and the value of A+B+C+D is 1,089 ohms.

$$\therefore \frac{e(A+B)}{A+B+C+D} = e \times 0.013$$

Thus the error introduced is only 1.3% of e.

If the parasitic e.m.f.s arising in the contacts of each of the decades IV, IV′ and V, V′ have a maximum value of e′, consider the position of the switch contact such that there is least resistance around each decade circuit (i.e., contact is made with stud No. −1 in decades IV and V′ and No. 10 in decades IV′ and V). Then the e.m.f. e′ is in series with 81.64 ohms and a 1 ohm coil which is shunted by a resistance greater than 1,000 ohms. The potential drop produced is therefore approximately equal to:—

$$\frac{e'}{82.6} = 0.012e' = 1.2\% \text{ of } e'$$

From these considerations it is apparent that errors due to parasitic e.m.f.s are negligible and that the potentiometer is well suited for application to the measurement of small electromotive forces.

The Welding of Thermocouple Junctions(60)

There are five methods whereby a thermocouple junction may be made for use in creep testing; they are as follows:—

1. Soldering or brazing the wires together.
2. Butt welding of the wires.
3. Oxy-gas welding of the wires.
4. Resistance welding of the wires.
5. Arc welding of the wires.

The first method produces thermocouples of reasonable reliability and permanence but the maximum temperature to which the junction may be heated is, of necessity, lower than the melting point of the solder.

The second method is reputed to give extremely good joints(60); the two wires are clamped in a jig and each is connected to one terminal of a charged electrolytic condenser, one wire is then made to approach the other at a high velocity and discharge takes place on contact, thus effecting a weld. The equipment and technique required for this process is of a specialized nature.

The third method is not a good one, particularly when applied to fine wires. It can be used only on wires which are resistant to oxidation and which do not react chemically with the constituents of the flame. There is always a tendency for the wires to weaken just behind the weld and the process has nothing to recommend it, since far better welds can be produced on a much wider range of couples by other methods.

The fourth method produces good welds which are permanent at the highest temperature the component wires will withstand. The apparatus used for the process must be kept very clean to avoid contamination of the junction.

The fifth method is the most convenient. By its means robust junctions may be produced whose contact is very permanent. The wires are held together, side by side, and they form one electrode of the arc. The process should take place in a reducing atmosphere, preferably hydrogen, and care should be taken to avoid contamination of the weld either from the other electrode or from the atmosphere. Platinum thermocouples are particularly susceptible to the presence of iron and carbon.

Attachment of the Thermocouples to the Specimen

At least three thermocouples should be attached to the specimen, one at each end and one in the centre. If the readings taken by these couples are reliable and they differ from one another by not more than 1° C. then the temperature distribution along the length of the specimen may be regarded as near uniformity as it is possible to achieve, under normal testing conditions.

Thermocouples may be attached to the specimen by the following methods:—

1. Hammering the junction into a small hole drilled in the specimen.
2. Welding the junction to the specimen.

3. Threading the junction into a socket drilled into the specimen.
4. Clipping the junction on to the surface of the specimen by means of a spring clip and metal shield.
5. Tying the junction on to the surface of the specimen with asbestos string.

Hammering the junctions into small holes suffers the disadvantage that, during the process, the junction and wires near it are unavoidably deformed, or cold worked, with a result that the calibration is changed; in addition, attachment only to the shoulders of the specimen may be made in this manner, since the surface of the gauge length must be unimpaired to avoid the existence of stress raisers.

The welding of thermocouples to the specimen suffers similiar disadvantages. The process must be limited to the shoulders of the specimen and, again, it introduces the possibility of a change in calibration during the welding.

The attachment of couples by threading the junctions into sockets drilled into the specimen must also be limited to the shoulders. The method is reliable, however, provided that care is taken to ensure that the couple junctions fit tightly into the end of the socket and that the two wires, or elements, of the couple are insulated from one another as far as the junction. More reliable results seem to be obtained if the couples are threaded into sockets in the manner shown in Figure 4a rather than in that shown in Figure 4b.

The method of tying, or clipping the couples on to the surface of the specimen, is the only way of attaching them to the centre of the gauge length, and provided certain precautions are taken, which are described below, reliable readings can be obtained.

The elements of the thermocouples should be of small gauge to reduce the rate of heat flow away from the hot junction. The elements should be insulated from one another right to the hot junction, and care should be taken to shield this junction, and the wires, from direct radiation from the furnace wall. In addition, the elements of the couple should be led out of the furnace in a direction along the length of the specimen and as near to it as possible; this avoids passage through temperature gradients near the hot junction.

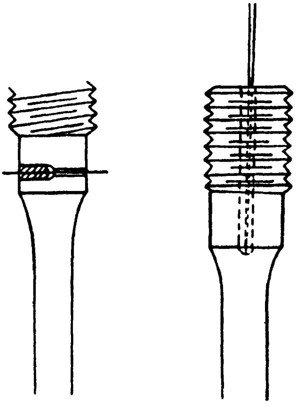


Figure 4a. Figure 4b.
Methods of attachment of thermocouples to the test-piece. 4a gives more reliable readings than 4b.

The recorded temperature of points along the length of the specimen should be independent of the method of attachment of the thermocouples. As will be seen, it is doubtful whether such a condition can be achieved, and the only way of acquiring any degree of confidence in the temperature observations is by running a number of trials in which the couples are attached by different methods. A common practice[61] is that of assembling a dummy specimen in the testing machine prior to each test. The design of the dummy is shown in Figure 5. The thermocouples A, B, C, D, E and F, are tied on and record the surface temperature of the shoulders and centre of the gauge length; thermocouples G, H, K and L, are threaded into sockets and record the internal temperatures of the shoulders and the centre of the gauge length. If the group readings of the individual thermocouples forming each of the groups: A, B, G, H, and C, D, K, L, and E, F, agree within, say, 0.1° C. it may be assumed that the recorded

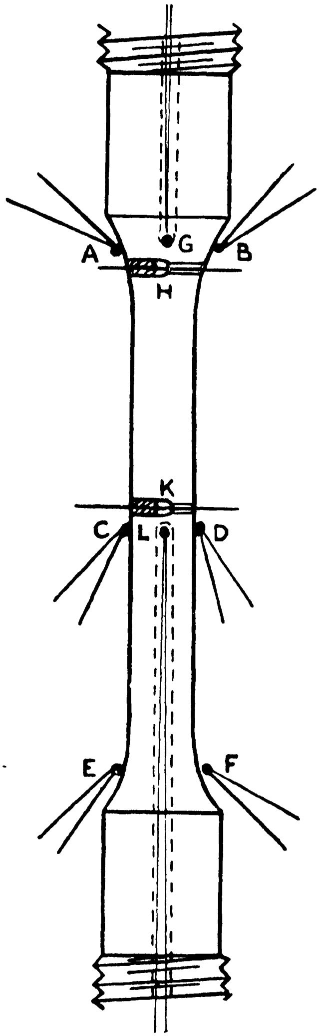


Figure 5.
The attachment of thermocouples
to a dummy test-piece.

temperature is independent of the method of attachment. If the readings between groups agree within 1° C. then the distribution of temperature along the length of the dummy may be regarded as satisfactory.

In actual practice it is a matter of some difficulty to obtain agreement between the readings of the thermocouples forming a group, and often involves several adjustments of the attachment; for example, thermocouples tied on loosely rarely agree with those tied on tightly, and thermocouples threaded into the type of socket at H and K rarely agree with those threaded into the type of socket at G and L. The closest agreement usually exists between the couples H and K and those which are tied on tightly, provided the latter make good contact with the surface and are well shielded from direct radiation from the furnace wall.

When the dummy is removed and replaced by an actual test specimen, the degree of contact in the screw threads of the specimen holders may change, thereby altering the rate of heat flow and, consequently, the temperature distribution along the specimen. Unless, then, a similar comparison to that described above can be made, an element of doubt must always exist as to whether the correct readings of temperature are being recorded. Thus, to eliminate this doubt, a number of trials should be carried out in which the dummy is taken out of the machine and put back again and, from a comparison of the various temperature readings taken, an estimation of the uncertainty involved can be evaluated.

In view of the difficulties which have been described, it can hardly be said that recorded temperatures are independent of the accepted methods of attachment of the thermocouples, even when considerable care is devoted to the process. It appears, therefore, that if strict comparison of results of creep tests submitted by different laboratories is to be made possible, a specified method of attachment should be universally adopted. The most favourable method is that of tying the couples to the surface, for no other alternative can be used at the centre of the gauge length, and, furthermore, it may be argued(51) that the region of the furnace enclosing the specimen is of uniform temperature, therefore, thermocouples placed anywhere near the specimen will record the correct temperature. Actually, the region enclosing the specimen can never be made completely uniform in temperature but, if the thermocouples are bound tightly on to its surface the error in readings should be negligible. As an additional safeguard, a specification for the attachment of thermocouples should state precisely how they are to be tied on and the method chosen should be based on concrete results from a series of tests.

Furnace Equipment

THE FURNACE

The design of the creep furnace is based on the necessity for obtaining and maintaining uniform temperature along the gauge length of the test specimen. This having been achieved, an additional requirement is that of minimizing variation in temperature due to the cyclic action of auxiliary temperature control equipment and to the variation of conditions external to the apparatus, i.e., the day-to-day changes.

Uniform temperature distribution along the gauge length of the specimen is usually effected by grading the spacing of the furnace windings and by choosing a furnace sufficiently long for it to be possible to produce a region of uniform temperature in its centre zone.

There is no means, other than that of experience, for predetermining the correct grading of the furnace windings for the production of a zone of uniform temperature, and success is accomplished only after a series of trials. Each trial involves dismantling the whole of the furnace assembly, a tedious task which rarely meets with complete success. It is more satisfactory, therefore, to obtain a centre zone of an approximate uniform temperature distribution (this can by done by grading the winding into three zones, the two outer zones being more closely wound than the central zone) and then to arrange for a fine adjustment to be made by means of external variable resistances which are tapped across sections of the furnace winding. The provision of tappings to the furnace winding provides a flexibility of control which is necessary in creep testing since, no matter how much care is taken in assembling the apparatus, conditions vary from test to test. For example, the rate of heat flow from the specimen to the holders (through the screw threads) is an uncontrollable quantity, and the lagging used for plugging the ends of the furnace can never be repeated exactly in any two tests.

A refinement which is often worth while, particularly in laboratories where the atmospheric temperature is not controlled, is that of winding small heaters on to the specimen adaptors, just above and just below the furnace, and arranging for the variations in air temperature to control the power input to these heaters so that the temperature of the adaptors in the vicinity of the furnace is kept constant.

Nichrome wire is most commonly used for the furnace winding and, provided that a suitable gauge of wire is chosen, and that the temperature of the winding is not allowed to go above that recommended by the makers of the wire, the need for replacement arises very rarely. If an A.C. supply is used it is advisable to eliminate the presence of the alternating magnetic field by making the winding non-inductive.

The choice of material for the furnace tube is governed by the working temperature. In the testing of aluminium alloys where the maximum test temperature does not exceed 400° C. it is preferable to have a metal furnace tube and to insulate the windings from it by a thin layer of insulating material such as asbestos paper, mica or insulating cement. (With such a construction the use of an electric supply of low voltage to earth is to be recommended as a safety precaution.) A metal tube has the advantage that any uneven temperature distribution, or hot spots, in the central zone of the furnace caused by faults in the winding are, to some extent, smoothed out by conduction of heat along the tube. For the testing of steels the working temperatures are often such that furnace tubes must be made of silica or alundum; when this is the case it is advantageous (for reasons just given) to have a metal liner which is easily replaceable. The use of silver liners has been reported(22); if the expense of such liners can be afforded there seems no reason for the presence of a silica tube since this only introduces a poor heat-conducting material between the windings and the interior furnace wall, which is a disadvantage from the point of view of temperature control, as will be seen later.

The furnace tube is usually mounted in a relatively large outer metal casing which is, preferably, polished to reduce any heat loss by radiation. The space between the tube and the casing is filled with material of good heat insulating properties, such as powdered silocel, and the dimensions of the casing are such that the thickness of the heat insulating material reduces the lateral heat flow to negligible proportions.

THE CONTROL OF FURNACE TEMPERATURE

The maintenance of uniform temperature of the creep test specimen over long periods of time becomes increasingly difficult as the allowable temperature variation is decreased. Starting with a knowledge of the allowable scatter of points on the creep curve, and the coefficient of linear expansion of the material under test, it is possible to compute approximately the allowable temperature variation.

Owing to the difficulties entailed in adequate temperature control, consideration has been given to the alternative of accepting small temperature variations and applying corrections to the observed results(23). On account of the uncertainties which must be introduced by the application of such corrections it is generally considered advisable to avoid them as far as possible.

The degree of temperature control which can be attained is dependent upon the sensitivity of the auxiliary automatic temperature control equipment used, as well as upon the design of the furnace. Good control should maintain the recorded temperature to within ± ¼° C. throughout a test, and the choice of auxiliary equipment for achieving this degree of control may be made from a wide selection of apparatus which has been invented for the purpose(62). When due consideration has been given to the merits of apparatus for application to particular testing conditions the choice becomes, largely, one of personal preference.

In order to control the temperature of a furnace, the first requirement is a form of thermometer the action of which depends on the variation of some suitable physical property with temperature. The magnitude of variation of the property must be such that, by its aid, changes in temperature within the allowable limits may be detected. The effect produced by the variation of the property must be such that, by suitable amplification, it may be used to control, automatically, the power supply to the furnace in such a manner that the system, as a whole, possesses a sensitivity adequate for combating all temperature changes greater than those which mark the allowable limits.

The thermometers, or sensitive elements, of temperature controllers most commonly used in creep testing depend on the three following types of physical change:—

1. The change in the e.m.f. developed by a thermocouple when the temperature of one junction is raised above that of the other.
2. The increase in electrical resistance of a metal (in the form of wire) as its temperature is raised.
3. The linear expansion of a metal when its temperature is raised.

CONTROL BY THERMOCOUPLES. The e.m.f. developed by a thermocouple may be applied directly to a galvanometer system the deflection of which actuates a relay controlling the power supply to the furnace(62) or, the out-of-balance voltage of a potentiometer system, connected to the thermocouple, may be applied to a relay(62).

The choice of thermocouple should be such that:

1. The e.m.f. continues to increase with rise in temperature throughout the range of testing temperature.
2. The e.m.f./temperature relationship is as near linear as possible throughout the range of testing temperature. This condition simplifies the construction of the remainder of the auxiliary equipment.
3. The thermo-electric power is adequate for providing the sensitivity required.
4. The elements of the couple neither volatilize nor oxidize excessively when exposed to elevated temperatures.
5. The thermo-electric power remains constant when the couple is exposed to elevated temperatures for a long time.
6. The heat capacity of the hot junction and wires of the couple is negligible.

The use of a thermocouple as sensitive element of a temperature controller involves controlling the temperature of the furnace by observations made at one point of the heating system; while the error involved in doing this is likely to be small in a carefully constructed furnace, it is preferable, in the author's opinion, to forgo the accuracy of control of one point and adopt a method whereby the sensitive element of the controller is exposed to the whole of that part of the heating system which surrounds the test specimen.

CONTROL BY RESISTANCE THERMOMETERS. In the author's opinion the resistance thermometer is the most favourable instrument for use as the sensitive element of a temperature controller in creep furnaces. It may readily be incorporated in electronic circuits and, thereby, very small changes in temperature may be detected and amplified to such a degree that an extremely sensitive system of automatic control may be established(63); furthermore, the element may be designed so that it is exposed to any proportion of the heating system of the furnace.

The choice of wire for the resistance thermometer should be such that:

1. The resistance increases as near linearly as possible with temperature throughout the range of testing temperatures. (This condition is not essential but the remainder of the auxiliary equipment is of simpler construction if it is fulfilled.)
2. The temperature coefficient of resistance fulfils the required conditions of sensitivity.
3. The resistance, as measured at room temperature, remains constant after exposure to elevated temperatures for a long time.
4. The temperature coefficient of resistance remains constant after exposure to elevated temperatures for a long time.
5. The heat capacity of the thermometer is negligible.

CONTROL BY MECHANICAL METHODS. The use of the expansion of a metal as a means of detecting changes in temperature has the advantage that, of the methods described, it provides, up to the highest testing temperatures, the most linear relationship with temperature. It is usual to employ the expansion of the metal furnace tube, or the expansion of a wire mounted in the furnace, for actuating the power supply relay system; this is accomplished by mechanical amplification of the change in length of the tube or wire, and if mechanical amplification is preferred, it can be made to yield results which are as good as those obtained by electrical means. Using such a method, control of creep test temperatures to within ± ½° C. has been reported(36).

THE RELAY SYSTEM. The relay system is usually arranged so that it exerts an "On and Off" control on the power supply to the furnace. There are two positions of the control; in the "On" position the temperature attained is too high and in the "Off" position it is too low, so that a permanent oscillation takes place about the mean required temperature, the amplitude and frequency of which depends on the characteristics of the furnace and on the extent of variation in conditions caused by the controller.

In the "On and Off" system of control small deviations of temperature cause the same movement of the relay as do large deviations and, in order to maintain stability it is sometimes necessary to sacrifice sensitivity. To overcome this difficulty a system of "Proportional Control" may be adopted in which the "On and Off" system is modified so that the power supply is regulated by the extent of temperature deviation, i.e., the change in power supply to the furnace is proportional to the temperature deviation(62), (63).

CONDITIONS FOR GOOD TEMPERATURE CONTROL. The conditions which must be fulfilled in the design of heating apparatus for the maintenance of constant temperature of a creep test specimen may be summarized as follows:—

1. The furnace should possess high thermal capacity so that it is slow to respond to changes in power supply and to variations in external conditions.
2. The sensitive element of the control device should be located as near as possible to the furnace heating coil so that there is a minimum time lag between changes in furnace element temperature and the detection of the changes by the sensitive element of the controller. In this connection the thermal capacity of the sensitive element should be kept as low as possible. (In actual practice the author has found, with the use of metal furnace

tubes, that best results are obtained when the sensitive element is placed against the wall of the tube inside the furnace.)

3. The control equipment should possess maximum sensitivity without producing self-sustained oscillation.

4. The time lag for the transfer of heat from the furnace winding to the furnace wall should be small so that changes in furnace wall temperature quickly effect the flow of heat from the windings and, conversely, changes in temperature of the windings rapidly influence the furnace wall temperature.

5. The internal diameter of the furnace tube should be kept as small as possible in order to minimize the time lag for the transfer of heat from the furnace wall to the specimen and to minimize the temperature gradient, which must always exist between the furnace wall and the specimen unless the heat flow from the specimen along the adaptors can be stopped. By keeping the diameter of the furnace tube small, the specimen temperature is more nearly equal to that of the furnace wall and, furthermore, variations in specimen temperature, due to changes in external temperature conditions which cause variable heat flow along the adaptors, affect the furnace wall temperature in minimum time and thereby affect the control system in minimum time.

6. The specimen adaptors should be made from material with as low thermal conductivity as possible in order that the rate of heat flow from the specimen shall be as small as possible.

7. The ends of the furnace should be carefully lagged to ensure against loss of heat by convection.

CHAPTER II

THE CREEP CURVE

General Characteristics

THE EXPERIMENTAL requirements for the study of the creep characteristics of metals and alloys have been discussed and it has been pointed out that the normal test involves the application of a fixed static load to a test specimen maintained at a fixed temperature, and the measurement of the deformation at regular intervals over a long period of time. In order to establish a relationship between the four variables: stress, strain, temperature and time, a series of such tests must be carried out employing different stresses and temperatures. From the results obtained two families of deformation/time curves may be drawn, one showing the curves obtained with different stresses (the temperature

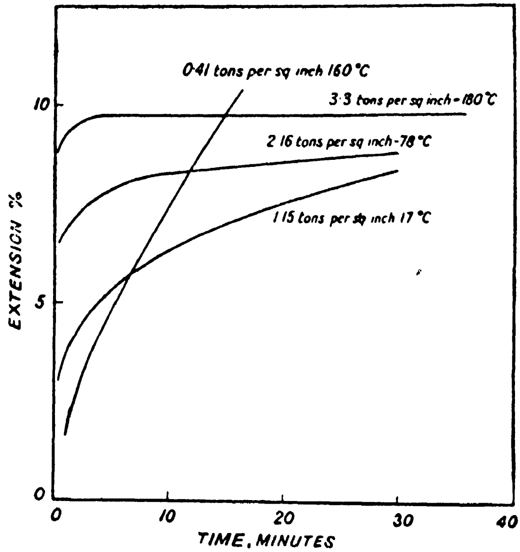


Figure 6.

Creep curves for Lead at different stresses and temperatures. (Andrade, ref. 4.)

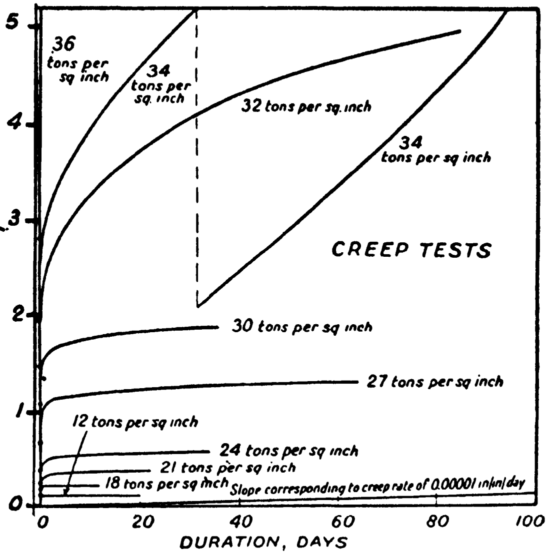


Figure 7.
Creep curves at different stresses for 44-ton steel at 300°C. (Tapsell, ref. 8.)

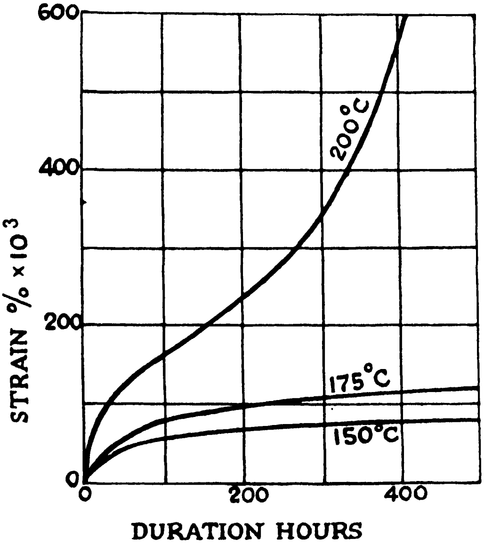


Figure 8.
Creep curves at different temperatures (stress 7 tons sq. in.) for a duralumin-type alloy. (Stanford.)

remaining constant) and the other showing the curves obtained with different temperatures (the stress remaining constant).

The form of these deformation/time curves has been the subject of considerable investigation in the past and it has been shown(8) from experimental data that both families display a common form, the characteristics of which are present at all stresses and temperatures, although the degree of each characteristic varies with stress and temperature for any particular metal or alloy. Figures 6, 7, 8 and 9 show curves which have been plotted from actual experimental results, and from an examination of these experimental curves it is possible to draw an ideal schematic family of curves(64) as they might appear were they plotted from results of tests lasting several years. Such a family of curves is shown in Figure 10, and they depict, from bottom to top, the different

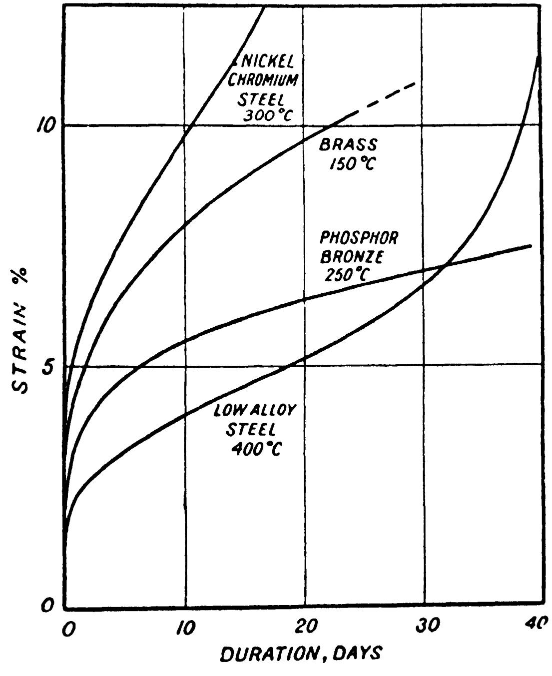
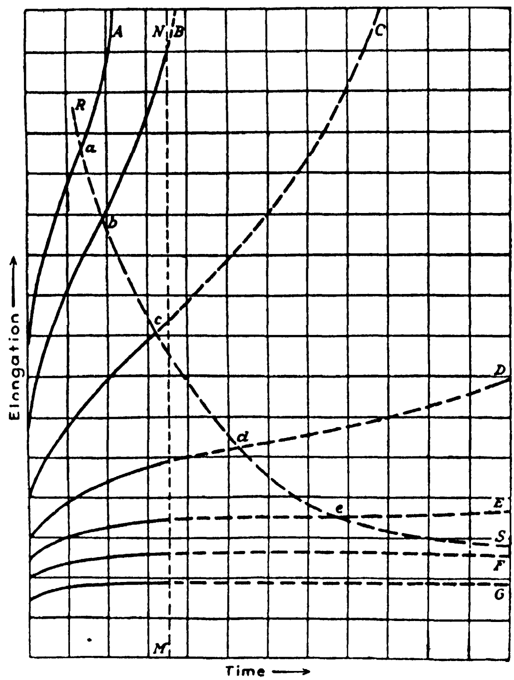


Figure 9.

Creep curves for different materials tested at different temperatures and stresses. (Tapsell, ref. 8.)

curves obtained with increasing loads at one temperature, or, equally well, increasing temperatures at one load. The vertical line MN represents the customary length of time of the ordinary creep test and the spacing between the curves, for given increments of load, or temperature, is a property of the material under test. It is seen that as stress and temperature are varied individually from high values to low values the deformation/time curve changes from that of indicating rapid extension to breaking, as in curve A (Figure 10), through the intermediate curves B, C, D, etc., to the curve G which shows a continually decreasing rate of deformation throughout the testing period.



EXTRAPOLATED CREEP CURVES

Figure 10.
Schematic family of creep curves.
(McVetty, ref. 64.)

The characteristics of the normal creep curve may be exemplified from the curve shown in Figure 11, which illustrates the gradual-deformation of a metal, by creep, to breaking within the testing period. The curve is divided into four stages. The first stage represents the initial deformation on loading and is given by the ordinate at zero time; this initial deformation may be composed of an elastic (reversible) deformation plus a plastic (non-reversible) deformation. The second stage represents a decreasing rate of deformation with time. The third stage represents a rate of deformation which remains almost constant at a value which approximates (within the possible accuracy of measurement) to the minimum value at the point of inflexion of the curve. The fourth stage represents an increasing rate of deformation which, if allowed to continue, leads to extreme deformation and ultimate failure. These characteristic stages of the creep curve will be considered separately below.

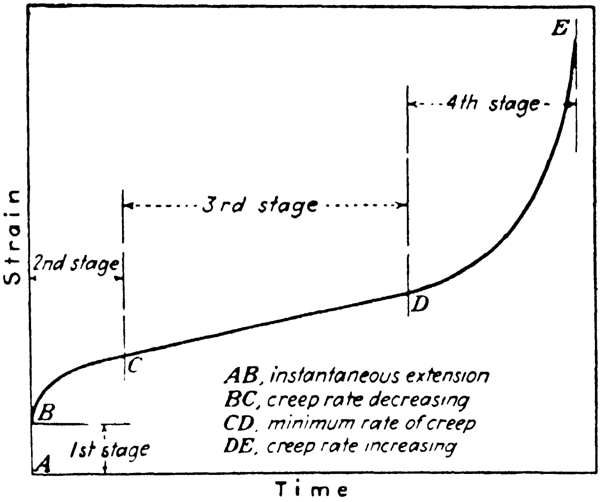


Figure 11.
The creep curve. (Hanson, ref. 111.)

It is impossible to fix exactly where one stage ends and another begins, since the deformation is a discontinuous function of time and any mathematical analysis which may be applied to the curve is, of necessity, approximate. Some investigators choose to omit the first stage and to consider the second stage as linked to the fourth stage by a stage of constant creep rate; thus they consider the creep curve to be composed of three stages, instead of four, but, although the first stage does not represent creep strain, it is usually added to the total creep strain, in the presentation of results, in order to show the total deformation under a given stress in a given time.

The question now arises as to whether, or not, the creep curves, such as 'A' and 'G,' which form the extremes of Figure 10, exhibit the same characteristic stages as those described by reference to Figure 11. By taking readings of deformation at sufficiently short intervals of time it may be shown, experimentally, that a curve such as 'A' (Figure 10) does, in fact, exhibit these characteristics. It remains to be decided, however, whether, from tests of limited duration, it may be determined that a curve such as 'G' (Figure 10) will also ultimately pass through each of the stages.

The decision on this latter point is of practical importance since, as has already been mentioned, the design engineer is not only interested in knowing that a material will not fail within a given time, but he requires to know, in addition, that under a given stress and at a given temperature, a material will not deform beyond certain allowable limits. Since exact mathematical analysis cannot be applied to the creep curve, extrapolation beyond the testing period must be treated with the utmost reserve and until a considerable number of tests, of several years' duration each, have been carried out it cannot be said with certainty that creep will persist at low stresses and temperatures in a manner characterized by the curve in Figure 11, although such long-term tests as have been made (22), (65), (66), (67), (68) indicate that this is most probably the case. In this connection the possibility of brittle fracture must always be considered, in which case the fourth stage might not be entered and fracture might take place without warning.

Design must be based, therefore, on a combination of knowledge gained from past experience of the behaviour of metals in service at elevated temperatures and the deductions derived from the results of creep tests carried out in the laboratory. From such a combination of knowledge a "working stress" may be defined as that load per unit of original area of a test specimen which, after application for a given number of years, will not produce a strain greater than that set by the tolerance limits in question, and a "limiting creep stress" may be defined as the load per unit of original area which will just not break a test specimen when allowed to remain on the material for a very long time (measured in years)(8).

The First Stage of Creep

It has already been pointed out that the first stage of the creep curve does not represent deformation which takes place by the process of creep. It is a measure of the deformation of the test specimen which occurs during the loading and its magnitude may vary with the speed at which the load is applied (this is particularly the case at the higher temperatures and stresses). At stresses and temperatures compatible with long life in service the initial deformation on loading often forms a large percentage of the total deformation during the normal period of a creep test, and it is important that the relationship between the rate of initial loading and the values quoted for total strain be studied.

The Second Stage of Creep

Bailey(69) expressed the view that the form of the creep curve may be explained on the assumption that it represents a running balance between the effect of strain hardening on the one hand and temperature softening on the other. Thus the decreasing rate of deformation which characterizes the second stage may be regarded as being due to the fact that the rate of increase of hardness produced by strain exceeds its rate of removal by temperature softening, during the interval of time over which the second stage extends.

This conception offers a plausible explanation which may be used to explain the form of the creep curve in very broad terms. Undoubtedly the phenomenon of strain hardening and its removal by temperature softening, involving the processes of crystal repair (or " recovery ")(70) and recrystallization(71), (72) is responsible in considerable measure for the form of the creep curve, but until satisfactory theories have been developed concerning the mechanism of these processes, the explanation of any phenomenon in which they are involved must, of necessity, be vague.

In a polycrystalline test specimen the second stage of the creep curve may be considered as associated also with a readjustment of stress distribution among the individual crystals as deformation proceeds(5). Here again a plausible explanation may be offered, this time in terms of what is known of the mechanism of plastic flow in solids. It may be considered that, in general, a condition of random orientation exists among crystals of a polycrystalline metal and that the application of stress causes slip to occur first in those crystals whose planes of easy slip coincide with the direction of maximum shear stress. (In support of this argument Baker, Betty and Moore(73) have shown that, for single crystals of lead, creep varies not only with load, but also with the maximum resolved shear stress on the planes of easy slip.) As slip, and consequent hardening, occurs in the most favourably oriented crystals, a redistribution of stress takes place thereby causing slip to occur in a larger number of slightly less favourably oriented crystals. This process continues until a condition is reached which approaches uniform stress distribution among the individual crystals which have now become more resistant to flow. As a consequence of slip, and during the process of stress readjustment, it is conceivable that movement, by rotation, of the crystals themselves occurs; evidence of rotation of the grains in specimens of lead in the early stages of creep is provided by Moore, Betty and Dollins(74). The amount of movement which may take place in a single grain is, however, restricted by its neighbours whose presence has the effect of " keying " it into position.

The explanation given above does not appear to be completely logical. As slip occurs in individual crystals hardening takes place until all the crystals of the aggregate approach a state of equal resistance to flow, either by virtue of their hardening or by virtue of their unfavourable orientation for slip. On this basis alone there is no cause for redistribution of stress. If slip occurs in some crystals more readily than in others then, during the slip process in a given crystal, strain will be promoted in its neighbours, and in the crystal aggregate, as this slip is accommodated; a redistribution of stress will thus occur which has no tendency towards uniformity. In fact there appears to be little ground for the assumption that uniformity of stress distribution among the crystals is ultimately approached; it seems more likely that a state of dynamic balance is set up between the weakening and strengthening of the individual crystals.

Little is understood of the part played by the crystal boundaries during the process of stress readjustment, or, for that matter, during the whole process of creep, but Chalmers(75) came to the conclusion, from his experiments on bi-crystal specimens of tin at room temperature, that, in itself, the boundary between the two crystals had no effect on the deformation under stress but acts merely as a zone of transition from one lattice to the other, and that the primary variable influencing the deformation is the angle between the lattices of the two crystals, owing to the mutual interference of their slip systems across the boundary.

It has been established that the course of the second stage of the creep curve varies, in some cases, from specimen to specimen of one material taken from the same bar(76-87) and, furthermore, that the course of the third stage is not indicated by that of the second(76-87). For want of a better explanation, these facts may be accounted for on the basis that the orientation of the crystals of a test specimen is random and the number of crystals whose orientation is favourable to slip is a matter of chance. This line of argument is not convincing, particularly when applied to fine grained material, but a more satisfactory explanation has yet to be produced.

The second stage of the creep curve is of importance from the point of view of engineering design. Estimations of permissible stress with reference to allowable tolerances in service are made from the deformation/time curve and preference is given to values of stress which will not cause creep beyond the second stage during the expected life of a material. The onset of the third stage represents, from the engineer's point of view, a period in the creep history of a material when its ultimate failure may be relatively imminent.

The Third Stage of Creep

The third stage of the creep curve contains the point of inflexion at which the rate of deformation becomes a minimum; variations in rate of deformation throughout this stage are usually so small that they are not detected by strain measuring equipment and, consequently, the stage is said to be one of constant creep rate. The measured constant rate is quoted as the minimum creep rate and it appears to represent a definite characteristic of a material, in a given condition, which is a function of stress and temperature.

By referring to the deformation/time curves shown in Figures 6, 7, 8, 9, and 10, the following points may be noted:—

1. The minimum rate of creep occurs after a longer and longer period of time as the stress is decreased.
2. The minimum rate is maintained for a longer period of time the lower the stress.
3. At the lower stresses, or temperatures, the minimum rate of creep may never be attained.
4. At higher stresses, or temperatures, the minimum rate may be masked by the large rate of deformation, or, the third stage exists only at a point of inflexion.

The apparent constant rate of deformation which characterizes the third stage of the creep curve has served as justification for it to be regarded as representing a period of viscous type flow, which implies the manifestation, by a metal, of homogeneity throughout this period of its deformation and that its physical properties should be unaffected by strain. Kanter(88) plotted a curve showing the connection between load and minimum creep rate; reference to this curve (Figure 12) shows that the load is not a linear function of the creep rate and that if viscous type flow does occur it is limited to a range of very low creep rates. Some evidence in support of the assumption of viscous type flow has been provided by Chalmers(89) (Figure 24) who has shown, from observations made on single crystals of tin, that a relatively abrupt transition from apparent viscous flow to plastic flow occurs as the rate of elongation is increased from low values. Hanffstengel and Hanemann(90) and Moore and Betty(91) have submitted similar evidence with respect to the flow of lead under low stresses.

According to Bailey's hypothesis(69), to which reference has already been made on page 47, the explanation of the period of constant creep rate arises from the assumption that the strain hardening produced by flow is automatically compensated by softening during this stage of creep. Clark and White(92) and Jenkins(93) make the explanation more definite by referring to the period of constant creep rate as being characterized by the continuous recrystallization of strain hardened material. The work of Hanson and Wheeler(94), in which they observed a constant hardness for aluminium during the third stage of creep, offers experimental evidence in favour of a balance between strain hardening and temperature softening.

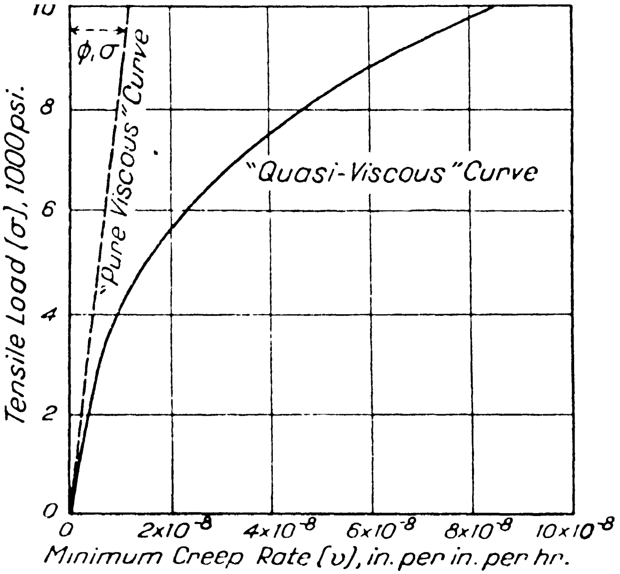


Figure 12.

Creep rate load curve according to the " Viscous " interpretation. (Kanter, ref. 88.)

Recrystallization is a time-temperature phenomenon and it is reasonable to suppose that the removal of strain hardening by recrystallization may take place at temperatures below the normally accepted recrystallization temperature as determined for short-time annealing. It would seem more likely, however, that the removal of strain hardening by recrystallization is confined to higher values of stress and/or temperature than are of practical interest from an engineering point of view and that, in the normal creep test, strain hardening is balanced by other time-temperature phenomena of a character closely associated with that of recrystallization. These phenomena are not completely understood and they are broadly described by such terms as: "recovery," "crystal repair" or "healing." One conception(72), (95-98) of the mechanism of the process of temperature softening is that grain fragmentation takes place in such a manner that a relatively small amount of energy is required to bring the fragments into line with the original crystal lattice and so to restore the unstrained properties of the individual crystals. Jefferies and Archer(99) were among the first to point out the distinction between recrystallization and recovery. Carpenter and Robertson(100) regarded recovery as a part of the recrystallization process. It has been suggested that a definite amount of deformation must be attained before recovery can take place(90), (101) but it appears that complete softening of a metal does not occur unless it recrystallizes(71), (102), (103). Thus an explanation of the form of the creep curve might be considered on the basis of recovery and recrystallization. During the second stage of creep the strain may be regarded as building up to the threshold value required for the onset of recovery, during which time the creep rate would be expected to decrease. and the third stage may be regarded as a period of gradual transition of the temperature softening process from recovery to recrystallization, the completion of which is accompanied by the onset of the fourth stage.

Wilson and Thomasson(104) consider that "creep is essentially plastic deformation which occurs at such a temperature that strain-hardening is partially offset by the recovery process" they produce some qualitative evidence in favour of this view from experiments on α- and α/β-brasses; their conclusions are that the poor creep resistance of α/β-brasses is due to the presence of the β-phase which has a much lower temperature of recovery than has the α-phase.

The Fourth Stage of Creep

The fourth stage of the creep curve is characterized by an accelerating creep rate to fracture. Kanter(88) has shown that the increase of stress due to "necking" of the specimen does not entirely account for observed rates of increase of creep and it appears that other factors, the nature of which is still obscure, must be taken into consideration.

Experiments carried out by Hanffstengel and Hanemann(90) and by Russell(78), (79), (80) on lead show that creep rate increases rapidly with the onset of recrystallization; thus the transition from the third to the fourth stage of creep may be accompanied by a change in the mechanism of temperature softening.

Hanson and Wheeler(94) have suggested from the results of experiments on single crystals of aluminium that the onset of the fourth stage of creep is accompanied by a change in the mechanism of flow and probably by the beginning of fracture.

Tapsell(8), (18) believes that accelerating creep rate begins with the development of cracking within the test specimen and McAdam, Geil and Woodward(105) support this belief. Kanter(88) has put forward the view that as flow proceeds, throughout the earlier stages, grain rotation takes place thereby giving rise to preferred orientation which may, in some measure, help to account for the increase in rate of flow.

General Conclusions

From the evidence which has been produced it appears that before a satisfactory explanation of the form of the creep curve can be given, further investigation into the processes of strain hardening and temperature softening is required. More definite information concerning time-temperature effects, such as recovery and recrystallization, and their inter-relationship must be obtained and a more complete understanding of the mechanism, or mechanisms, of flow in metals must be reached. In the study of the effect of grain rotation and grain boundaries much more information is needed on the creep of single crystals.

CHAPTER III

THE FACTORS WHICH INFLUENCE THE CREEP OF METALS

Grain Size

THERE IS ample evidence(7), (65), (76-86), (90), (106) that variations in grain size influence the creep characteristics of metals and alloys, but, in spite of this evidence, very little progress has been made towards an understanding of the exact nature of the influence, or of the mechanism whereby the influence is effected.

A problem which confronts all investigators who are concerned with the determination of the influence of grain size on the properties of materials is that of developing a satisfactory method of measuring the actual grain size. A condition of absolute uniformity of grain size in any material is impossible to achieve and it would seem that an investigation which aims at providing reliable information must be put on a statistical basis. Practically the whole of the work which has been carried out has been concerned with comparing the effects of coarse and fine grain. When grain size is classified in such broad terms the results obtained can only lead to conclusions of a generalized nature.

Information which has been gained on this subject has led to the conclusion that, in general, coarse-grained material is more creep resistant than fine-grained material at the higher temperatures and that at the lower temperatures the reverse is the case. Beyond this statement existing knowledge does not go very far.

Kanter and Spring(7) and Clark and White(106) reported evidence of this generalization in the case of steels. Cross and Lowther(107) have studied the effects of factors which influence the creep characteristics of steels and have shown that true austenitic grain size is an important variable for killed steels; they were able to vary the creep resistance of the same steel from good creep resistance to poor creep resistance by altering the treatment to produce coarse or fine austenitic grain.

Clark and White(108) suggested that coarse-grained material is more creep resistant than fine-grained material in the temperature range above the minimum temperature of recrystallization; but, since recrystallization is a time-temperature phenomenon the interpretation of this suggestion is somewhat vague.

The danger of making generalizations was pointed out in the introduction and, as with most considerations in connection with creep, there are plenty of exceptions to the rule for grain size. Hanffstengel and Hanemann(90) found that, as the temperature of specimens of lead is raised, coarse-grained material does not continue to give better creep resistance than fine-grained material. Jenkins, Tapsell, Mellor and Johnson(109) found a killed steel of composition: 0.19% carbon, 0.40% manganese and 0.08% silicon which was not improved by grain coarsening, and Cross and Lowther(110) found that the creep resistance of Carbon-Vanadium steel was not improved by grain coarsening.

Hanson and Sandford, and Pell-Walpole(85), (111), (112) investigated the effect of grain size on the creep of tin and tin alloys. They pointed out the importance of taking into consideration the cross-sectional area of the test-piece, for, in the extreme case, the most coarse grain size would be that of a single crystal, and single crystals are noted for their poor creep resistance. Thus Hanson and Sandford recorded their grain sizes as the number of grains in the cross-section of the test-piece. From measurements of time to fracture and of minimum creep rate under constant load they were able to establish an optimum grain size, for the dimensions of their specimens, for longest life and smallest value of creep rate.

Parker and Riisness(113) carried out creep tests on oxygen free high-conductivity copper at 200° C. in which the grain sizes ranged from 3 to 500 grains per square millimetre in test-pieces of 0.160, 0.375 and 0.505 inch diameter. The creep strength was found to be independent of grain size for each size of test-piece; the two larger sizes of test-piece were of equal strength but the smallest size was weaker, particularly at the higher stresses. This difference was attributed to the relatively larger effect of oxidation of the smallest test-pieces, due to their higher ratio of surface to volume. It seems that this contention was largely correct since the creep strength of the smallest test-pieces was considerably improved when they were protected with a coating of cadmium. Thus it appears that oxidation must be taken into account when considering the factors which influence creep.

It is difficult to evaluate the effect of grain size, as it is difficult to evaluate the effect of any factors which influence creep, since it is almost impossible to isolate any one factor. For example, when a metal, or alloy, is subjected to different treatments to produce different grain sizes it may be argued that the investigation is not being carried out on the same material in each case. Greenwood and Russell(76), (81-84) carried out experiments on samples of very pure lead (99.99%, or better); all samples were of the same grain size but considerable variations in creep characteristics were found among the samples; these variations were attributed to the presence, and form, of minute impurities. Again, Cross(114) found little difference in the creep behaviour of coarse-grained cast and fine-grained wrought austenitic 18:8 steel, which may be attributed to the difference in form (or, perhaps, manufacture) of the two materials, or to a difference in minor chemical composition.

In view of the difficulty of obtaining exact information on the effect of grain size on the creep characteristics of metals and alloys it is not surprising that no satisfactory explanation has been advanced to account for the various results which have been obtained. The earliest theories were based on the doubtful generalization that at low temperatures fractures in metals were of a transcrystalline type and at high temperatures they were of an intercrystalline type. Rosenhain(115-118) supported the conception that grain boundary material could be likened to an amorphous cement which is very hard at low temperatures and which behaves like a viscous fluid at high temperatures. Jeffries and Archer(99) introduced the conception of an equicohesive temperature which marks the temperature range where the transition takes place from transcrystalline to intercrystalline fracture. Equicohesive temperature has been used in the literature on creep to explain the difference between the behaviour of coarse-grained and fine-grained material; but, without the slightest justification, these explanations go far beyond the original concept of Jeffries and Archer, and must be rejected on the grounds that they lack both theoretical and experimental support.

The flow of crystalline solids is discussed in Chapter V where the scarcity of information with respect to the flow of polycrystals is emphasized. Before a satisfactory theory of grain size effect can be formulated, more comprehensive knowledge is required of the mechanism of: (1) flow in crystals, (2) the flow properties of grain boundaries, and (3) the restrictions to flow impressed on individual grains by the presence of neighbouring grains and by the flow of the grain aggregate.

Heat-treatment and Structural Stability

The importance of the effect of heat-treatment on the creep resistance of metals and alloys has been recognized since the days of the pioneer work on creep testing(8), but so little attention has been devoted to this aspect of the subject in subsequent experimental work that information on the relationship between heat-treatment and creep resistance is still very lacking in essential detail.

The influence which heat-treatment has on the creep resistance of metals and alloys is indirect and is manifested by the effect of other factors which arise as a direct result of heat-treatment; among such factors, grain size and structural stability are the most prominent, but, here again, further sub-division can be made and the difficulty associated with the consideration of any one factor by itself is immediately evident.

The problem of determining the best type of heat-treatment for metals and alloys for use in service at elevated temperatures was first studied by means of short-time tests. Investigations were carried out in which the effects of tempering quenched steels at different temperatures for different lengths of time were examined(8). It was assumed that those heat-treatments which gave, to steels, a high degree of their quenched strength were the most suitable for high temperature service. The danger of this assumption was pointed out by Tapsell(8) who stated that the amount and rate of temper softening with time for a particular steel is not fully indicated by the effects of short period re-heats.

It seems reasonable to assume that if a material, under stress, is to be exposed to an elevated temperature for a long enough time, its structure must ultimately attain a condition which, for all practical purposes, might be regarded as an equilibrium structure the nature of which is dependent upon the stress and temperature in question. If the temperature is sufficiently high an equilibrium structure similar to that of the material in the annealed condition might be expected. Again it seems reasonable to assume that creep resistance decreases with the onset of structural instability and, although it is dangerous to generalize to this extent, some support for the assumption is to be found from the work carried out on lead by Hanffstengel and Hanemann(90) and by Russell(78), (79), (80).

When consideration is given to the influence exerted by heat-treatment on the creep resistance of some alloys, through its effect on stability of structure, it is usual to find the following phenomena discussed:—

(a) The phase changes which can take place at the particular temperature in question.

(b) The rate of precipitation of the solute constituents from solid solution at the temperature in question.

(c) The type of precipitation (whether continuous or discontinuous(119)).

(d) The size of the precipitated constituents.

It is doubtful whether heat-treatment effects can be tied down so neatly into so few categories, but more information gained from systematic studies of these phenomena in relation to creep would do much to throw light on the role played by heat-treatment in influencing creep resistance. (The effect of grain size on rate of precipitation should be noted; evidence appears to show(119) that, for the same material, the rate of precipitation increases as grain size decreases.)

When considering the significance of structural stability it must be borne in mind that this factor is influenced by temperature, by time at temperature and probably by stress. Thus for materials, which are required for service at temperatures at which phase changes and precipitation take place very slowly, greater creep resistance may be obtained after such treatments as quenching and tempering, or solution treatment followed by ageing. An investigation of the effects of these treatments involves the determination of the degree of stability of the resultant structure (after heat-treatment) at the temperature of operation and again after a period of time exceeding the expected life of the material.

From results of investigations on steels(108) it has been found that at temperatures above approximately 500° C., annealing is the most suitable treatment (here normalizing is regarded as a form of annealing). Bailey and Roberts(36) studied this aspect of the subject, however, and drew attention to the fact that annealing treatments do not always produce a structure which is stable throughout the service life of steels at elevated temperatures. They showed that pearlite may change during service from the lamellar to the spheroidized structure, which change is accompanied by a decrease in creep resistance.

Hanson and Sandford(85) studied the effect of different types of annealing treatment on the creep resistance of tin alloys (containing 2 to 7% cadmium and 3 to 9% antimony). They found, from the creep resistance point of view, the most beneficial annealing treatment was dependent upon the chemical composition of the particular alloy in question. Thus it appears that in the case of materials which may be annealed satisfactorily (according to the normally accepted tests) by different methods, further qualification of the process is necessary when creep has to be taken into consideration.

In the studies of the effect of precipitation-hardening on the creep resistance of alloys, Miller, Campbell, Aborn and Wright(120) found that heat treatments which produced finely divided particles of carbide dispersed throughout the matrix of carbon-molybdenum steels, made a marked improvement in the creep resistance. From investigations on the creep of lead at room temperature, Greenwood(81-84) reported that finely disseminated copper particles enhance creep resistance while finely distributed particles of lead telluride do not.

It follows from the above observations that it is dangerous to regard finely divided particles as having a " keying " action on the slip planes. This expression is often used to describe the effect of precipitation-hardening and it is an analogy which adds little to the understanding of the mechanism of the process. Precipitation-hardening is a time-temperature phenomenon and when an alloy which exhibits precipitation-hardening characteristics is exposed to an elevated temperature, as in a creep test, the precipitation of finely divided (sub-microscopic) constituent throughout the matrix appears to increase the resistance to flow but, with the passage of time, the particles of precipitate grow in size, and number, and the material becomes over-aged, which condition is accompanied by a decrease in resistance to flow.

Jenkins, Bucknall and Jenkinson(121) investigated the inter-relation of age-hardening and creep performance of a copper-nickel (2.4%)-silicon (0.6%) alloy, in the temperature range 15° C.—625° C. The material was given a solution treatment of 2 hours at 900° C., which was followed by quenching and different ageing treatments to produce various conditions of hardness ranging from completely soft material (obtained by slowly cooling, over a period of 4 to 7 weeks, from 900° C. to 200° C.) to fully hard material (obtained by ageing for 2 hours at 500° C.). The results of their tests led to the following conclusions:

1. At temperatures above, approximately, 230° C. the superiority, in creep resistance, of the fully hardened material, over the fully soft material, disappears.
2. At temperatures of 300° C. and 450° C. the creep resistance of the fully hard material is poor in comparison with that of the fully soft material. (At 450° C., and under a stress of 4 tons per square inch, the life of the fully soft material may be eight times as great as that of the fully hard material.)
3. At 600° C. the fully soft material is still superior but, at this temperature, material in all conditions ages rapidly and, in tests lasting 1,000 hours, all samples tend to reach the same condition and to exhibit the same creep resistance.

It appears, therefore, that in this particular case, age-hardening ceases to be beneficial, from the point of view of creep resistance, at temperatures above 230° C. The reason for this was attributed to the instability of the structure of the hardened material at the higher temperatures.

An investigation of the effect of heat-treatment on the creep characteristics of aluminium alloys would be valuable in providing information to fill in the gap in existing knowledge on this subject. Kennedy(38) carried out an investigation of the elevated temperature properties of materials for use in aircraft engines and obtained results which indicate that, for aluminium alloys, in the temperature range 205° C.—315° C. the annealed condition is probably the best for maximum creep resistance. Apart from this work of Kennedy's there is little, or no, published information available.

After the solution treatment of a duralumin type aluminium alloy the rate of precipitation of solute elements from solid solution is almost infinitely fast at the maximum temperature (approximately 350° C.) at which these alloys find service and is almost infinitely slow at room temperature. Thus a knowledge of the relationship between heat-treatment and the service temperature of aluminium alloys is a matter of prime importance.

Basing assumptions, with respect to aluminium alloys, on the work of Jenkins, Bucknall and Jenkinson (referred to above) it might be expected that at temperatures above approximately 150° C. (in which range rapid precipitation from solid solution takes place) annealed material would be superior, in creep resistance, to fully heat-treated material. At temperatures below 150° C. the reverse may be the case. There is a danger, however, in assessing the value of heat-treatment from tests of shorter duration than the expected life of the material, and the decision on the comparative merits of various types of heat-treatment will probably depend on the establishment of a range of stress and temperature in which the onset of over-ageing does not occur within the expected life. In this connection some idea might be formed of the degree of structural stability of aluminium alloys, after various types of heat-treatment, by comparing the creep characteristics of specimens, in various conditions, at stresses which bear a constant ratio to the proof stress in each case.

Chemical Composition

Perhaps the most obvious approach to the determination of a means of improving the creep resistance of materials is that of studying the effects of alloying elements in relation to the production of desirable structures. Very little systematic work has been published along these lines, but it may be said that, in general, success has been achieved by the addition of alloying elements which raise the softening temperature, or which produce precipitation-hardening effects at the working temperature, or which slow up the rate of phase changes.

It is a reasonable supposition that elements which raise the lowest temperature of recrystallization of a material will also increase its creep resistance at temperatures below the recrystallization range. Here again it must be remembered that recrystallization is a time-temperature phenomenon and it is dangerous to carry the supposition too far without the support of experimental evidence.

At temperatures below the lowest temperature of recrystallization, improved creep resistance in steels is obtained by the addition of elements, such as nickel and manganese, which enter into solid solution with the ferrite(108). Elements which form carbides, such as chromium, tungsten, molybdenum and vanadium, also bring about a marked improvement in creep resistance(108); this is particularly so in the case of molybdenum(5), (106). The carbides formed by these elements are distributed, in the form of finely divided particles, throughout the matrix and at the grain boundaries; their action in resisting creep is visualized as a " keying action " (again for want of understanding of the action).

Since the room temperature strength of plain carbon steels may be increased by raising the carbon content it might be expected that such a procedure would also increase the creep strength. Norton(21) found that the creep strength of carbon steels increased with carbon content between the limits of 0.08% and 0.42% at 538° C. Tapsell(8) found increased carbon content beneficial up to a temperature of, approximately, 450° C.; above this temperature, however, variations in carbon content had little influence. Clark and White(122) investigated the effect of variations in carbon content on the creep characteristics of a series of plain carbon and alloy steels; their results show that, for plain carbon steel, increase of carbon is beneficial at temperatures up to 538° C. but, at the three temperatures considered (427° C., 538° C., 650° C.), the creep strength of alloy steel is lowered by increased carbon content.

The addition of alloying elements to steel will slow down the rate of diffusion and also the rate of physical changes (e.g., the balling up of pearlite) and phase changes (e.g., the tempering of martensite) due to heat-treatment. It is possible that effects such as these are as important as recrystallization.

It is recognized(123), (124) that killed steel is superior in creep resistance to rimming steel but, so far, no satisfactory explanation of this fact can be offered. It must be concluded, therefore, that there are still more factors, as yet unknown, which influence the creep of metals.

Clark and White(106) commented on the fact that electric-arc melted steel appears more creep resistant than basic open-hearth steel and that induction-melted steel may be more creep resistant than either. They also pointed out that ingot size and degree of reduction in hot working may be important variables. So it appears that variations in method of manufacture influence the creep of metals.

Very little is known of the influence which alloying elements have on the creep characteristics of non-ferrous alloys and a systematic investigation of the relationship between creep strength and chemical composition would be of the utmost value. Extensive work has been carried out on the affects of alloying elements on the age-hardening of the duralumin-type alloys(119) and it would be interesting to observe whether the variation in creep characteristics and age-hardening characteristics brought about by variations in chemical composition are in any way related. Aluminium alloys containing high nickel (of the order of 2%) possess relatively good creep resistance and the alloys of the duralumin type containing high magnesium (of the order of 1%) show indication of greater creep resistance than those containing low magnesium(125), but conclusions drawn from this evidence must be regarded with the utmost reserve, since in the materials used, variations in other alloying elements have occurred at the same time, and the difference in behaviour is probably a result of the effect of a total variation in a number of elements.

Creep behaviour is undoubtedly related to such factors as: solid solubility, the amount of intermetallic compounds present, the form, size and distribution of intermetallic compounds, and the rate of precipitation. A systematic research in which careful observation of such variables was made would do much to provide better understanding of the subject.

CHAPTER IV

THE PRESENTATION OF CREEP RESULTS

Engineering Requirements

IN THE design of engineering equipment it is necessary to establish a safe working stress for each component under service conditions. To do this, information must be supplied concerning the behaviour of the materials used when subjected to stress under service conditions. In applications where the total deformation during service life is not important, and the only concern is that the applied stress shall not lead to rupture, the required information can be supplied from relatively crude tests; but, as Tapsell has pointed out(18), for applications where the amount of permissible deformation determines the stress which can be applied, a detailed analysis of creep results, obtained under the most rigid conditions, is essential.

The creep requirements of various structural components are to be found in the literature on the subject(5), (8), (36), (126). The highest creep rates tolerated are in such applications as the construction of superheater tubes, or the lead linings for sulphuric acid chambers, where a total deformation of 5%, in the expected life, might be allowed. Turbine engineers demand that the permissible creep rate for turbine blades shall not exceed 10^{-9} in./in./hour, and in most applications, where creep is a factor to be considered, a creep rate of 0.1% in a life of from 10 to 20 years is an outside figure.

In most cases the service life demanded of a material exceeds the amount of time which can possibly be devoted to laboratory tests by many thousands of hours; it follows, therefore, that the results of laboratory tests must be extrapolated to times equivalent to the expected life. The need for extrapolation is a serious handicap, since, in carrying out the process, the assumption has to be made that the stage of accelerating creep rate will not be entered. This assumption can by no means be made with any degree of certainty as is shown by the curves in Figure 13 obtained by Clark and White[65]; the 17,500 lb. per sq. in. curve showed a fairly constant creep rate from 1,000 to 4,000 hours, but from 4,000 to 7,000 hours the rate increased.

Thus the results of creep tests can serve only as an indication of the most probable behaviour of a material during its life in service (due regard being paid, of course, to other factors such as fatigue, corrosion and complexity of temperature conditions, which are not always stable). Tapsell[8] has stated that if the applied stress is such that a stage of constant rate of strain of the order of 10^{-5} in./in./day is maintained, and it is known that the material is metallurgically stable (this is important), then such a rate of strain can be associated with long life of many years under conditions similar to those of the test. Support for this contention may be found from the work of Cross and Lowther[67], [68] whose results on 18 : 8 steel, and an annealed 0.35% carbon steel (see Figures 14 and 15), showed that the creep rate diminished for 4,000 hours and then became very constant for the remainder of the 20,000-hour test; the samples used were exceptionally uniform and were, initially, in a very stable state.

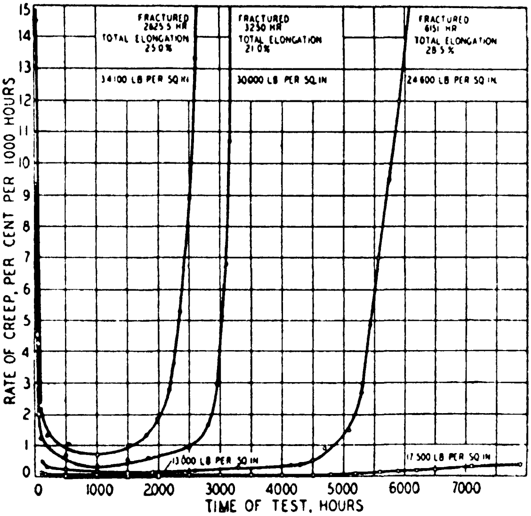


Figure 13.

Creep rate/time curves. (Clark and White, ref. 65.)

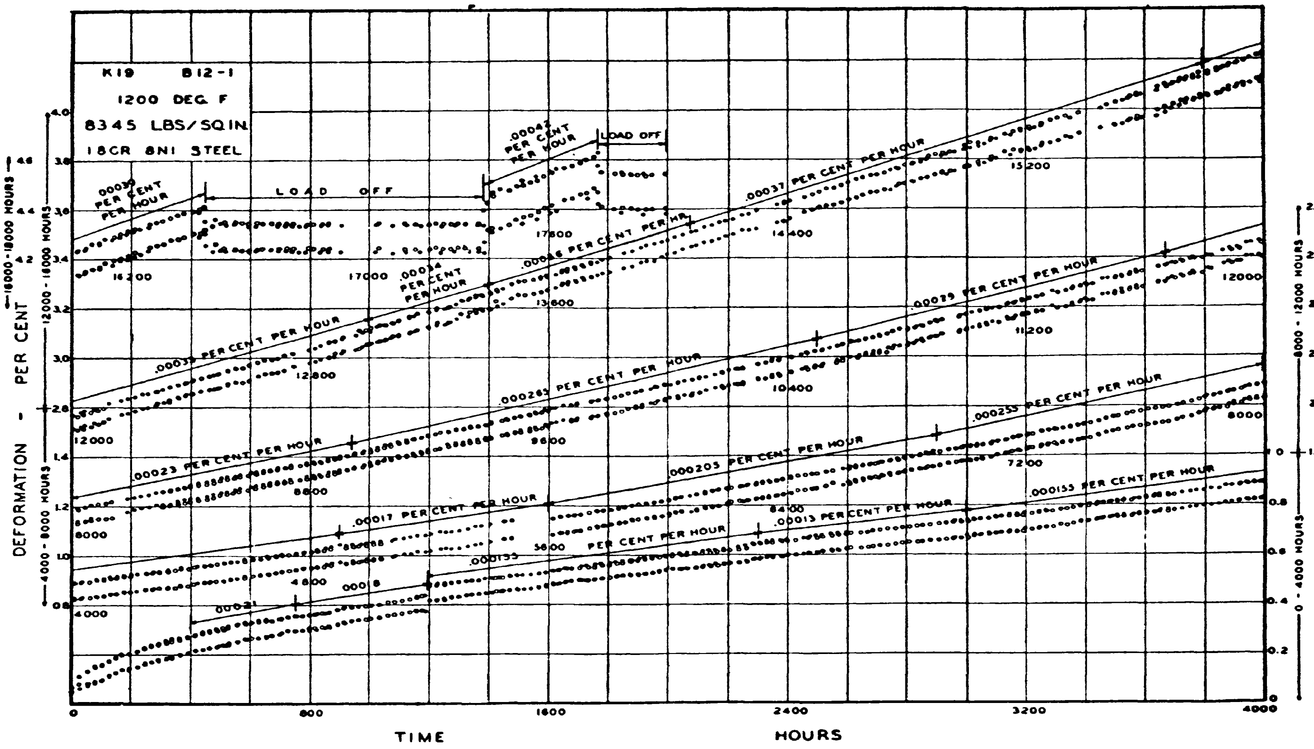
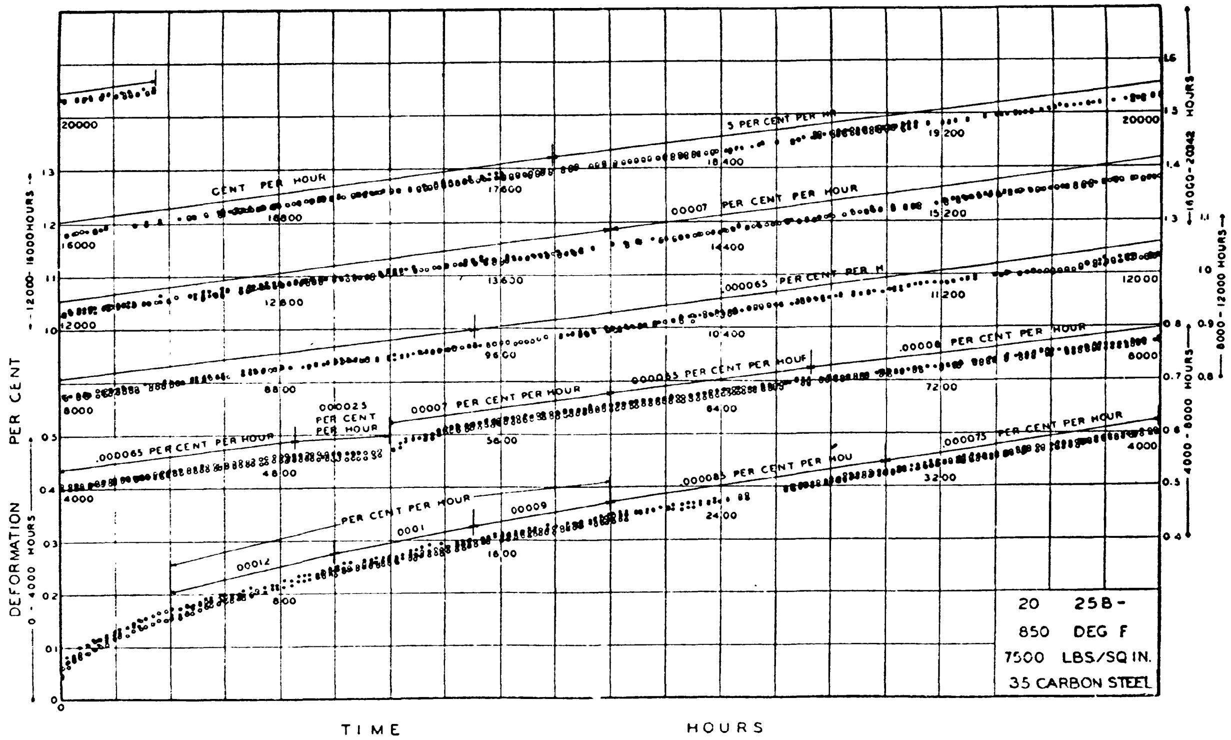


Figure 14.
Creep test on 18% Cr, 8% Ni for 17,000 hours, showing long maintenance of the constant-rate stage.
(Cross and Lowther, ref. 67.)



TIME HOURS

Figure 15.

Creep test on 0.35% C steel for 20,000 hours. (Cross and Lowther, ref. 68.)

Until sufficient experience has been gained, from the use of materials in service at elevated temperatures, to enable a correlation between working stresses and creep test measurements to be made, design data must be established on such a basis as that defined by Tapsell.

The Presentation of Creep Results

Many workers, particularly in the 1920-30 period, devoted attention to the " time to fracture " of a material under given conditions of stress and temperature. Their results were presented in the form

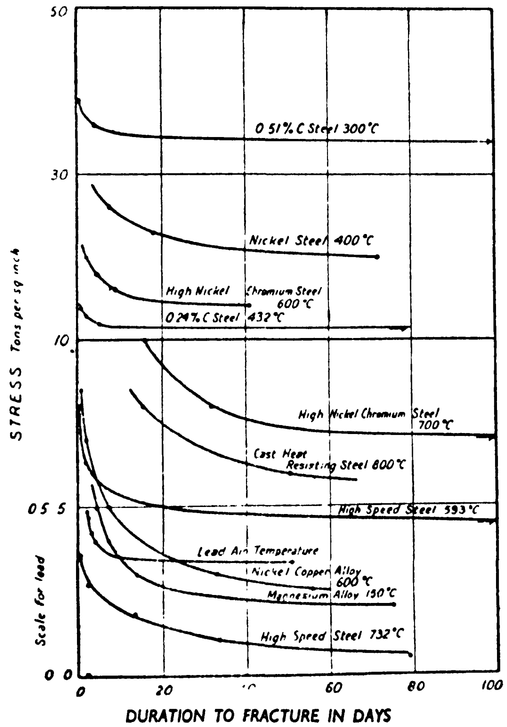


Figure 16.
Stress/duration curves to fracture.
(Tapsell, ref. 8.)

of curves showing the relationship between initial stress and " duration to fracture " at a constant temperature, or, between temperature and " duration to fracture " at a given initial stress.

Typical curves of this kind are shown in Figure 16. The form of these curves immediately suggests one of three possibilities for the relationship between initial stress and " duration to fracture." The first possibility is that stress may be expressed as an exponential function of the duration, indicating that ultimate failure occurs at all stresses. The second possibility is that stress may be expressed as a hyperbolic function of the duration, which function is asymptotic at positive values of stress, thus indicating the existence of a " limiting creep stress." The third possibility is that the duration may be expressed as a power function of stress of the form: $D = a S^n$.

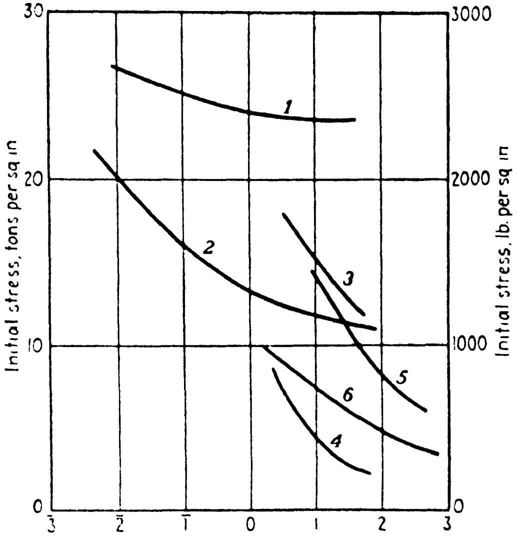


Figure 17.

1. 0.24% C steel, 293°C.
2. 0.24% C steel, 432°C.
3. Nickel-copper alloy, 500°C.
4. Nickel-copper alloy, 600°C.
5. Tin containing 1% bismuth.
6. Tin containing 0.04% silver.

(Hanson, ref. 111.)

The validity of these possibilities may easily be checked against results obtained from the literature on the subject and, while any one of the relationships may hold for short durations, mathematical analysis persistently yields results for long durations which are not consistent with experience. In Figure 17 experimental results are plotted with initial stress as ordinate and log (duration) as abscissa and it is clear, from the fact that straight line plots are not obtained, that, in these instances, the exponential relationship does not hold. Tapsell[(6)] has discussed the possibility of a hyperbolic relationship and has pointed out that it gives a stress for infinite duration which is considerably greater than the stress for long life estimated on the basis of creep rate. Figure 18 shows the results used in Figure 17 replotted on a log.log scale; again, straight line plots are not obtained and it is evident that the third relationship is not generally applicable.

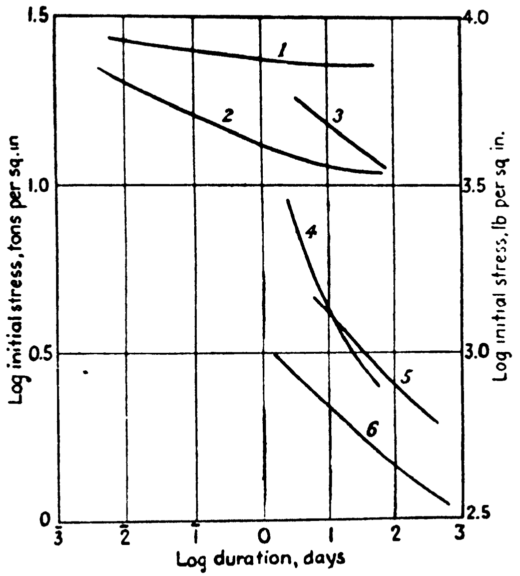


Figure 18.
Stress/duration curves plotted on log.log scale from results used in Figure 17. (Hanson, ref. 111.)

French[(2)] was among the first to draw attention to the fact that, from an engineering point of view, the study of the progress of deformation by creep is of greater importance than the determination of "time to fracture." In 1927 he, together with Cross and Peterson, published a paper(127) in which the opinion was expressed that creep test results could best be represented by a series of curves showing the relationship between stress and temperature for a definite amount of creep strain in a given period of time, and a diagram was constructed based on the stress required to produce 0.1% deformation in 1,000 hours. This work, although lacking the refinement customary to creep testing nowadays, was of fundamental significance in that it drew attention to the importance of deformation, and indicated how observations of deformation could be presented in a form which would show the maximum values of stress and temperature which a material could be expected to withstand without creep becoming objectionable.

Most laboratories are equipped with a small number of testing machines and the full investigation which is necessary to produce the required information for the construction of a working stress diagram, of a type referred to above, is a task of some magnitude. It is sometimes convenient and expedient, therefore, in gaining information as to the relative temperature limitations of materials in a given application, to conduct tests with temperature as the independent variable. In this manner the suitability of a material at a working stress can be determined directly; the experimental procedure is that of observing creep under a picked stress at, say, three successively higher temperatures until an increment has been reached at which the specimen shows creep rates beyond those of interest.

For applications where the maximum total deformation permissible is of the order of 0.1% it is essential that primary creep should be given full consideration since 0.1% strain often occurs during the early stages of a test, the conditions of which yield an almost negligible rate during the steady rate period.

It has already been pointed out in Chapter I that it is important for laboratories, engaged on creep testing, to be in a position to compare results. In the fulfilment of this requirement the need for accurate records of testing conditions and metallurgical details of materials tested cannot be too greatly emphasized. In the author's opinion a standard procedure should be adopted, in reporting results, on a basis similar to the scheme given below*:—

The details of testing conditions should be supplied and should include, at least, the following items:

1. Gauge length of specimen.
2. Diameter of specimen.
3. A record of the day-to-day temperature readings of the specimen (it is preferable that this record should be shown on a graph and that readings of the temperature of the specimen should be taken at regular intervals throughout each day of the testing period).
4. A record of room temperature should be supplied with item 3.
5. The type of extensometer used and the smallest extension which can be measured by its use should be quoted.
6. The method of loading the specimen should be given.

The results of the tests should include, at least, the following information:

1. The type of material tested (whether wrought or cast).
2. The form of the material (whether forged slab, extruded bar, etc.).
3. The direction of the specimen relative to the material.
4. The details of heat-treatment.
5. The chemical composition.
6. Details of initial and final micro-structure.
7. An estimate of the grain-size.
8. Details of the room temperature physical properties before creep (tensile and hardness).
9. Details of the room temperature physical properties after creep.
10. A description of the details of manufacture of the material.
11. The initial strain on loading the test-piece should be quoted.

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* In suggesting this scheme, and in the execution of creep investigations, the author has been guided by the Tentative Method of Test for Long-Time (Creep) tests laid down by the A.S.T.M.(52).

12. Sufficient data should be given to enable the reconstruction of the creep curve to be made.

13. A statement of the minimum creep rate should be given for each combination of stress and temperature.

14. The total creep strain taking place in a fixed time (usually 1,000 hours) at the various temperatures and stresses should be given, or, alternatively, the time required for some fixed value of creep strain to be attained should be given.

15. An estimate should be made of the extent to which the test specimens truly represent the material under investigation.

16. A working stress diagram should be constructed and the stages in the development of this construction should accompany the final diagram.

The Extrapolation of Creep Results

In view of the length of time involved in obtaining results from creep tests, and the need for the extrapolation of these results, it is not surprising to find that a considerable amount of effort has been expended in trying to develop reliable, empirical, mathematical relationships between creep variables. Mathematical analysis cannot be regarded as having achieved any success, but until such time as a reliable procedure for extrapolation has been developed on a basis of service experience, certain methods are commonly employed to assist in forming an idea of the probable behaviour of material under service conditions. These methods are summarized below.

At the lower stress values the creep/time relation becomes, in many cases, essentially a straight line which is represented by the line AB in Figure 19. Thus the creep " ϵ " in a given time " t " is expressed by:

$$\epsilon = .\epsilon_0 + Mt \quad .. \quad .. \quad .. \quad (IV.1)$$

where $\epsilon_0 = OA$ and $M = d\epsilon/dt$.

Most of the methods used for interpreting creep test data begin with this assumption, and the further assumption is made that the creep/time relation given by AB (Figure 19) can be extrapolated as a straight line to periods of time beyond those covered by the tests.

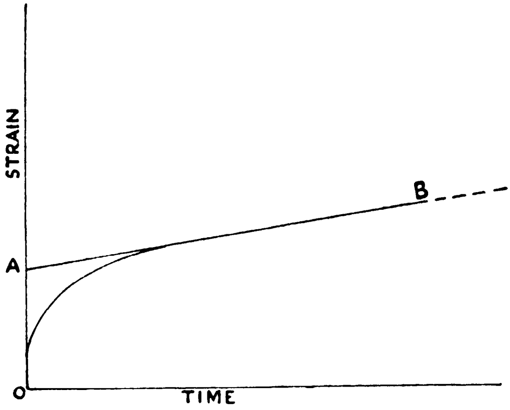


Figure 19.
The creep curve.

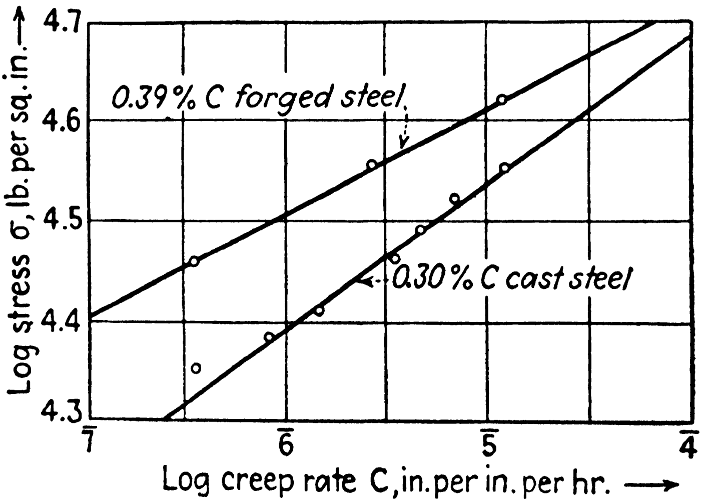


Figure 20.
Creep rate/stress relation. (Marin, refs. 129, 130.)

THE LOG.LOG METHOD: In the log.log method the assumed constant creep rates are measured for a series of stress values and a straight line is drawn through points on a graph of log (creep rate) plotted against log (stress). It may be seen from Figures 20 and 21 that the points often do lie on a straight line to a good degree of approximation; additional evidence of this can be found in the "Compilation of Available High Temperature Creep Characteristics of Metals and Alloys" published by the A.S.T.M. and the A.S.M.E.(128). That the log.log plotting does not always yield straight lines, however, may be seen by reference to Figure 21, which is taken from Clark and White(106), and in Figure 22, which is taken from Moore, Betty and Dollins(74). Kanter(88) considers that the log.log plotting is not exact enough to use with assurance for large extrapolation.

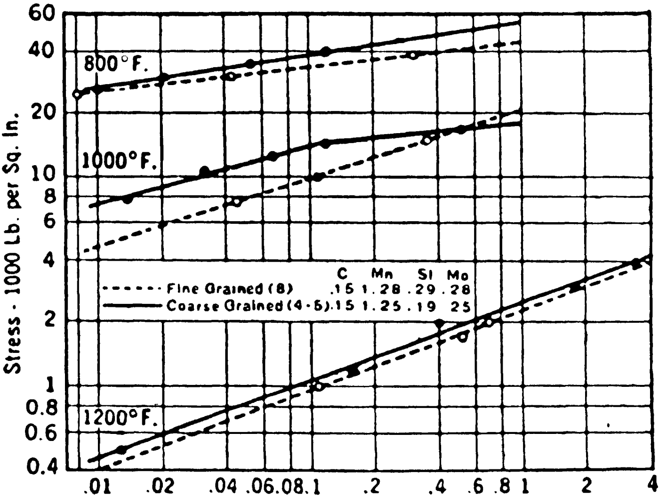
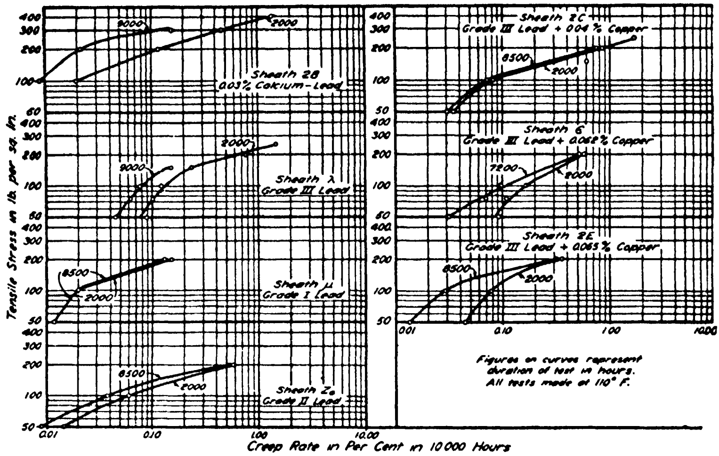


Figure 21.

Log.log plot of stress against creep rate.
(Clark and White, ref. 106.)



Figures on curves represent duration of test in hours.
All tests made at 110° F.

Figure 22.
Log.log plots. (Moore, Betty and Dollins, ref. 74.)

Mathematically, the log.log method assumes a stress/creep deformation relationship of the form:

$$\log(d\epsilon/dt) = n\log(S) + \log(A) \qquad .. \qquad (IV.2)$$

where ϵ = strain, S = stress, t = time and n and A are constants which are determined from experiment, that is:-

$$d\epsilon/dt = AS^n \qquad .. \qquad .. \qquad .. \ (IV.3)$$

THE LOG METHOD: This method is based on similar assumptions to those made in the log.log method but, instead of using a straight line relationship between log (creep rate) and log (stress), a straight line graph is drawn through points on a log (creep rate)/stress diagram. Thus, in this case,

$$\log(d\epsilon/dt) = AS + B \qquad .. \qquad .. \ (IV.4)$$

Some test data(129), (130) appear to show better agreement with this form of plotting but, again, exceptions to the rule are found in abundance.

McVETTY'S METHOD(131): In this method a creep/time relationship is assumed of the form:

$$d\epsilon/dt = A + Be^{-kt} \quad \ldots \quad \ldots \quad \ldots \quad (IV.5)$$

where ϵ = creep deformation, t = time and A, B and k are constants for the material under the particular conditions of stress and temperature considered.

By integrating equation *(IV.5)* a relationship of the following form is obtained:

$$\epsilon = \epsilon_o + At - \frac{B}{k}e^{-kt} \quad \ldots \quad \ldots \quad (IV.6)$$

and as "t" becomes large, ϵ approaches the value:

$$\epsilon = \epsilon_o + At \quad \ldots \quad \ldots \quad \ldots \quad (IV.7)$$

which is equation *(IV.1)* and represents the asymptote to the creep curve given by equation *(IV.6)*.

McVetty(132) has shown that actual creep/time curves derived from experimental results can be represented quite accurately by equation *(IV.6)*. For times beyond the range covered by tests he assumes that creep deformation is defined with sufficient accuracy by equation *(IV.7)*, and he verifies his assumption by applying data obtained by Kanter and Spring(37).

BAILEY'S METHOD(133): In this method a series of tests are carried out at different temperatures and stresses so that the times required to attain a given creep strain under the different conditions of temperature and stress can be determined. From this information a family of curves may be plotted on a temperature/log (time) graph, each curve of the family corresponding to a particular stress. By extrapolation of these curves, to times equivalent to the expected life of the material in service, a working diagram may be constructed showing the relation between stress and temperature for a given strain in a given time.

The Method Used by Metropolitan-Vickers Ltd.(134): The method used by Metropolitan-Vickers Ltd., for the extrapolation of creep test results, is based on the work of Bailey(133). In this method it is considered that closest agreement with experimental results is obtained by use of an expression of the form:

$$\frac{d\epsilon}{dt} = \frac{AS^n}{\epsilon^p} \quad .. \quad .. \quad .. \quad (IV.8)$$

where ϵ = creep deformation, t = time, S = stress and A, n and p are constants of the material at the particular temperature in question and are determined by experiment.

By integrating equation *(IV.8)* the following relation is obtained:

$$\epsilon^{(p+1)} = A(p+1)\, tS^n + \text{constant} \quad .. \quad .. \quad (IV.9)$$

If now a series of tests are carried out at different stresses and from each test the time to attain a fixed strain is measured then:

$$tS^n = \text{constant (since } \epsilon = \text{constant)} .. \quad (IV.10)$$

and a plot of log (stress) against log (time for the given strain) approximates to a straight line the slope of which gives the value of n.

A knowledge of n is considered important for it not only indicates how creep resistance varies with load (thus enabling a suitable factor of safety to be established) but it also indicates how a section will behave when varying stress acts on the cross-section, as in the case of bending.

The value of " p " may be determined by plotting log (strain) against log (time) for a given stress. This again approximates to a straight line the slope of which gives the value of $(1/p + 1)$.

Weaver's Method (135): Weaver assumed that creep rate varies inversely with time and that it approaches a constant value asymptotically, thus:

$$\frac{d\epsilon}{dt} = \frac{A}{t} + B \quad .. \quad .. \quad (IV.11)$$

where A and B are constants, and it follows from this equation that:

$$\epsilon = A \log t + Bt + \text{constant} \quad .. \qquad (IV.12)$$

STURM'S METHOD (136): Sturm carried out experiments on wires of copper, of steel and of aluminium alloys, and he showed that, within the limits of his tests, the creep behaviour of his specimens was closely represented by a straight line relationship between log (strain) and log (time).

SODERBERG'S METHOD (137): Soderberg considered the problem of expressing creep strain in terms of stress and time when temperature remains fixed. He considered creep/time curves as being geometrically similar and expressed creep strain as:

$$\epsilon = f(S)\,F(t) \quad .. \quad .. \qquad (IV.13)$$

where f(S) represents a function of stress only and F(t) represents a function of time only.

Soderberg's method for determining f(S) and F(t) was based on an observation made by Ludwik(138) that creep curves approach asymptotically to a minimum creep rate and that arithmetic increments in stress produce geometric increments in creep rate. Thus he obtained the following expressions for f(S) and F(t):

$$f(S) = A\,(\exp.\frac{S}{S_1} - 1) \quad .. \quad .. \qquad (IV.14)$$

$$F(t) = t + t_0\left\{1 - \exp\left(-\frac{t}{t_1}\right)\right\} \quad .. \qquad (IV.15)$$

A, S_1, t_0 and t_1 are experimental constants.

NADAI'S METHOD (139): Nadai made the assumption that creep rate is related to stress at constant temperature by a hyperbolic sine function. Thus

$$d\epsilon/dt = A \sinh S/S_1. \quad .. \quad .. \qquad (IV.16)$$

where A and S_1 are constants.

In a comparison of this method with that of the log.log method, Nadai points out that one advantage of his equation is that it satisfies the condition of zero creep rate at zero stress.

By the application of any of the methods described above for the extrapolation of creep results, a working diagram can be constructed in which stress is plotted against temperature for a given deformation within a fixed time, and, from such a diagram, the value of a safe working stress at the temperature of operation may be estimated. It should be noted, however, that each of the methods will, in general, yield different working diagrams for the same set of results. Marin (129) drew attention to this fact by comparing the working stress diagrams obtained by several of the methods with that obtained by McVetty's method, and he showed that considerable differences occur.

There is no doubt that the use of empirical relationships can lead to erroneous conclusions with respect to the creep behaviour of materials under the conditions of prolonged loading which occur in service. The engineer, nevertheless, has still to face the problem of deciding how best to apply the results of creep tests to design, and the knowledge that mathematical and graphical extrapolation is unreliable is cold comfort without some other alternative. The alternative is a procedure, not unfamiliar to engineers, and is embodied in a statement by Gillett (5).

> " We can still use common sense, apply proved metallurgical principles and come out with a reasonable guess, which is often pretty good engineering."

CHAPTER V

THE PHYSICAL STUDY OF THE MECHANISM OF CREEP

Introduction

SO FAR attention has been devoted, mainly, to the treatment of the phenomenon of the creep of metals from the engineering standpoint. This has been done because more extensive studies have been made of this aspect of the subject, by reason of the fact that engineering requirements have created an immediate need for knowledge of the behaviour of commercially interesting materials under conditions of service in which creep is an important factor.

To conclude this survey a review is given of the progress which has been made in the physical study of the mechanism of creep. Such a review gives emphasis to the deplorable lack of experimental data which is available in support of theoretical treatments of the subject, and indicates the lines along which future investigation would be of most value.

Creep is a particular form of flow in solid materials. In the case of a polycrystalline metal it appears likely that the individual grains flow in a manner characterized by the flow of single crystals, and that the grain boundaries flow in a manner more characteristic of viscous flow. It seems logical, therefore, that the investigation of the mechanism of creep should begin by attempting to combine the interpretations of the observed facts with respect to the deformation of single crystals, with the knowledge which has been gained from the study of the flow of liquids and gases.

Thus, before dealing with the theories which have been put forward to explain the mechanism of creep, a brief outline is given of information concerning the flow of single crystals and polycrystals which is relevant to the study of creep.

The stage of apparent constant creep rate bears similarity to viscous flow in gases and liquids since both are distinguished by a constant rate of flow under stress at a given temperature. If the stress is divided by the gradient of the stage of apparent constant creep rate, a quantity is obtained which is a direct analogue of the coefficient of viscosity as defined for gases and liquids. It must be remembered, however, that this stage of creep is only apparently constant and that the limits set for the onset of, and departure from, constant rate are dependent, among other factors, upon the sensitivity of the strain measuring apparatus. In addition, the quantity which may be termed the " analogous coefficient of viscosity " is not an independent constant of the material as is the coefficient of viscosity of gases and liquids and, as may be seen from Figure 12, when rate of strain is plotted against load a linear relationship is not obtained. Qualitatively the relationship is similar to that for non-Newtonian liquids which is such that, under conditions of constant temperature, the rate of shear increases more rapidly than does the stress.

According to Maxwell(140) the viscosity of a liquid may be regarded as the product of its elastic modulus and its time of relaxation, and if, in the case of a liquid, the time of observation could be made small in comparison with the time of relaxation a value of the elastic modulus could be determined. Maxwell implied that the difference between the elastic behaviour of a solid and liquid is due to a difference in time of relaxation. In certain types of solid it would appear that this implication is significant, for Kühn(141) has shown how the elastic behaviour of high polymers may be accounted for by assuming a product of a constant elastic modulus with a range of relaxation times; further evidence in support of Maxwell's implication was obtained by Ramen and Venkateswaren(142) who found that for mechanical disturbances of sufficiently high frequencies a liquid behaves essentially as a solid.

In the development of a theory for the creep of metals it is not sufficient to focus attention on a viscous type of flow, for the problem must also be regarded in the light of knowledge which has been gained on the plastic flow of crystalline solids. From experimental observation of the plastic flow of single crystals of metals and crystalline inorganic solids, it has been determined that plastic flow occurs by either, or both, of the processes of " slip " or " twinning," which processes are regarded as being fundamental. In the slip process relative movement of crystal lamellæ takes place;

this movement is usually regarded as being abrupt, it appears to be confined to a succession of certain planes leaving the intervening blocks of crystal undeformed, and it is most commonly recognized by the presence of slip lines formed at the intersections of the slip planes with the surface of the crystal. In the process of twinning a long sequence of neighbouring planes, within the crystal lattice, move relatively to one another by distances which are not an integral multiple of the atomic spacing; the relative motion of successive parallel planes is not random but takes place in such a homogeneous manner that the region in which deformation occurs remains a single crystal.

The mechanism of twinning produced by mechanical means has been the subject of much less theoretical investigation than has the mechanism of slip. The two processes are similar in that plastic flow is produced in both cases by stresses many times smaller than the critical shear stress which would be required to translate atomic planes as a unit. The processes differ in that the conditions for the initiation of a twin appear to be so structure sensitive that no reproducible resolved critical shear stress for twin deformation has been obtained in work upon single crystals; furthermore, the process of twinning is to be contrasted with slip in that it can be produced by static load in only comparatively few substances and it frequently occurs during, or as a result of, recrystallization.

When single crystals are subjected to static loading, the deformation which takes place with time, for values of stress which are below the critical shear stress, is considered to occur by the process of creep. (The critical shear stress may be defined as the stress at which the rate of slip becomes readily measurable with comparatively crude equipment; such a rate of shear strain is, for example, 10^{-6} sec.$^{-1}$). As the definition implies, however, the value of the critical shear stress determined by any one investigator is dependent upon the sensitivity of the apparatus used; thus it would appear that there is no definite line of demarcation between creep and slip and it seems that there is an intimate connection between the two processes.

Seitz and Read(143) have pointed out the fact that in a metal such as bismuth, which apparently exhibits twinning and not slip when strained rapidly beyond the elastic limit, it is quite possible that the creep mechanism has a closer resemblance to twinning than to slip.

The ease with which slip will take place in a crystal is dependent upon the orientation of the planes of easy slip with respect to the applied shear stress. When a single crystal test piece is strained in a tensile machine the constrains of the grips keep the ends in line and the crystal is forced to deform in a manner shown (in the simple case where slip occurs on one set of planes) in Figure 23. It will be noticed that, in the central portion of the test piece, rotation of the lattice planes takes place such that the slip direction becomes more nearly parallel to the axis of tension as deformation proceeds.

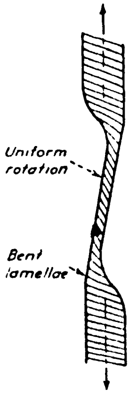


Figure 23.
Rotation and bending in a single crystal elongated in a tensile testing machine. Homogeneous rotation of the necked-down region ends in bend gliding near the grips. (Barrett, ref. 150.)

It has been pointed out that one of the distinguishing features of the slip process is the production of slip bands; thus, if there is an intimate connection between slip and creep it is reasonable to expect that slip bands might appear during the progress of creep in single crystals. Very little attention has been paid to this point in experimental work. Chalmers(89), (144), in his experiments on single crystals of tin, failed to observe any slip bands; it is possible however, that, for the small deformations with which he was concerned, the magnification and resolving power of his microscope were not high enough to reveal the presence of slip bands. On the other hand Dushman(145) has reported that he observed slip bands on creep test pieces of single crystals of aluminium. Hanson(111) described experiments on single crystals of aluminium in which he and Wheeler found that rate of straining was a deciding factor in the production of slip bands. He produced evidence to show that no slip bands occur in a single crystal of aluminium when slowly extended at a rate of 1.8% in 52 days at a temperature of 250° C. and he described the flow, by which deformation took place, as " slip-less " flow. He qualified this description, however, by remarking that further study may show that slip-less flow is essentially the same as normal distortion in which the shear is spread over a large number of planes but is nowhere sufficient to be detected by optical methods.

The importance of attempting to relate the processes of slip and creep on the grounds of the observation of slip bands in single crystal specimens is, however, somewhat in doubt. Information on the characteristics of slip bands is, in itself, far from complete. According to Andrade(146) the displacement at an individual slip plane may amount to as much as 0.1 mm., thus it follows that, at slip bands, shifts amounting to thousands of atom distances may be involved. Andrade and Roscoe(147) found, from experiments on lead, that slip bands were spaced apart with considerable regularity, and they quoted a probability distribution around a spacing of 4.2×10^{-4} cm. for parallel slip bands. They noted that the spacing was independent of the following factors: (1) temperature (between 0° C. and 100° C.), (2) diameter of crystal and (3) rate of deformation (which was varied by a factor of 3,000). In addition, the spacing of the slip bands did not change with increasing deformation in crystals of lead, cadmium and mercury, indicating that, in these metals, the deformation proceeded along, or adjacent to, the original slip planes. On the other hand, had these experiments been carried out at temperatures at which the metals concerned exhibit pronounced strain hardening characteristics, or, more precisely, at temperatures at which the rate of recovery is considerably less (as is the case for aluminium at room temperature) it would be expected that the number of slip bands would have increased and become more closely spaced as deformation proceeded.

Rosenhain(148) observed from an examination of slip in crystals of iron that sharp discontinuities occurred on the surface of the crystals at the intersections with the slip planes, indicating that movement was concentrated on a single plane. Greenland(149), however, found that the slip bands on single crystals of mercury were rounded off at the surface, suggesting that movement had occurred in a series of closely grouped planes. The differences in resolving powers and magnification of the microscopes used in these investigations may account for the different observations, and the more recent work on this subject(150) seems to establish the fact that slip occurs on a group of closely associated planes.

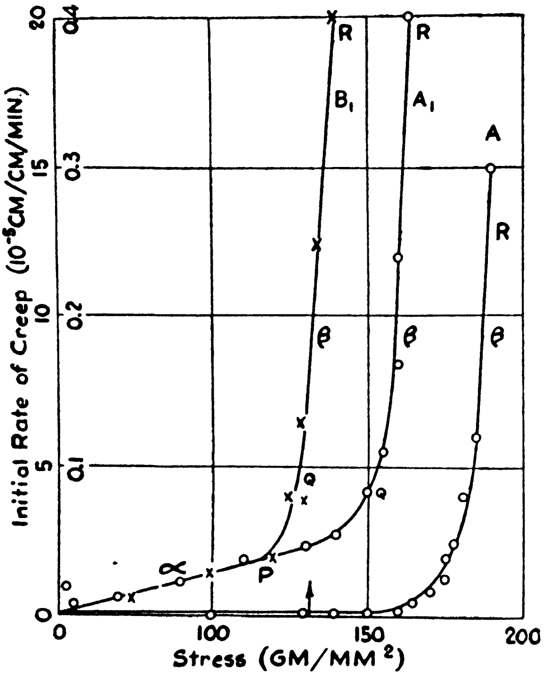


Figure 24.

The dependence of initial creep rate upon stress for specimens of tin. Curves A_1 and B_1 have the right-hand scale of ordinate, curve A has the left-hand one. A_1 and A were obtained for Champur tin, curve B_1 for Williams-Harvey tin. The vertical arrow indicates the critical shearing stress as measured by other workers. (Chalmers, ref. 89.)

It would appear possible, then, that during the course of creep in single crystals small movement may occur by slip on a relatively large number of planes and that the amount of movement on each active plane may be so small that it cannot be detected by means of an optical microscope; this is more likely to be the case in the range of temperature where the rate of strain hardening is greater than the rate of recovery.

Chalmers(89) carried out creep experiments on single crystals of tin. He plotted the rate of initial strain against stress and it may be seen from his curves (Figure 24) that the rate of strain increased linearly with stress for stresses below 100 gm./mm.^{2} (the range which has become known as the range of micro-creep) thereafter the rate of strain increased more rapidly.

The region of the curves where departure from a linear relationship between initial creep rate and stress takes place seems to represent a transition from creep to slip, and more extensive work of this type should prove valuable in relating the two processes.

When information gained from single crystals is applied to the flow of polycrystals many complicating factors have to be considered. The most important of these are the effects of grain boundaries and the constraints imposed on the flow of a grain by the flow of the variously oriented neighbouring grains in relation to the flow of the aggregate. The homogeneous strain of the single crystal becomes replaced by an inhomogeneous strain which varies from grain to grain and from point to point within a grain.

A polycrystalline metal, in general, offers more resistance to deformation than do single crystals(150). The type of strain hardening observable in single crystals plays an appreciable part in the hardening of polycrystals; it is possible, however, that additional unknown factors which are related to the existence of grain boundaries also enter into the hardening process, and that their influence becomes more pronounced as the number of grains per unit volume increases.

During the creep of a polycrystal each grain of the aggregate cannot change its shape in an arbitrary manner, nevertheless each grain tends to elongate in proportion to the elongation of the specimen as a whole and to contract laterally in proportion to the contraction of the specimen(111). Taylor(151), (152) computed the stress-strain curve for polycrystalline aluminium from a knowledge of slip in single crystals of that metal by assuming that, for each of the slip systems which he considered operative, the shear-hardening curve is similar to that of a slip system in a single crystal and, also, that a state of uniform strain exists in all grains. Taylor's computed curves agree well with the observed curves, but experimental determination of the direction of rotation of the grains shows a discrepancy with the direction of rotation predicted by the theory. This discrepancy was attributed to the influence of grain boundaries and the constraints imposed by neighbouring grains.

Very little is known of the nature of grain boundaries(153). Rosenhain favoured the theory that grain boundaries consist of a type of amorphous cement, but modern evidence fails to substantiate this theory. It appears more likely that a grain boundary is merely a transition region where the atom positions represent a compromise between the crystalline arrangements of the adjoining grains(153). A value for the thickness of this transition region may be calculated, and one estimate(154) puts the upper limit at a thickness of five interatomic distances.

Rosenhain(115-118) recognized the importance of viscous grain boundary flow and performed experiments demonstrating the existence of such a flow. Since his work, which was done some 30 years ago, no attempt seems to have been made, until quite recently, to determine the relation between rate of flow across a grain boundary and the stress which causes such flow. In a recent paper, however, T'ing-Sui Kê(203) has described some notable work which he has carried out, at the University of Chicago, in an investigation of the mechanical behaviour of grain boundaries in metals; he has produced experimental evidence which shows that under conditions of shear, at low stresses, grain boundaries behave in a viscous manner; from his measurements he has estimated the value of the coefficient of viscosity of the grain boundary material in a polycrystalline specimen of aluminium at its melting point, and has found it to be of the same order of magnitude as that determined from experiments on the viscosity of the molten metal. In performing his estimations, T'ing-Sui Kê assumed that the thickness of the grain boundary is of the order of magnitude of one atomic distance, thus, this work provides additional evidence of the extreme thinness of the grain boundary region.

The amount of strain at various distances from a grain boundary has been measured(155) and it has been found that near a boundary there is less deformation than in the centre of a grain. Chalmers(75) conducted experiments with tensile specimens composed of two crystals of tin in which the boundary between the pair of crystals extended longitudinally throughout the specimen. His results show that the critical shear stress depends upon the difference in orientation between the two crystals, being a minimum when the lattices in the two crystals are similarly oriented and a maximum when the orientations differ the most. These results indicate that when slip planes in two adjacent grains are not parallel it is difficult for the block movement in one grain to cross the boundary into the next.

The irregular atomic array at the grain boundaries appears to possess higher energy and less stability than the grain interior. Possible evidence in support of this is obtained from observations(156) that atoms diffuse into many polycrystals more rapidly along grain boundaries where the potential barriers, which must be overcome during diffusion, are, presumably, lower than those in the interior of grains. Additional possible evidence is provided from practical experience where, for example, the properties of the boundaries are influenced by impurities which segregate at these places during freezing, and by the penetration of oxygen and corrosive elements along the boundaries; the latter effect, however, is more probably due to electro-chemical action. Chalmers(157) attempted an investigation of the properties of grain boundaries between crystals of pure tin and found that the melting point of the boundary material was about 0.14° C. below the melting point of the bulk material.

Hanson and Wheeler(94) studied the nature of the deformation in polycrystalline specimens of pure aluminium by observing the changes produced in the microstructure. They found that when a specimen was strained slowly (at a rate of approximately 4% in 20 minutes) under an initial stress of 1.4 tons per square inch and at a temperature of 250° C. the crystal boundaries became well developed, but there was no evidence of slip-bands on the surface of any crystal. After 210 hours the same specimen had elongated 21%, by this time the grain boundaries were very prominent and slip-bands were almost, or entirely, absent. Hanson(111) pointed out the difference between this observation and a similar observation made on another specimen which was strained more rapidly (at a rate normally used in short time tests) to 21% at the same temperature. The microstructure of this latter specimen showed prominent slip-bands and the grain boundaries were but moderately well defined. No satisfactory explanation of this difference in behaviour has yet been put forward, but the observation is of sufficient significance to warrant further investigation in which the rate of deformation required for slip-less flow in polycrystals and single crystals should be carefully compared. X-ray examination of the specimens, similar to that carried out by Wood (see below), might also provide important evidence concerning the mechanism of creep.

Wood and Tapsell(158) have described work carried out at the National Physical Laboratory which provides useful information in connection with the mechanism of creep. Under normal static loading of a polycrystalline metal, Wood(159) found that the grains break down into crystallites characterized by widely differing orientations and, for a given metal, a particular lower limiting size. Under similar loading he found(160) that in a single metal crystal a dislocation of the mosaic structure occurs, but that, in general, the mosaic elements remain approximately parallel; this condition he described as the " parallel crystallite formation." It was considered that this difference in behaviour of a polycrystal and of a single crystal might be a factor distinguishing the normal short-time deformation of a polycrystalline metal (in a tensile test with increasing stress) from continuous deformation at the much slower rate in a creep test. This has proved to be the case. A specimen of aluminium, initially annealed, was strained at 300° C., in a normal tensile test, to 0.9%, the extension being completed in approximately 2 minutes. A similar specimen was allowed to creep under a load of $\frac{1}{2}$ ton per square inch at the same temperature, until the same extension was reached, the extension being completed in 50 minutes. X-ray examination of the two specimens showed random crystallite formation in the tensile specimen and parallel crystallite formation in the creep specimen. Thus in creep the polycrystalline specimen deformed by a process characteristic of single crystals.

Such information as that provided by Wood and Tapsell and by Hanson and Wheeler should play an important part in the development of a theory which satisfactorily accounts for the mechanism of creep, and it should guide theoretical investigators in their search for the fundamental processes which are active during creep.

The Theoretical Aspect of Creep

ANDRADE'S WORK:

The work of Andrade[(4)] published some thirty years ago may be regarded as marking the beginning of the study of the physical nature of the creep of metals on a quantitative basis. He carried out tests of short duration, ranging from a few minutes to several hours, on wires of lead, tin, iron, copper, magnesium, mercury (at −78° C.), a lead-tin alloy, brass and German silver. Precautions were taken to avoid increase of stress, due to the decrease in cross-sectional area of the test pieces as extension proceeded, by the use of a hyperbolic weight dipping into a liquid. The extension measurements were made with an extensometer reading to 0.01 mm.

Andrade found that the length " L " of the wire test piece at any time " t " could be expressed by the empirical formula:

$$L = L_o(1 + Bt^{1/3})e^{Kt}. \qquad .. \qquad .. \quad (V.1$$

L_o being the length of the test piece immediately on loading, B the measure of an initial type of flow which decreases with time (as expressed by $t^{1/3}$) and K the measure of the final flow which proceeds viscously. The exponential form of the viscous term is analogous to the case of a very viscous liquid and follows from the condition that:

$$\frac{1}{L} \cdot \frac{dL}{dt} = \text{constant}.$$

It was found that the various forms of extension/time curves, which were observed for different metals at room temperature, could be obtained with the one metal, lead, by suitably choosing the temperature. This fact is of considerable importance and further extensive investigation of the matter over a wide range of metals and alloys would be of great value.

At a given temperature, with increasing stress, Andrade found that the factor B tended to a constant value and the factor K increased at a rate which itself increased to a constant value. From experiments on lead it appeared that the limit to which B tended with stress did not increase with temperature, and at low temperatures this limit could not be attained owing to the wire breaking first.

Andrade(161) maintains that for all pure metals, at temperatures below their recrystallization range, the use of constant stress in place of constant load results in the elimination of the final stage of creep. He presumably draws this conclusion from his experiments on lead; some doubt exists, however, as to whether such a conclusion is justified in view of the short time tests which he carried out. Moore(162) has pointed out that if tests on lead are carried out for long enough periods of time the last stage of creep is entered. The evidence given in Chapter II appears to disagree with Andrade's contention, and Tapsell(18) has provided experimental results in support of the view that other factors, such as the onset of cracking, contribute largely to the causes of a stage of accelerating rate of creep. Additional evidence of a similar kind may be found in the work of Hanson(111).

It is important to bear in mind that during creep tests, particularly at elevated temperatures, structural changes are going on within the specimens and, as Andrade(161) pointed out, at the end of a test, the test specimen is not composed of the same material as it was at the beginning. Thus hopes of establishing a general equation of state for metals grow dim as consideration is given to the influence of such a factor.

The qualitative explanation which Andrade gave, originally, to his results was based on the co-existence of two different phases in his metals: the crystalline phase and the amorphous phase. He has since modified this explanation(161) to conform to an interpretation on the basis of present-day knowledge. He considers that the B-flow represents a type of transient flow associated with the rotation of crystals during the period of stress redistribution (see

Chapter II, and page 84), while the K-flow represents a uniform rate of flow which, by analogy with the theory of viscous flow(163), (164), is governed by an activation energy. Thus an expression for the rate of K-flow in terms of temperature is proportional to:

$$\exp(-E/kT)$$

where E is the activation energy required for flow and T is the absolute temperature.

Following the argument of the Becker-Orowan theory (which is discussed later) the activation energy E necessary for flow in single crystals may be expressed in terms of the applied stress by a relationship of the form

$$E = A(S_0 - gS)^2 \quad .. \quad .. \quad .. \quad (V.2)$$

where S_0 is the critical shear stress necessary to cause slip on a slip plane, S is the applied stress, A is a constant which includes the shear modulus and g is a stress concentration factor.

If, now, the rate of flow is considered to be a function of the applied stress at constant temperature, then rate of strain may be expressed by an equation of the form

$$\frac{d\epsilon}{dt} = C \exp\left\{-\frac{A(S_0 - gS)^2}{kT}\right\} \quad .. \quad .. \quad (V.3)$$

This is the expression which originated in a modification, by Orowan, of Becker's equation (see page 97), and which has found some confirmation from the results of creep tests on single crystals in the region where the applied stress is small(89). It must be emphasized, however, that by representing the K-flow in terms of the above expression, the assumption is made that this type of flow is composed only of block movement within crystals. It is most unlikely that such is the case, and furthermore, the implication that the B-flow accounts for the total grain boundary flow cannot be accepted.

It may be argued that, since the grain boundary is a region of irregular atomic array, grain boundary flow is, presumably, a type of viscous flow with similar characteristics to that of non-Newtonian liquids, and the factor K might be represented in terms of some function combining non-Newtonian flow with the slip type of flow; but, even if consideration of structural stability is disregarded, neither of the factors B nor K take into account such influences as: grain size, strain hardening and recovery, variation in chemical composition, etc. Thus, although it may be possible to establish an empirical formula for the creep of one batch of metal, the same formula will not necessarily apply to another batch of even the same material, unless the empirical equation takes account of at least the variation of such factors as have been discussed in Chapter III.

The establishment of empirical equations from experimental work is not, therefore, of great value in helping to discover the mechanism of creep. The theoretical approach to the subject must begin by considering flow from a fundamental standpoint and by reasoning out first the mechanics of the simplest cases. Such an approach was made by Becker who attempted to explain, quantitatively, the flow in single crystals.

BECKER'S THEORY(165):

In order to account for the fact that the observed critical shear stress in single crystals is of the order of 10^{-3} of the value calculated on the assumption that slip occurs by the relative block movement of two adjacent planes of atoms in the lattice(165), (166), (167), (168), Becker put forward the theory, in 1926, that since the atoms in the lattice are in a state of oscillation at temperatures above absolute zero, it is possible that these oscillations influence the shear strength of crystals. He pointed out that the stress at any point within a crystal will not be constant at finite temperatures but will fluctuate as a result of the thermal oscillations of the atoms and, occasionally, local fluctuations should become sufficiently large to cause groups of atoms to slip past one another and so produce a small amount of local slip.

The elastic energy per unit volume in a region of a crystal where the shear stress is equal to " S " is:

$$\frac{S^2}{2G}$$

where G is the shear modulus of the material of the crystal. In accordance with the Boltzmann theorem(169), (170), (171) (which states that when a body is considered as made up of a large number of particles having a definite total energy and an absolute temperature " T," the relative probability that any particle is in an " accessible " state having an energy E is given by:

$$p = C \exp\left(-\frac{E}{kT}\right) \quad \ldots \quad \ldots \quad \ldots \quad (V.4)$$

where k is Boltzmann's constant), the relative probability that a given volume " V " of the crystal will be under a stress " S " (in the absence of an externally applied stress) as a result of thermal fluctuations is approximately:

$$C \exp\left(-\frac{S^2V}{2GkT}\right).$$

If, now, the crystal is subjected to an externally applied stress of average value S′ then, on the assumption that the external stress acts constantly on all atoms, for slip to occur in the direction of the applied stress it is only necessary for thermal fluctuations to add a stress equal to $(S_o - S')$, where S_o is the value of the critical shear stress, and the relative probability that slip will occur in the direction of the applied stress is:

$$C \exp\left\{-\frac{(S_o - S')^2V}{2GkT}\right\}$$

The rate of flow in the direction of the applied stress is then given by the equation:

$$\frac{d\epsilon}{dt} = C' \exp\left\{-\frac{(S_o - S')^2V}{2GkT}\right\} \quad \ldots \quad \ldots \quad (V.5)$$

where C′ is a constant involving the reciprocal of the time required for the atomic planes to slip past one another. This is the equation which Becker derived.

The physical interpretation of the volume V in Becker's equation is puzzling. Orowan(168) has given an interpretation in terms of the radius of a stable dislocation line, the dislocation being considered to start from a point and to spread out in a fan-like fashion as it moves forward. The significance of the volume V(183), (194) and (195), is that its value must be such that the thermal stress fluctuation occurring within it gives rise to a local slip process which can spread over the slip plane; if this volume is too small, then, after the stress fluctuation has ceased, the propagation of local slip is arrested by the elastic forces in the surroundings.

When slip occurs in the ordinary static test, if it is assumed that $d\epsilon/dt$ is constant (it should be pointed out, however, that the temptation to make this assumption arises only from the probable insensitivity of extensometer equipment used, and that there is no concrete evidence for making the assumption) then, from Becker's equation, the following relationship must be satisfied:

$$S' - S_0 = -\text{Const.}\ (T)^{1/2} \quad .. \quad .. \qquad .. \quad (V.6)$$

From this equation it follows (as would be expected, of course) that S_0 should be the value of the critical shear stress at absolute zero of temperature.

Schmid and Boas(172) studied the temperature dependence of the critical shear stress for zinc and cadmium and the relationship predicted by equation *(V.6)* is in qualitative agreement with their observations. Unfortunately, however, the value of S_0 obtained by extrapolating the observed points to absolute zero of temperature is of the same order (namely 10^7 dynes/cm.^{2}) as the values of S' at room temperature. Thus it has to be concluded that imperfections introduced by local thermal fluctuations are not sufficient to account for the low values of observed shear strength of crystals.

Smekal(173) suggested that the observed low values for the shear strength of crystals is intimately related to the existence of the mosaic structure. He considered that the mosaic boundaries are a source of weakness, thus the variations in structure within the mosaic pattern allow for variations in stress from one region to another and, at certain locations, miscroscopic stress concentration is set up. In this connection it is worth noting that Griffith(174) explained the discrepancy between the microscopic and macroscopic tensile strength of brittle solids on the basis that in the neighbourhood of minute internal cracks, or other defects, local stress concentration is set up.

Masing(175) observed that the difficulty encountered in the application of Becker's equation at absolute zero of temperature is removed if it is assumed that mosaic boundaries are responsible for the introduction of a stress concentration factor of the order of a thousand at some point within the crystal.

Working on the basis of the well-known "Dislocation Theory"(176), (177), (178), (179), (180), (181), (182) Orowan(178) regarded dislocations as being responsible for the existence of stress concentration and accordingly he modified Becker's equation to the form:

$$\frac{d\epsilon}{dt} = C \exp \cdot \left\{ - \frac{(S_o - gS')^2 V}{2GkT} \right\} \quad \ldots \quad \ldots \quad (V.7)$$

where g is the stress concentration factor.

There are fewer objections to this form of equation than there are to the original equation; it is to be noted, however, that, for a constant applied stress, equation *(V.7)* requires that the creep rate shall have a constant value; this fact cannot be reconciled with the form of the Andrade Curve, and, in view of the fact that the phenomenon of strain hardening is regarded as having considerable influence on the progress of creep, it is not surprising to find that further modification of the Becker equation should be attempted by the introduction of a term to represent strain hardening. Orowan(183) has shown that the most obvious way of introducing such a term is by making the simple assumption that the increase in critical shear stress, caused by strain hardening, is a function of plastic strain only, and by adding this function to the term S_o. In this manner equation *(V.7)* now becomes:

$$\frac{d\epsilon}{dt} = C \exp \left\{ \frac{-(S_o + f(\epsilon) - gS')^2 V}{2GkT} \right\} \quad \ldots \quad (V.8)$$

and this equation, as Orowan has remarked, is a special case of the more general relationship

$$\frac{d\epsilon}{dt} = f(\epsilon,S,T) \qquad .. \qquad .. \qquad .. \; (V.9)$$

suggested by Ludwik(138).

Zener and Hollomon(184) have examined this suggestion put forward by Ludwik, and they consider that such a mechanical equation of state exists for metals in regions of temperature and strain where factors such as grain growth, precipitation, recovery, recrystallization, etc., are not operative. If this be the case, then rate of strain in a metal depends upon the instantaneous values of strain, stress and temperature, and not upon their past values. In order to establish the existence of an equation of state,

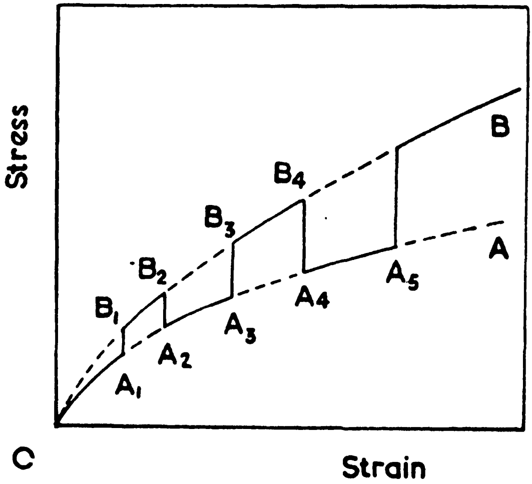


Figure 25.
Curves illustrating the method used for investigating the possibility of existence of an equation of state for metals. (Orowan, ref. 183.)

Hollomon(185) carried out a series of experiments on mild and alloyed steels. Reference to Figure 25 will show the principle upon which Hollomon worked. OA and OB represent the stress-strain curves for a given material at two different temperatures; suppose the straining of a specimen is begun at the higher temperature along the curve OA to the point A_1 where it is interrupted and continued, from the point B_1, on the curve OB at the lower temperature. If an equation of state exists, alternating the conditions of straining in this manner will yield a plot of the segments OA_1, B_1B_2, A_2A_3, B_3B_4, etc., of the curves OA and OB. Hollomon found that his experimental results gave a plot of segments of curves lying between OA and OB, but he considered that these curves were sufficiently close to OA and OB to justify the conclusion that over a wide range of strain at low temperatures, and for small strains at elevated temperatures, the stress required for plastic flow depends only upon the instantaneous values of strain, strain rate and temperature, and not upon the past history of these variables.

Orowan(183) has discussed the subject of a mechanical equation of state for metals in terms of an experiment of the type carried out by Kochendörfer(186) on the plastic deformation of naphthalene crystals. In Figure 26 the co-ordinates of the point P represent a

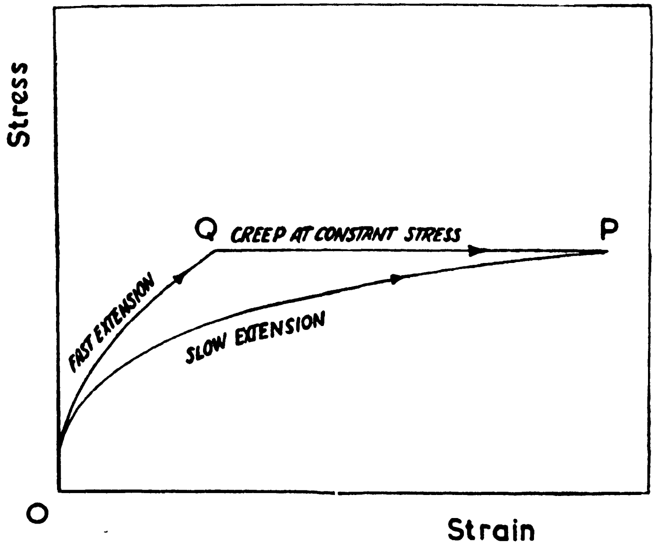


Figure 26.

Curves illustrating an experiment which shows that strain rate is a function of past values of strain in addition to the instantaneous values of strain, stress and temperature. (Orowan, ref. 183.)

given state of stress and strain in a material at a fixed temperature. Let the conditions represented by the point P be reached in a first specimen by a constant slow rate of straining along the curve OP, and, in a second specimen, by rapid rate of straining along the curve OQ to the point Q followed by creep at constant stress along QP. Experiments show that the creep rate at the point P in the second case is much less than the slow rate of straining at the point P in the first case. This evidence, and the additional evidence which Orowan has provided from results of experiments, conducted by Los and himself(183), in which the behaviour of high conductivity copper and pure aluminium was studied, under conditions similar to those used by Hollomon in the case of steels, shows fairly conclusively that Hollomon was unjustified in his conclusions regarding the existence of an equation of state for metals, and that such an equation does not exist even approximately.

THE RECOVERY THEORY(183):

When describing the characteristics of the normal creep curve in Chapter II, it was shown how the conception, put forward by Bailey, of the progress of creep representing a running balance between strain hardening on the one hand, and temperature softening (or recovery) on the other, gave a plausible, qualitative explanation of the form of the creep curve in very broad terms. According to this conception, the process of strain hardening involves some mechanism which is responsible for the creation of undefined obstacles, within the crystal lattice, which impede the progress of deformation, and the process of recovery involves an opposing mechanism which supplies the means of removing these obstacles.

Orowan(183) has provided an account of the development of this conception in what has now become known as the Recovery Theory, and he has shown that such a theory cannot account for the behaviour of metals under conditions of creep.

In the Recovery Theory it is assumed that, in the absence of thermal softening, the relationship between stress and plastic deformation may be represented by an equation of the form:

$$S = f(\epsilon) \quad .. \quad .. \quad .. \quad .. \quad (V.10)$$

where S is the stress required for the further deformation of a specimen which has been strain hardened during preceding deformation of magnitude ϵ. A coefficient of strain hardening, H, may now be defined by the slope of the stress-strain curve, at the point (ϵ, S), that is, by the derivative of f(ϵ) in equation *(V.10)*. Thus

$$H = \frac{dS}{d\epsilon} \quad \cdots \quad (V.11)$$

The removal of strain hardening, by temperature softening, may be measured by the reduction of yield stress per unit time, thus giving the rate, r, of thermal recovery, at a given temperature, in the form:

$$r = \frac{dS}{dt} \quad \cdots \quad (V.12)$$

It follows, therefore, that the decrease in yield stress, in a time interval dt, due to thermal recovery is given by the quantity r.dt. In the same interval of time, and under conditions of constant applied stress, the specimen must deform by an amount represented by a strain dϵ, such that the decrease in yield stress, r.dt, is just compensated by the increase in yield stress, H.dϵ, due to additional strain hardening. This is expressed in the equation:

$$r.dt = H.d\epsilon \quad \cdots \quad (V.13)$$

from which it follows that the creep rate is given by:

$$\frac{d\epsilon}{dt} = \frac{r}{H} \quad \cdots \quad (V.14)$$

An immediate difficulty which arises in connection with equation *(V.14)* is that of expressing the term r as a function of the factors which influence the rate of thermal recovery. It is evident that r cannot be a function of stress and temperature only, for, if this were the case, equation *(V.14)* would demand a constant creep rate under conditions of constant stress and temperature, and this cannot be reconciled with the initial stages of creep. If it is to be assumed that r varies with time, then a complete understanding of the phenomenon of recovery is essential before an expression for r can be found. Until such understanding is established, as Orowan has remarked, the use of the term r only introduces an auxiliary variable which does not help in the development of the theory.

From an experimental point of view, it is well known that slight variations in stress, which have negligible effect on the slope, H, of the stress-strain curve, often have considerable effect on the creep rate. Furthermore, determinations of the values of the terms r and H(178) show that the two sides of equation *(V.14)* can differ by many powers of 10. Orowan has also drawn attention to yet another difficulty in connection with this theory; if at any moment, during creep, the stress is suddenly decreased by a small amount, then, if the theory is correct, creep should cease immediately, because the stress would now be below the yield stress of the material; further deformation should set in equally suddenly when the yield stress has fallen, due to thermal softening, to the reduced value of the applied stress. Experiments carried out by Orowan and Los(183) have shown that, in actual fact, a small reduction of stress merely reduces the creep rate and that no trace of a temporary arrest in the progress of creep can be observed. Evidence in support of this result is found from the work of Lea(187).

SUMMARY OF THE RESULTS OF INITIAL STUDIES OF THE MECHANISM OF CREEP

Before proceeding with an account of the more recent theories concerned with the mechanism of creep it is worth while summarizing the results of the initial studies which have been described.

Andrade established the form of the creep curve for all pure metals (see Figure 6) and showed that the various types of curve, observed for different metals at room temperature, could be obtained with one metal by suitable choice of stress and temperature. This form of curve has become known as the "Andrade Curve" and the question of whether, or not, it is valid for prolonged periods of straining is a matter of little importance at this stage of theoretical development.

Andrade showed that the creep curve can be resolved into two components; one component, the K-component, represents a type of flow which proceeds in a viscous manner, the other, the B-component, represents a transient type of flow which starts rapidly at the origin and dies away to zero in a relatively short time. At temperatures which are low relative to the melting point of a metal, the flow is almost entirely transient; at higher temperatures the transient component is soon masked by the viscous component. (It has become the custom, nowadays, to describe the K-component as quasi-viscous flow, since it does not obey the condition, that strain rate is proportional to stress, which characterizes Newtonian viscosity.)

Neither the Becker Theory nor the Recovery Theory explain the decreasing rate of creep shown by the Andrade curve. This difficulty may be overcome by modifying the Becker-Orowan formula to take account of the phenomenon of strain hardening; this procedure, however, leads to the conclusion that an " equation of state " for metals should exist and experiment shows that this is not so.

In the description, which follows, of the more recent theories of creep, it will be seen that Andrade's resolution of the creep curve into two components of flow is regarded as being of fundamental significance, and that particular attention has, so far, been devoted to the study of transient flow.

Recent Studies of the Mechanism of Creep

KAUZMANN'S THEORY(188):

Kauzmann published a discussion of an equation in which creep rate is expressed as a function of temperature and stress. The argument leading to the development of this equation is based on the general theory of shear rates which Eyring(189) applied to the viscous flow of liquids. Kauzmann began by considering a " unit of shear process " which involves the motion past one another of " units of flow "—units of flow being regarded as generalized elementary structures within a solid, whose motions constitute the shear process. It was assumed that in order for two units of flow to pass one another an energy barrier must be overcome so that the unit shear process cannot take place unless the units of flow become activated. Thus the unit shear process was regarded as taking place in the form of a jump as two units of flow pass one another and, if λ is the average distance through which two units of flow move relative to one another in one jump, D is the average distance between layers of units of flow and N the net number of jumps per second made in the direction of shear, then:

The shear rate is equal to

$$\frac{d\epsilon}{dt} = \frac{\lambda N}{D} \quad \ldots \quad \ldots \quad \ldots \quad (V.15)$$

The value of N is dependent upon the number of activated units of flow present and, in accordance with the Boltzmann theorem (see page 95) N is proportional to:

$$\exp. (-E/kT)$$

where E, k, and T have the same meanings as before.

In the absence of applied stress a state of statistical equilibrium exists between the activated units of flow, and the number of jumps in any one direction is equal to those in the opposite direction.

According to the theories of Andrade(190) and Eyring(189):

$$N = \frac{kT}{h} \exp\left(-\frac{\Delta F}{RT}\right) \quad \ldots \quad \ldots (V.16)$$

where ΔF is the difference in free energy of the units of flow in the normal and activated states, k is Boltzmann's constant (in ergs per degree absolute), h is Planck's constant (in erg-seconds) and N is the number of jumps per second made by one unit of flow past other units of flow.

If ΔE and $\Delta \Theta$ are the changes in energy and entropy respectively, associated with the unit shear process then:

$$\Delta F = \Delta E - T.\Delta\Theta \ldots \quad \ldots \quad \ldots (V.17)$$

When an external shear stress is applied, Kauzmann assumed that the activation energy for motion of units of flow in the direction of shear (referred to as the positive direction) is lowered by an amount proportional to the applied stress, whilst that in the opposite direction (referred to as the negative direction) is raised by an equal amount.

Thus:

$$N'(+) = \frac{kT}{h} \exp\left(-\frac{\triangle F}{RT} + \frac{aS}{kT}\right) \quad \ldots \quad \ldots (V.18)$$

$$N'(-) = \frac{kT}{h} \exp\left(-\frac{\triangle F}{RT} - \frac{aS}{kT}\right) \quad \ldots \quad \ldots (V.19)$$

where S is the applied stress and "a" is a constant of proportionality.

If N is the net number of jumps in the positive direction per unit of flow per second, then

$$N = N'(+) - N'(-) = \frac{2kT}{h} \exp\left(-\frac{\triangle F}{RT}\right) \sinh\left(\frac{a}{kT} \cdot S\right) \ldots (V.20)$$

Substituting in equation *(V.15)*, the shear rate is given by:

$$\frac{d\epsilon}{dt} = \frac{2\lambda kT}{Dh} \cdot \exp\left(-\frac{\triangle E}{RT}\right) \exp\left(\frac{\triangle \Theta}{R}\right) \sinh\left(\frac{a}{kT} \cdot S\right) \quad \ldots (V.21)$$

This is the general equation for shear rate developed by Eyring and applied to the phenomenon of creep by Kauzmann.

For small values of the externally applied shear stress this equation reduces to the form:

$$\frac{d\epsilon}{dt} = K \exp\left(-\frac{\triangle E}{RT}\right) \exp\left(\frac{\triangle \Theta}{R}\right) \cdot S \quad \ldots (V.22)$$

That is, the shear rate is proportional to the shear stress which is the condition characterizing Newtonian flow.

For large values of S on the other hand, the equation takes the form:

$$\frac{d\epsilon}{dt} = K \exp\left(-\frac{\triangle E}{RT}\right) \cdot \exp\left(\frac{\triangle \Theta}{R}\right) \exp\left(\frac{a}{kT} \cdot S\right) \quad (V.23)$$

which leads to the empirical equation which was derived by Ludwik(138) namely:

$$\log(\text{rate of strain}) = \mathrm{A} + \mathrm{BS} \quad .. \qquad .. (V.24)$$

where A and B are constants at any given temperature and for any given material.

The Kauzmann theory goes no further than the Becker theory in elucidating the mechanism of plastic flow in metals. It is worthy of note in so far as it presents another approach to the subject but the interpretation of the various terms involved (e.g., Θ and " a ") is vague, and in the discussion on the theory(188) Kauzmann himself points out that his explanations are based on generalizations for which there is no experimental justification.

In the application of the equation to experimental data(145), (188), it was found necessary to assume that the constant " a" varied with temperature, but no satisfactory interpretation is evident for such a variation.

Seitz and Read(143), (154) applied the Kauzmann theory to the motion of dislocations which they specified as the units of flow. The argument used was that since the Dislocation Theory is the most successful theory to-date in accounting for the slip process and, since the processes of creep and slip appear to pass continuously from one to the other as the stress is raised, it is reasonable to consider the mechanism of creep as based on the motion of dislocations of a type defined by Taylor(180).

Seitz and Read consider the resolution of the creep curve (i.e., the strain/time curve) into two component curves, one of which passes through the origin, is linear and represents a steady rate of flow, the other, starts linearly at the origin and asymptotically approaches a constant value; this latter curve is analogous to a type of transient flow since its contribution to the velocity is negligible at long times (c.f. Andrade's resolution of the creep curve). They point out, however, that there is good evidence(172), (191), (192), (193) against this type of resolution being generally applicable, since the onset of the stage of increasing creep rate makes it impossible to resolve the creep curve in this manner.

Using similar theoretical reasoning to that used by Kauzmann, Seitz and Read derive the following equation for the transient flow:

$$\frac{d\epsilon}{dt} = \frac{N\lambda^2}{t} \exp\left(-\frac{E}{kT}\right) \sinh\left(\frac{a}{kT} . S\right) \qquad .. (V.25)$$

where N is the density of dislocations cutting across a plane which is normal to the slip plane and extends in the slip direction; λ is the slip distance associated with the passage of a dislocation, and " t " is the time required for the transition from one equilibrium position to the next in the event that a dislocation receives the required activation energy E.

During the steady state flow, Seitz and Read consider that dislocations are generated and annihilated at the same rate. They point out that the steady state rate is strongly dependent upon temperature and appears to obey the law:

$$\frac{d\epsilon}{dt} = A \exp\left(-\frac{E}{kT}\right) .. \qquad .. \qquad .. (V.26)$$

where A is a constant and the other symbols have the same meaning as heretofore.

This modification of the Kauzmann theory does not lead to a satisfactory expression for creep rate. It is not possible to assign values to the terms, such as N, λ, t and a, which appear in equation *(V.25)*, and although this equation has been fitted to Chalmer's results(89) in the range of micro-creep, Seitz and Read have pointed out that a simple hyperbolic function for stress is inadequate in the region of Chalmer's curves, Figure 24, where deviation from linearity occurs in the relationship between initial creep rate and stress.

OROWAN'S THEORY OF TRANSIENT CREEP(183):

Orowan(183) has suggested a new theoretical approach to the problem of creep which does not attempt to provide a detailed molecular picture of the process; it makes no assumptions as to the mechanism of strain hardening but merely assumes that such a process exists. It is not concerned with the discrepancy between theoretical estimates and actually measured values of yield stress, but accepts as a fact that, for some reason, plastic deformation takes place even at the lowest temperatures and at stresses far below the estimated shear strength of crystals.

According to Orowan, the basic law of plastic deformation of a material is that given by the stress-strain relationship at absolute zero of temperature. Suppose the curve OPQ, in Figure 27, represents the stress-strain relationship for a given material at absolute zero. The slightest increase in stress, ΔS, above that represented by the point A will produce further, slight, plastic deformation $\Delta\epsilon$ beyond that represented by AP; therefore, Orowan suggests, the small macroscopic increase in deformation $\Delta\epsilon$ represents the total contribution of all small volumes within the material where the activation energy, required for further deformation, is small and, as a result, local increased deformations have taken place. In order to continue with the increase in strain, beyond that represented by AP, it is necessary to increase the stress in accord-

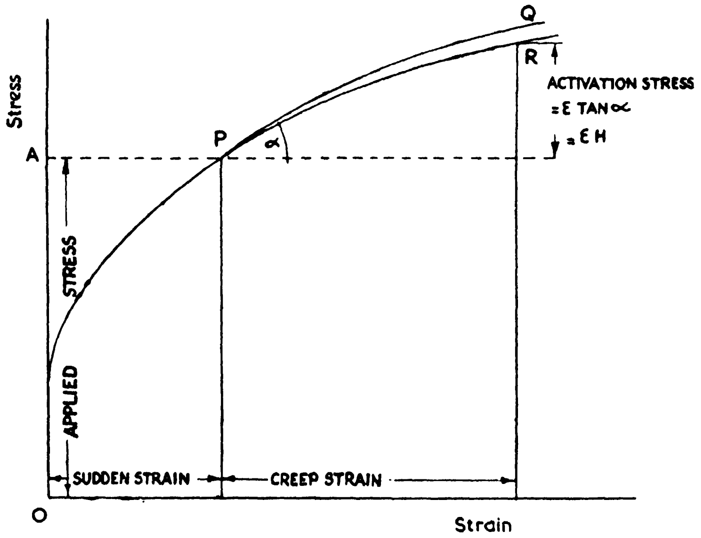


Figure 27.
The stress-strain relationship of a material at absolute zero of temperature. (Orowan, ref. 183.)

ance with the stress-strain relationship; thus Orowan's conception leads to the conclusion that the volume of the material is composed of a large number of small volumes which require a distribution of activation energies, for further plastic deformation, ranging from very small values up to some maximum value.

In a test, at a temperature above absolute zero, in which a stress is applied to a specimen and then held constant, if the effects due to the inertia of testing equipment are neglected, the sudden application of the stress will immediately produce the same strain as would be produced at absolute zero, and, in addition, a distribution of small volumes will appear, throughout the volume of the specimen, for which the activation energy required for further plastic deformation will vary in a manner already described. If, now, it is assumed, as in the Becker theory, that thermal fluctuations are responsible for the production of local stress concentrations, then, immediately following the application of the external stress, these local stress concentrations become effective in producing the required activation energies for further local deformations. Each local deformation which takes place contributes a given amount to the macroscopic deformation of the specimen, and the local value of the stress required to produce such a deformation has been called, by Orowan, the activation stress.

In connection with the conception of activation stress, Orowan has pointed out that its magnitude depends on the magnitude of the volume in which it is produced. Corresponding to a given local increase in stress there is a certain volume for which the production of local deformation has the highest probability, and Orowan regards this most probable value as being given by a mean activation stress.

The continued macroscopic deformation of a specimen under constant stress (that is, the creep of the specimen) is, therefore, visualized as occurring by the continual contribution of local deformations within the volume of the specimen. Suppose after a given time, activation stresses up to a value ΔS have been exhausted by the process of local deformations. If, at this moment, the absolute zero temperature yield stress of the specimen could be determined, it would be found to exceed the corresponding value before creep by an amount approximately equal to ΔS. Thus, at each moment during the creep process, the zero temperature yield stress of specimen rises by an amount equal to the lowest activation stress still available (that is, the material strain hardens). But, Orowan explains, this rise in yield stress with creep strain will not exactly follow the zero temperature stress-strain law, because the action of the process of creep is that of eliminating activation stresses, whereas the application of external stress creates a complete distribution of activation stresses. Thus, instead of the stress-creep strain relationship following the curve PQ in Figure 27, it will follow a curve, such as PR, of somewhat different slope.

As in the Recovery Theory, the slope of the zero temperature stress-creep strain curve, PR, defines the coefficient of strain hardening, H, and after a small creep strain ϵ, the zero temperature yield stress increases by an amount $H\epsilon$. Thus the lowest activation stress available at this point is $H\epsilon$. For large creep strains Orowan assumes that the activation energy for deformation increases approximately proportionally to the creep strain.

Following the argument used in the development of the Becker theory (cf. page 95), the number, N, of local deformation processes taking place in unit time is given by

$$N = C \exp\left\{-\frac{H^2\epsilon^2 V}{2GkT}\right\} \quad \ldots \quad \ldots (V.27)$$

and the creep rate is given by

$$\frac{d\epsilon}{dt} = N\lambda \quad \ldots \quad \ldots \quad \ldots (V.28)$$

where λ is the macroscopic strain produced by one local deformation process.

The difficulty which now arises is that of fixing a value for λ. It cannot be assumed, as in the Becker theory, that local deformation, caused by thermal stress fluctuations, is equal to one interatomic distance, and that this deformation is propagated over the whole cross-section of the specimen. The picture of the process, which Orowan presents, is that, at the instant when the loading of the specimen is completed, the resistance to deformation just balances the applied stress; owing to the presence of small volumes where the activation stress for further deformation is small (the probability of existence of such volumes will be high) a high rate of creep begins at once due to the rapid exhaustion of low activation stresses. Orowan likens the process to a glide avalanche which, initially, spreads over the whole specimen consuming all low activation stresses as it goes, and, as the gap between applied stress and yield stress widens, the propagation distance of the glide decreases due to the existence of barriers of high activation energy.

From a study of experimental creep curves, Orowan has observed that by making the assumption that λ is inversely proportional to the square of the creep strain, an equation for creep rate can be developed which is in best agreement with measured values. Thus, Orowan suggests, the creep rate, $\frac{d\epsilon}{dt}$, after a creep strain ϵ, is given by the following equation:

$$\frac{d\epsilon}{dt} = \frac{A}{\epsilon^2} \exp\left\{-\frac{H^2\epsilon^2V}{2GkT}\right\} = \frac{A}{\epsilon^2} \exp\left\{-\frac{B}{T}\epsilon^2\right\} \quad .. (V.29)$$

where A and $B\left(= \frac{H^2V}{2Gk}\right)$ are constants provided that the creep strain remains small enough for the slope H, of the stress-creep strain curve, to be regarded as constant.

The relationship between creep strain and time is derived from equation *(V.29)* by integration

$$At = \int_0^{\epsilon} \epsilon^2 \exp\left(\frac{B}{T}\epsilon^2\right) d\epsilon$$

$$= \frac{T}{2B}\epsilon \exp\left(\frac{B}{T}\epsilon^2\right) - \frac{T}{2B}\int_0^{\epsilon} \exp\left(\frac{B}{T}\epsilon^2\right) d\epsilon.$$

or

$$\frac{2BAt}{T} = \epsilon \exp\left(\frac{B}{T}\epsilon^2\right) - \int_0^{\epsilon} \exp\left(\frac{B}{T}\epsilon^2\right) d\epsilon \quad .. (V.30)$$

In equations *(V.29)* and *(V.30)*, B is the only term which contains H and is, therefore, the only term which depends on stress. Orowan has pointed out that, although H is probably different from the slope of the stress-strain curve, it is certain that the curves PQ and PR, in Figure 27, are of the same character. Thus, for metals with stress-strain curves similar to OPQ (e.g., aluminium and copper), if sufficiently small values of stress are chosen, the value of H, and therefore the value of B, can become as large as is desired. So that, with increasing temperature, the value of $\frac{B}{T}$ can be kept constant by suitable reduction of the applied stress.

The constants A and B can be determined empirically, and Orowan has shown that the theoretical curve, for creep strain against time, is almost identical with the single experimental curve obtained by Los, from short-time creep tests on copper and aluminium at different temperatures and stresses, by choosing the stress, at each temperature, to be such that the value of $\frac{B}{T}$ remained constant.

The observation that, in general, provided no structural changes occur during creep, a given transient creep curve can be obtained with any material, at any temperature (or stress), by suitable choice of stress (or temperature), follows immediately from this work. This observation was first made by Andrade and reference is made to it on page 91.

Examination of equation *(V.30)* shows that, in order to account for the total strain of a specimen, a third constant is involved which represents the point at which sudden strain is assumed to end and creep to begin. Thus Orowan's equation, like that of Andrade, contains three empirical constants, and, in the region of $\epsilon = O$, the exponential factor in equation *(V.29)* varies slowly compared with $\frac{1}{\epsilon^2}$, so that, regarding this factor as approximately constant, equation *(V.29)* becomes

$$\epsilon^2 d\epsilon = \text{const. } dt$$

or, by integration,

$$\epsilon = \text{const. } t^{\frac{1}{3}} \quad \ldots \quad \ldots \quad \ldots (V.31)$$

This is identical with Andrade's equation for transient creep. Orowan considers, therefore, that his equation *(V.30)* is equivalent to Andrade's equation in the initial stages of transient creep; he states, however, that the exponential factor appearing in equation *(V.29)* provides a creep rate relationship which fits the rapid disappearance of transient creep better than does Andrade's equation.

In the description of his theory, Orowan showed(183) how it is related to the Becker theory and to the Recovery Theory. In the Becker theory thermal stress fluctuations are assumed to raise the value of the applied stress locally to the value of the yield stress at the absolute zero of temperature, whereas, in this new theory, this value of stress is assumed to be reached at the moment when creep at constant stress begins, and thermal stress fluctuations represent additional, transient, local stresses which produce continued deformation and, thereby, increase the yield stress in accordance with a stress-creep strain relationship. In the Recovery Theory, creep is attributed entirely to the removal of obstacles by thermal agitation; according to this new theory, however, creep can arise by thermal stress fluctuations providing the means of overcoming these obstacles without removing them.

SMITH'S MODIFICATION OF OROWAN'S THEORY(194)

Orowan's theory of transient creep has, so far, led to the derivation of an equation for creep rate in terms of the two empirical factors $\frac{A}{\epsilon^2}$ and $\exp\left(-\frac{B\epsilon^2}{T}\right)$. While this equation has been obtained from consideration of a theoretical model of the process of transient creep, the presence of these empirical factors is a handicap in establishing the validity of the model, and, at this stage, it cannot really be said that any advance has been made in the understanding of the mechanism of creep, beyond Andrade's experimental study of the subject. It seems that, for the present, the factor involving the contribution of local deformations to the macroscopic strain must remain empirical, but the factor $\exp\left(-\frac{B}{T}\epsilon^2\right)$, is capable of replacement by one which is rationally derived, and this Orowan stated to be the next step in the development of his theory. Such a step has been attempted, recently, by Smith(194).

Smith's treatment of transient creep is similar to that proposed by Mott (see Mott's theory on page 117). He argues that if E represents the activation energy required for the local deformation of a small volume within a material subjected to conditions of constant stress and temperature, then the number of small volumes with activation energy E will depend upon the value of E; thus the number of small volumes with activation energies between E and (E + dE) may be represented by the function f(E)dE. (cf. Mott's theory, page 122.)

During the process of plastic deformation, the distribution of activation energy is continually changing, that is, f(E) changes with time. Thus Smith considers a function, f(E, t) as representing the distribution of activation energy at time t after the onset of creep. At time t, therefore, the number of small volumes with activation energies between E and (E + dE) will be f(E, t)dE. At time (t + dt), this number will have decreased, due to local deformations, by an amount, dN, given by:

$$dN = C.f(E,t)dE \exp. \left(-\frac{E}{kT}\right) dt \ldots \qquad \ldots (V.32)$$

provided that no new small volumes with activation energies between E and (E + dE) are created during the process. Here Smith follows Orowan in making the assumption that this is the case.

The quantity dN represents the change in the number f(E, t)dE during the time dt. Therefore

$$dN = d\left\{f(E,t)dE\right\} = -\ C.f(E,t)dE. \exp. \left(-\frac{E}{kT}\right)dt \quad \ldots (V.33)$$

(the minus sign on the right-hand side indicates the decrease).

Equation *(V.33)* has a solution of the form:

$$f(E,t) = f(E,O) \exp(-\alpha t)$$

where f(E, O) represents the activation energy distribution function, of the type E, (E + dE), at time t = O and $\alpha = C \exp\left(-\frac{E}{kT.}\right)$

If λ is the contribution of one local deformation to the macroscopic strain of the specimen, then the macroscopic strain produced in time dt, from small volumes with activation energies between E and (E+dE) is given by

$$d\epsilon = \lambda dN = -\lambda.C.f(E,t)dE.\exp\left(-\frac{E}{kT}\right)dt$$
$$= -\lambda.C.f(E,O).\ \exp\ (-\ \alpha t).\ \exp.\left(-\frac{E}{kT}\right).\ dE.dt \quad ..\ (V.34)$$

An expression for creep rate is obtained by integrating over all possible activation energies E. Thus

$$\frac{d\epsilon}{dt} = -\lambda C\int_{E=O}^{E=\infty} f(E,O)\ \exp\ (-\ \alpha t)\ \exp\left(-\frac{E}{kT}\right)dE \quad (V.35)$$

Smith points out that the integral in equation *(V.35)* cannot be evaluated unless the function f(E, O) is known, and neither theoretical nor experimental work gives any indication of its form. In order to obtain a simple expression under the integral sign, Smith assumes that there is an equal number of small volumes within equal intervals of activation energy, dE, at all energy levels. On this assumption, therefore, f(E, O) = constant = p, and the creep rate at time t is given by:

$$\frac{d\epsilon}{dt} = -\lambda Cp\int_{O}^{\infty} \exp\ (-\ \alpha t)\ \exp\left(-\frac{E}{kT}\right)\ dE \quad (V.36)$$

But $\alpha = C\exp\left(-\frac{E}{kT}\right)$ and, therefore $dE = -kT\left(\frac{d\alpha}{C}\right)\exp\left(\frac{E}{kT}\right)$

so that

$$\frac{d\epsilon}{dt} = \lambda pkT\int_{C}^{O} \exp\ (-\ \alpha t)\ d\alpha = \lambda pkT\left[\frac{e^{-\alpha t}}{-t}\right]_{\alpha=C}^{\alpha=O}$$
$$= \lambda pkT\left\{\frac{1-e^{-Ct}}{t}\right\} .. \quad .. \quad ..\ (V.37)$$

Smith points out that it is not strictly correct for the integration in equation *(V.36)* to be performed between the limits : E = O and E = ∞ because small volumes with activation energies greater than some maximum value cannot be present. He argues, however, that the small volumes with the higher activation energies make negligible contribution to the creep rate and the error introduced by taking an upper limit $E = \infty$ must be very small.

According to equation *(V.37)*, creep rate is proportional to the absolute temperature T and to the function $F(t) = \left\{ \frac{1 - e^{-ct}}{t} \right\}$

The function F(t) describes the form of the transient creep curve at a constant temperature. Mott and Nabarro (see page 126) have shown that for a precipitation-hardened metal the value of C is of the order $10^8 - 10^9$ sec.$^{-1}$ and in metals of very high purity, Smith considers that the value of C would not be below $10^4 - 10^5$ sec.$^{-1}$, thus in the region where $Ct \gg 1$ the form of the function F(t) is not of practical interest. In the region where $Ct \gg 1$, F(t) approaches the value $\frac{1}{t}$. By plotting curves of $y = (1 - e^{-ct})/_{ct}$ and $y = {}^1/_{ct}$, Smith shows that there is very little deviation between $F(t) = \left\{ \frac{1 - e^{-ct}}{t} \right\}$ and $F(t) = \frac{1}{t}$ for values of $Ct > 3$, so by using the approximation $F(t) = \frac{1}{t}$ in equation *(V.37)* negligible error is introduced at values of t greater than, approximately, 10^{-4} sec.

At times greater than 10^{-4} sec. after the onset of creep the equation for creep rate is, therefore, given accurately by:

$$\frac{d\epsilon}{dt} = \frac{\lambda pkT}{t} \quad \ldots \quad \ldots \quad \ldots (V.38)$$

and, by integration, the creep strain-time relationship becomes

$$\epsilon = (\lambda pkT) \log (\text{const. } t) \quad \ldots \quad \ldots (V.39)$$

i.e., at constant temperature

$$\epsilon = (\text{const.}) \log (\text{const. } t) \quad \ldots \quad \ldots (V.40)$$

Smith has thus obtained a logarithmic relationship between transient creep strain and time. He has shown that, by suitable choice of constants, good agreement is obtained between the theoretical curves given by equation *(V.40)* and curves obtained by experiment. He states that the agreement is quite remarkable in view of the fact that f(E, O) was chosen arbitrarily. He considers, however, that before a more complete understanding of transient creep can be achieved, it is necessary to know the form of the function f(E, O) more accurately, and also the manner in which it changes as plastic deformation progresses; this involves an investigation of the mechanism of the formation of energy barriers and of the laws which govern their distribution.

MOTT'S THEORY OF TRANSIENT CREEP(195)

Mott is investigating the theory of creep from the standpoint, expressed by Seitz and Read (see page 106), that the motion of dislocations is the only effective mechanism known by which crystalline grains of a material can be deformed, and he has set out to examine the types of dislocation which can exist, the mechanism by which they are formed, the factors which influence their motion and the relationship between the temperature dependence of their motion and the resulting deformation of single crystals or grains of a polycrystalline material.

At present, only a small proportion of the work which is being carried out at Bristol University, under the direction of Professor Mott, has been published. The theory of the mechanism of transient creep, in materials hardened before test and subjected to small stresses at relatively low temperatures, has formed the subject of a recent paper(195); but the extension of this theory, to cover the region of larger stresses and higher temperatures, is in the process of investigation(196), so, also, is the theory of grain boundary creep.

In presenting the theory of transient creep, Mott and Nabarro(195) begin by summarizing the properties of dislocations. They show that the most general type of dislocation joining two points P and Q (Figure 28) in a glide plane of a crystal, is curved in an arbitrary manner. " Curly " dislocations of this kind are of a composite nature and consist of a series of Edge type (Taylor) dislocations and Screw type dislocations(195). Such composite dislocations have higher energy than have straight ones, and this energy is regarded as appearing in the form of a tension along the length of a dislocation.

Since the motion of dislocations is regarded as being responsible for plastic deformation in crystals, it follows that resistance to deformation must be due either to the difficulty of forming dislocations, or to the difficulty of moving them.

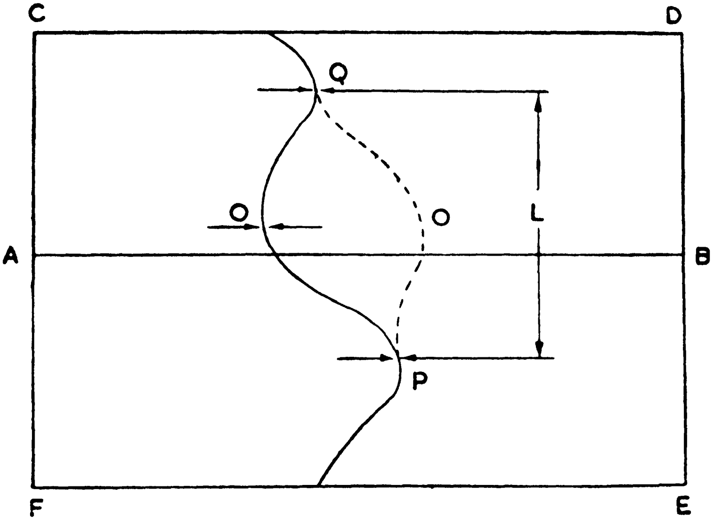


Figure 28.
A "Curly" dislocation crossing a glide plane. (The glide plane is in the plane of the diagram.) (Mott, ref. 195.)

The problems involved in the theory connected with the formation of dislocations are, as yet, unsolved. Frank(197) has suggested a possible mechanism whereby a dislocation, when set in motion, generates others, thus forming a slip band. Evidence, provided from the experimental fact that, in general, cold worked crystals are harder than annealed ones (according to the Dislocation Theory, the internal strains in cold worked crystals should assist in the formation of dislocations but should not hinder their motion) seems to indicate, however, that the difficulty of forming dislocations is not usually the important factor in producing resistance to deformation. Thus it appears that an attempt should first be made to explain resistance of deformation in terms of the stress required to set in motion the dislocations which are assumed to be already present in a real crystal.

Taylor(180) introduced the idea that the yield point of a crystal is determined by internal strains. Figure 29a is drawn to represent a section through a crystal containing internal strains; the line AB lies in a glide plane which is imagined to be at right angles to the plane of the figure; the small parallelograms are drawn to represent small parallelepipeds which, in a crystal containing no internal strains, would be cubes. In addition, the crystal will also contain some dislocations, and these will normally take up equilibrium positions in regions of low potential energy, such as at the point R. According to Taylor, slip cannot take place over the glide plane until an external strain is applied to the crystal which is sufficiently large to bring all the internal strains, in the direction AB, to the same sign. This condition is represented in Figure 29b. When this is the case a Taylor dislocation can move right across the glide plane in the direction AB. It follows, therefore, that if S is the stress at which yield along AB occurs, and S_i is the maximum internal shear stress at any point on AB, then $S=S$.

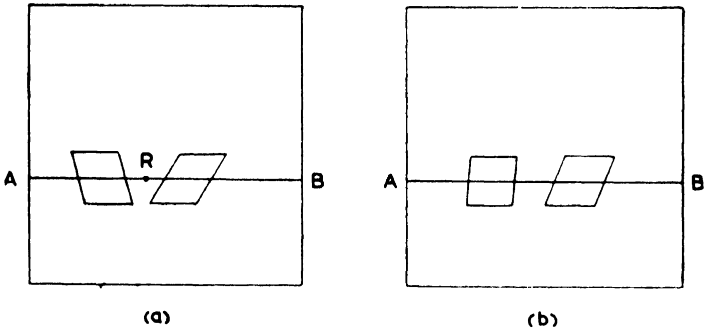


Figure 29.
Diagram representing internal strains in a crystal. (Mott, ref. 195.)

If Figure 28 represents the glide plane containing the line AB then, in the absence of externally applied strain, the component of shear strain parallel to CD will, in general, vary in magnitude and sign from point to point over the glide plane. According to Mott and Nabarro, a dislocation crossing the glide plane, between the points P and Q, will tend to follow a path of least potential energy, and this it will do provided the distance L (which is regarded as analogous to a wave-length) is sufficiently large to allow the internal stress to force it into a position of least potential against the reaction of its tension. If this is not so, the path of the dislocation will be constrained by its tension. Mott and Nabarro show that a dislocation will follow a path of least potential provided that its radius of curvature is the same order of magnitude as the wave-length L. A dislocation can be regarded, therefore, as being flexible, and its motion forward across the glide plane in the direction AB can be visualized as taking place by the straightening out of some loops and the formation of others in a sideways, snake-like fashion.

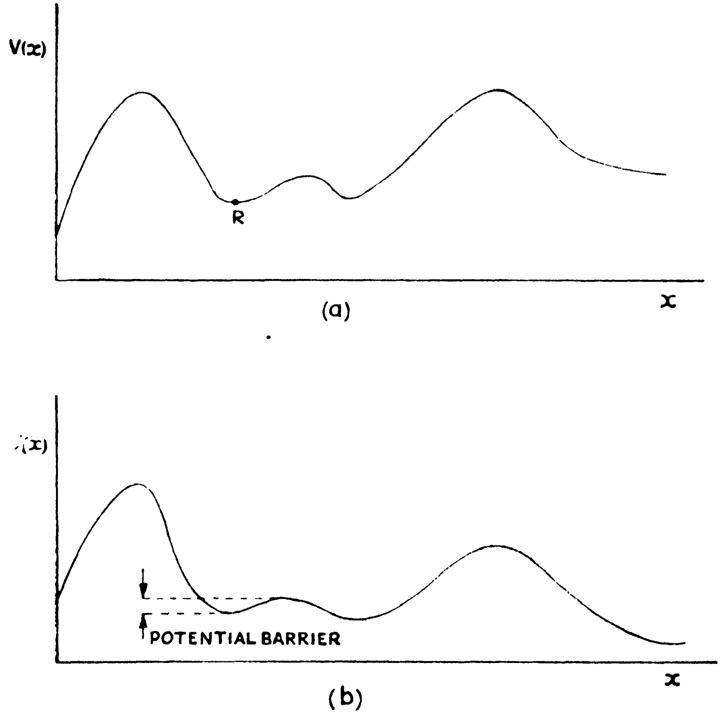


Figure 30.
Potential energy of a dislocation as a function of position.
(Mott, ref. 195.)

In the absence of an externally applied stress, any point along the length of a dislocation is assumed to be prevented from moving from its equilibrium position by potential energy barriers. Thus, along the line AB, in Figure 29, the distribution of potential energy, at any point distant x from A, is given by some oscillating function, V(x), which is represented, diagrammatically, in Figure 30. The point of intersection, R, of a dislocation with the line AB would lie, in its equilibrium position, somewhere in one of the troughs of the wave-like function V(x) as shown in Figure 30a; the exact position in the trough depends upon the constraint imposed by the tension in the dislocation (i.e., upon the value of L). In order to move the dislocation forward, its potential energy must be raised to the height of the barrier which prevents its motion, and Mott and Nabarro show that this cannot be done by the action of thermal stress fluctuations alone. When external stress is applied, however, it is assumed to have the effect of raising the potential of the minima of the function V(x) and of tilting the function, in the direction of the applied stress, in a manner shown in Figure 30b; thermal stress fluctuations are now capable of supplying the activation energy required for the dislocation to surmount a barrier such as that shown in Figure 30b, and it will " jump " forward into the next region of low potential.

Internal stresses in crystals are produced by (a) stress fields around dislocations(180), and (b) the presence of dissolved atoms, precipitates and Preston-Guinier Zones(198), (199), (200) within the crystal lattice. According to Taylor's theory of strain hardening, dislocations do not travel right across a slip plane during the straining of a crystal but they become stuck by the interlocking of one with another(201); thus, with increasing strain, the number of dislocations increases and, as a consequence, the potential function V(x) must change. Mott and Nabarro point out that the creep of soft single crystals, or of pure annealed materials, must, therefore, be interpreted in terms of the motion of dislocations in a stress field of potential V(x), which not only changes during straining but which is probably changed by the applied stress. They emphasize the fact, therefore, that an important first step in the study of creep should be the investigation of materials in which this is not the case, and in which the internal stresses, giving rise to V(x), change as little as possible during the test. They think of the strains around precipitates, which are assumed responsible for age-hardening, and suggest that an ideal test would be one carried out on a material in which ageing had taken place at a temperature higher than that of the creep test. They contend that tests on age-hardened materials, or solid solutions, may be simpler to interpret, and more likely to lead to advancement of the theory of creep, than tests on pure single crystals. Their present treatment of the mechanism of transient creep is concerned, therefore, with such materials.

A dislocation line is pictured as threading its way between particles of precipitate which are at distance L apart (i.e., one wave-length in Figure 28) and the case is considered in which, for the wave-length L, the internal stress, S_i, is sufficiently large to force the dislocation, against the reaction of its tension, into regions of least potential energy right across the glide plane.

The initial act which is envisaged as initiating creep is the motion forward of a loop of a dislocation from a position such as POQ (Figure 28) to a position PO′Q. The stress required to do this will be of the same order of magnitude as the maximum internal stress, S_i, in the region of the loop (cf. page 119, where it is pointed out that, according to Taylor's theory, slip cannot take place unless the applied stress is equal to the maximum internal stress).

Suppose an external stress S is applied to a crystal containing dislocations. All dislocation loops in regions where $S < S$ will move, without the aid of thermal fluctuations, giving a certain instantaneous deformation ϵ_0. Consider, now, a dislocation loop, such as POQ, where, it is assumed, the applied stress S is not quite large enough to move it. If it is to be able to jump to the position PO′Q, thermal stress fluctuations must supply an activation energy, E, which is dependent upon the value of the internal stress, S_i, and can be expressed by some function $W(S_i)$. The chance, per unit time, that, in unit volume of a crystal, a dislocation loop will surmount an energy barrier will also be a function, $\alpha(S_i)$, of the internal stress and is given by (cf. page 110)

$$\alpha\,(S_i) = C \exp\left\{-\frac{W(S_i)}{kT}\right\} \quad .. \quad .. \quad (V.41)$$

Mott and Nabarro show that the function $W(S_i)$ should have the form

$$W(S_i) = \text{const. } \lambda S_i L^2 \left(1 - \frac{S}{S_i}\right)^{3/2} \quad .. \qquad .. \ (V.42)$$

where λ is the macroscopic deformation produced by the passage of a Taylor dislocation across a glide plane, $S_i\lambda$ is the force acting on unit length of a dislocation in a material where the stress is S_i, and L is the length of a dislocation having two positions of equilibrium distant L apart.

Along each dislocation, and for the various dislocations in the crystal, the value of the internal stress, S_i, will vary, from loop to loop, over a wide range. The distribution of internal stress at time t = O (i.e., before the application of external stress to the crystal) is given by some function $N(S_i)$ which is assumed to be related to the internal stress by a curve of form similar to that shown in Figure 31. At time t = O, therefore, the number, n,

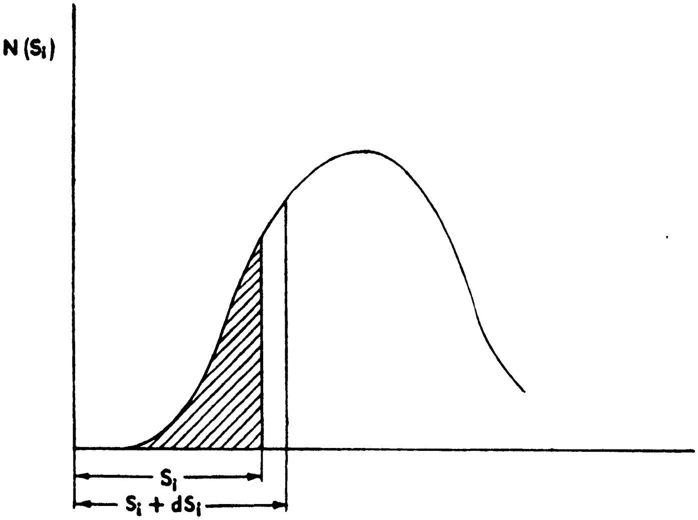


Figure 31.

Curve showing the distribution of the number of dislocation loops, per unit volume, for the movement of which a stress S_i is required at absolute zero of temperature. (Mott, ref. 195.)

of dislocation loops for which stresses between S_i and $(S_i + dS_i)$ are necessary in order to move them is given by

$$n = N(S_i)dS_i \quad .. \quad .. \quad .. (V.43)$$

this may be seen by reference to Figure 31.

When the external stress, S, is applied to the crystal, all loops for which $S_i < S$ will move, that is, all those within the shaded area in Figure 31. It is, now, assumed that the application of the stress and the subsequent plastic deformation by creep do not change the function $N(S_i)$; this will be the case when small stresses are applied to specimens which have been hardened, for example, by age-hardening, and tested at temperatures below that at which the age-hardening took place. At the end of the initial, instantaneous, deformation ϵ_0, the value of S_i around all dislocation loops is greater than the value of S, and the further plastic deformation (creep) which occurs does so by the aid of thermal stress fluctuations.

The number, n, of dislocation loops (for the movement of which stresses between S_i and $(S_i + dS_i)$ are required) present at time, t, after the onset of creep, is a function, n(t), of t; so that the change in the number of these loops during the time interval t = O and t = t, (t = O is taken, here, as marking the onset of creep) is given by

$$n(o) - n(t) = \int_o^t \alpha(S_i)n(t)dt \quad .. \quad .. (V.44)$$

or, by differentiation,

$$-\frac{d\{n(t)\}}{n(t)} = \alpha(S_i)dt \quad .. \quad .. (V.45)$$

Therefore

$$-\int_o^t \frac{d\{n(t)\}}{n(t)} = \int_o^t \alpha(S_i)dt \quad .. \quad .. (V.46)$$

that is

$$n(t) = n(o) \exp\{-\alpha(S_i)t\} \quad .. \quad .. (V.47)$$

But the number of dislocation loops of the type S_i, $(S_i + dS_i)$ at time $t = O$ is given by $N(S_i)dS_i$. So that the number of loops of this type which have not yet moved at time t after the onset of creep is given by

$$n(t) = N(S_i)dS_i \exp\left\{- \alpha(S_i)t\right\} \quad .. \quad .. \quad (V.48)$$

In the time interval dt after time t, the chance that a dislocation loop will surmount an energy barrier is, from equation *(V.41)*,

$$\alpha(S_i)dt$$

Thus the number of dislocation loops which will move in this time interval is

$$N(S_i)dS_i \exp\left\{- \alpha(S_i)t\right\}\alpha(S_i)dt$$

If λ is the macroscopic deformation of a crystal associated with the movement forward of one loop of a dislocation, then (considering unit volume) the strain produced in the crystal in the time interval t to (t + dt) by the movement of loops, for which stresses between S_i and $(S_i + dS_i)$ are necessary, is

$$\lambda N(S_i).\alpha(S_i).\exp\left\{- \alpha(S_i)t\right\}dS_i.dt.$$

and the strain, $d\epsilon$, produced by the movement of all loops is given by*

$$d\epsilon = \lambda \int_0^\infty N(S_i).\alpha(S_i).\exp\left\{- \alpha(S_i)t\right\}dS_i dt \quad .. \quad (V.49)$$

The creep rate is, therefore,

$$\frac{d\epsilon}{dt} = \lambda \int_0^\infty N(S_i)\alpha(S_i).\exp\left\{- \alpha(S_i)t\right\}dS \quad .. \quad (V.50)$$

The evaluation of this integral(195) gives an equation for transient creep curves of the general form

$$\epsilon = \text{const. } T^{2/3} (\log Ct)^{2/3} \quad .. \quad .. \quad (V.51)$$

---

* In order to take account of the possibility that each motion of a loop sets off an 'avalanche' of other dislocations, forming a slip band, Mott and Nabarro multiply λ by a numerical factor, p, which may be great compared with unity.

Mott and Nabarro consider that the coefficient, C, in equations *(V.41)* and *(V.51)* represents the frequency of vibration of a dislocation in its potential trough and they calculate that, for a precipitation-hardened alloy, its order of magnitude is 10^9 sec.$^{-1}$. It follows from equation *(V.51)* that the deformation in the time internal $t = 0$ to $t = 1$ sec. will be larger than that during an interval, say, $t = 1$ to $t = 10^6$ secs. Creep taking place in the first second after the application of stress will certainly be measured as instantaneous deformation; it seems, therefore, that a characteristic of creep which takes place by the " exhaustion " of dislocations, as described in this work, is that the instantaneous extension will always be somewhat greater than the total time-dependent extension; a curve of this type is shown in Figure 32. It remains for experiment to demonstrate that the curve of Figure 32 is typical of the behaviour of age-hardened alloys under conditions of relatively low stress and temperature.

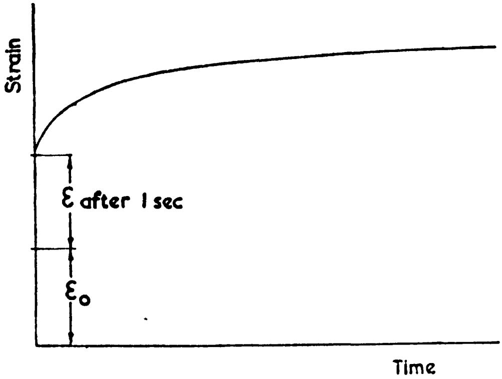


Figure 32.
A curve characterizing creep which takes place by the "exhaustion" of dislocations. (Mott, ref. 195.)

Conclusion

The foregoing survey has briefly summarized the present position in the development of a theory to account for the mechanism whereby the creep process occurs in metals and alloys. Advancement in theoretical studies has been slow, due, in large measure, to the lack of reliable experimental information and to the disregard which research workers have shown to the microscopic (submicroscopic would, perhaps, be more appropriate) behaviour of metals.

Kanter stated(68) that to express creep in terms of the various factors governing it would involve a function of staggering complexity. Such a function would be required, at least, to place the following factors in their correct inter-relation: stress, strain, strain rate, strain energy, temperature, time, strain-hardening, age-hardening, recovery, recrystallization, grain size, etc. It is little wonder, then, that Metallurgists and Engineers despair of receiving much immediate assistance from the Physicist in the solution of the problems which confront them in connection with the creep behaviour of service materials; they have good reason to adopt the attitude of reserve, towards the theoretical treatment of the subject, which was expressed by Allen(202) in these words:

> "It is simple to arrive at a formula which shall predict, with fair accuracy, an observed behaviour, but it is another matter to demonstrate that the processes assumed in the derivation of the formula are actually going on in the metal."

Author's Note

During the preparation of this book, from the author's manuscript, a paper, by A. H. Cottrell and B. A. Bilby, entitled "Dislocation Theory of Yielding and Strain Ageing of Iron," has been published in the Proceedings of the Physical Society (January, 1949, volume 62, page 49). This paper makes an important contribution to the application of the Dislocation Theory to the study of plastic deformation. Dr. Cottrell and Mr. Bilby show how the yielding and strain ageing of iron can be explained by the conception that carbon atoms form atmospheres round dislocations. The temperature dependence of the yield point of iron is explained on the assumption that thermal fluctuations enable small dislocation loops to break away from their atmospheres of carbon atoms, and strain ageing is interpreted in terms of the migration of carbon atoms to free dislocations.

APPENDIX I

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