SUMMARY OF D.Sc. THESIS AND ADDITIONAL PAPERS

by J. Glen, B.Sc., A.R.C.S.T., F.I.M.

A large number of creep tests on various steels were carried out to study the effect of composition, heat treatment, and testing variables on the creep resistance. Some of this work has been published [1,2,4]. It was found however that the order of creep resistance of various steels could be changed by altering the stress or temperature of testing. A similar effect can occur on comparing the same steel in the normalised and in the normalised and tempered condition. The former is always the better at high stresses and /or temperatures but the tempered steel may be the better at low stresses and temperatures. Very long time creep and rupture tests were carried out on 0.5% molybdenum steel, 1% chromium - 0.5% molybdenum steel and on molybdenum-vanadium steel. From these tests an estimate of the creep and rupture strength in 100,000 hours was obtained and the change in rupture ductility with increasing testing time was demonstrated.

The above tests indicated that the creep of commercial alloys was too complex a problem to be easily solved even by the accumulation of a large body of creep test data, and it was decided to approach the problem from a different angle. The effect of individual alloying elements was thus investigated by carrying out true stress-true strain tensile tests on various steels over a range of temperature [3,5]. It was found that, with/

with a simple iron-carbon alloy, a maximum in stress was obtained at about 200°C., and that this maximum was intensified if a little nitrogen (0.005%) was present. A minimum in reduction of area was obtained at about the same temperature. With the addition of either manganese chromium, molybdenum, tungsten or copper to the steel a second maximum in stress and a second minimum in reduction of area was obtained at a temperature higher than 200°C., the temperature depending on the element added. This second strain-age-hardening effect appears to be associated with some form of coherent precipitation of alloy carbide (or copper) in the dislocations formed during straining. It was thus concluded that the first strain-age-hardening effect was due to the precipitation of iron carbide and nitride. It was also found that a powerful carbide former tends to modify the effect of a less powerful one. This was attributed to the affinity of carbon and nitrogen atoms in solution for alloy atoms. Carbon and nitrogen atoms are fixed in interstices adjacent to alloy atoms and thus have less tendency to migrate into the dislocations during straining.

In the belief that similar effects must operate during creep testing a careful reappraisal was made of the shape of the many creep test curves which were available. It was found that unlike those of pure metals or simple solid solutions the creep curves of most alloys showed a rapid deceleration of creep rate over one or more intervals of time. These phenomena were called transitions in creep rate. In a simple iron-carbon alloy/

alloy only one transition in creep rate was obtained. If nitrogen was present two transitions were obtained. If manganese was added to the steel a third transition was obtained. The addition of chromium and molybdenum also resulted in transitions in creep rate. These results are convincing proof that strain-age-hardening phenomena do in fact operate during creep testing.

A simple theory was developed to explain the results obtained, log strain-log creep rate curves being used to present the data. The basic idea of the theory is as follows. Strain-hardening is the most important factor in the initial stage of a creep test, the movement and multiplication of dislocations being so rapid that strain-age-hardening has no time to occur. As the creep rate decreases, the rate of formation of dislocations decreases and the number stationary at any instant increases so that the strain-ageing constituent can precipitate in a few dislocations. When these are fixed other dislocations can pile up giving more time for further ageing. The process is thus self aggravating so that the creep rate slows down rapidly over an interval of time giving a transition in creep rate.

Apart from the steels above mentioned evidence is presented to show that alloys with a face-centred cubic structure (austenitic alloys and nimonic) and hexagonal close-packed alloys (titanium alloys) behave in quite a similar way.

From a family of strain-rate curves at constant temperature or constant/

constant stress it is possible to estimate the strain-rate curves for lower stresses or temperatures and by integration to obtain the ordinary creep curves. In this way extrapolation to long testing times can be made with some theoretical justification. Examples of such extrapolations are given.

Finally it is shown that strain-age-hardening can result in intercrystalline failure with a low extension. It was found that such failures occurred at or near to the beginning of a transition in creep rate. A tentative hypothesis is put forward to account for this phenomenon.

<u>THESIS</u>: "The Effect of Alloying Elements on Creep Behaviour with Particular Reference to Steel"

<u>PAPERS</u>: 1) "Abnormal Creep in Carbon Steels" (Journal I.S.I. April 1947, pp. 501-512).

2) "The Creep Properties of Molybdenum, Chrome-Molybdenum and Molybdenum-Vanadium Steels" (Journal I.S.I. January 1948, pp. 37-80).

3) "An Experimental Study of the Strength and Ductility of Steel at Elevated Temperatures" (A.S.T.M. Special Technical Publication 128, 1953, pp. 184-224).

4) "Some Additional Creep and Rupture Data" (Journal I.S.I. April 1955, pp. 320-336).

5) "The Effect of Alloying Elements on High Temperature Tensile Strength of Normalised Low-Carbon Steel" (Journal I.S.I. May 1957, pp. 21-48).

THE EFFECT OF ALLOYING ELEMENTS ON CREEP BEHAVIOUR WITH PARTICULAR REFERENCE TO STEEL

THE EFFECT OF ALLOYING ELEMENTS ON CREEP BEHAVIOUR

WITH PARTICULAR REFERENCE TO STEEL

By

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CONTENTS

ABSTRACT

Evidence is presented to show that, unlike pure metals or simple solid solutions, the creep resistance of alloys depends on some form of strain-ageing or strain induced precipitation. Because of such effects the shape of the creep curves alter, the creep rate slowing down rapidly over an interval of time. This phenomenon has been called a transition in creep rate.

A theory has been outlined making use of log strain - log creep rate curves to present the data. Detailed experimental evidence is presented to show that transitions in creep rate result from the presence of nitrogen, carbon, manganese, chromium and molybdenum in steel. The effect of vanadium, titanium and silicon is also considered. It is shown that similar effects occur in austenitic type alloys and titanium alloys.

From a family of strain-rate curves at constant stress and various temperatures it is possible to estimate the strain-rate curves for lower temperatures. Similarly the strain-rate curves for low stresses can be estimated from a series of strain-rate curves at constant temperature and various stresses. By integration the ordinary creep curves can be obtained. Examples are given showing how the method can be applied to obtain the creep and rupture strength in 100,000 hours or even longer times. An estimate of the accuracy of the extrapolation can be made.

From strain-rate curves to fracture it is shown that, when a steel fails in an intercrystalline manner and with a low elongation during a long time rupture test, failure was initiated at or near to the beginning of a transition in creep rate. At higher elongations failure may occur during a transition but may also result from some other mechanism. It/

It is shown that the addition of one alloying element may minimise the embrittling effect of another alloying element. A tentative hypothesis has been put forward to account for these results.

THE EFFECT OF ALLOYING ELEMENTS ON CREEP BEHAVIOUR WITH PARTICULAR REFERENCE TO STEEL

By J. Glen, B.Sc., A.R.C.S.T., F.I.M.

Ever increasing amounts of power are required for heating and lighting and to supply the needs of industry and transport. Most of this power is obtained by the conversion of heat energy to mechanical or electrical energy. Needless to say, therefore, metals and alloys capable of withstanding stress at high temperature are of vital importance in the construction of power plant. Since the efficiency of conversion increases with increasing temperature the demand for new alloys to resist ever increasing temperatures and stresses is never ending. The chemical and petroleum industries which are expanding rapidly are also important users of creep resistant alloys since many chemical reactions are endothermic and take place at high temperatures.

In 1910, Andrade[1] carried out the first scientific study of the creep of metals and demonstrated that the now well-known characteristic shape of a constant stress creep curve was the same for all the pure metals he studied. Since that time investigation of the problem of creep has been pursued with ever increasing vigour as the practical importance of the problem became more urgent. A brief survey of the literature suffices to indicate the immensity of the effort expended and the success which has been obtained in the development of new alloys to meet the needs of the engineer. On a purely theoretical basis much less progress has been achieved. This was perhaps only to be expected since until fairly recently no satisfactory theory existed to/

to explain the strength properties even at ordinary temperatures. The advent of the dislocation theory has, however, led to a renewed interest in the more theoretical aspects of the study of strength.

The complexities of the problem, with particular reference to hardening of metals have been emphasised by Cottrell[2] who states: "It (work hardening) was the first problem to be attempted by the dislocation theory of slip and may well be the last to be solved". It is not to be expected, therefore, that a quick solution will be found for the still more complex problem of the basic mechanism of creep, since this involves other phenomena such as recovery, recrystallisation, grain boundary slip etc., in addition to work hardening itself. The progress which has been made in the dislocation theory as applied to creep is well summarised in a recent symposium[3].

Because the basic mechanism is unknown empirical methods have to be used to study the creep properties of industrial alloys. These methods of necessity involve the choice of a simple criterion of creep resistance which in the light of experience is considered suitable for the comparison of different materials, the best alloy on this basis then being subjected to more prolonged testing to provide information on which to base working stresses. In the early days of creep testing much controversy arose because a different criterion was used by practically every investigator in the field, two well-known examples being the Hatfield Time yield and the Barr-Bardgett test. While giving due credit to the pioneer efforts of these workers, it must be stated that few, if any of these early tests are now considered of any importance.

Continual investigation led to the development of more sophisticated tests which at least have the advantage that they are usually of somewhat longer duration than the earlier tests. The author has shown, however,[4] that the order of creep resistance of several steels can be completely changed by altering the/

the stress or temperature of testing. Thus, even the modern short time tests are of doubtful value.

Because of the difficulties associated with testing, the effort which must be expended to develop new alloys increases rapidly as the complexity of the alloy increases. Even the alloys which have already been developed are not always used to full advantage because of uncertainty regarding their true strength under long time service conditions. This aspect of the subject has been reviewed by Bailey(5). His paper and the wide discussion which it provoked indicates that as yet no reliable method exists which will indicate the high temperature properties in, say, 100,000 hours from tests of much shorter duration. The situation is perhaps better illustrated by an example taken from work carried out on simple carbon steels by Tapsell & Ridley. In 1946(6) these authors reported the results of an extensive series of creep tests on three carbon steels at stresses of 5, 2 and 1 ton per square inch. These tests were of about 5000 hours duration. Fig. 1 shows the time required to give a creep strain of 0.1% and 0.3% at various temperatures and a stress of 5 tons per square inch for one of these steels (a superheater header steel of 0.24% C. and 0.63% Mn.). The hatched line shows how the curves were extrapolated to obtain estimated temperatures of 430°C. and 456°C. for 0.1% and 0.3% creep respectively in 100,000 hours. In 1953, the results of more prolonged tests on this steel were reported(7) including one of about 70,000 hours at 450°C. These results permitted a more accurate estimate of the temperature for 0.1% and 0.3% creep in 100,000 hours, these temperatures being 443°C. and 446°C. respectively. Thus, even for such a simple steel the original extrapolation underestimated the temperature for 0.1% creep whereas it overestimated that for 0.3% creep. What is perhaps more important is the fact that for the same sample of steel at the same stress the curve for 0.1% creep is concave upwards, whereas at a slightly greater creep strain the curve is concave downwards. This unpredictable change in/

in slope renders suspect all other extrapolated curves of similar types unless the tests are of such a duration that the trend of the curves is well established. In this connection it should be emphasised that the length of a curve in a log-time plot may be very misleading. Thus, if test results extend from 1 to 10,000 hours only a 25% extrapolatiom in the length of a log-time curve is required to reach 100,000 hours, whereas on a linear time scale the extrapolation is no less than 900%. Since time is all important in such phenomena as spheroidisation, which may cause a serious reduction in creep resistance, it is possible that quite an abrupt change may occur in the slope of an iso-strain temperature-log time curve in the time interval between 10,000 and 100,000 hours. An illustration of an abrupt change is shown in Fig. 2 where the elongation at fracture for 0.5% molybdenum steel has been plotted on a log elongation-log time basis. Robinson[8] in tests lasting 5000 hours at 482°C. plotted on this basis estimated by a straight line extrapolation that the elongation in 100,000 hours would be only 0.6%, whereas from the Author's results[4] of much longer time tests at higher temperatures, it seems obvious that at 482°C. the elongation in 100,000 hours will be nearer 10%.

A much more extensive review of the literature on creep could be given but would serve no useful purpose for the present paper. Sufficient evidence has been presented to indicate that from a practical point of view, the position is unsatisfactory, since an inordinate amount of work is required for each new advance in the subject.

For more than ten years the author has been making attempts to circumvent the above impasse and has carried out many hundreds of creep tests of short and long duration, many of which have not yet been published. It was realised, however, that a solution to the problem might be obtained more quickly by approaching the subject from a different angle. It was considered that if /

if more were known about the effect of individual alloying elements on the properties of steel there would be a greater chance of co-ordinating the huge mass of, in many cases, conflicting creep test data which has been compiled by so many laboratories. The principal test chosen for the study of alloying elements was the true stress true strain tensile test carried out over a series of temperatures. Some of these tests have now been reported[9 and 10] and, as a result, the author believes that he can now put forward a simple theory which helps to explain the behaviour of industrial alloys under tensile creep conditions. In view of the vital practical importance of this problem, it was considered essential to present the theory on its present form before reporting all the tests which have led up to it and without attempting to fill in obvious gaps in the knowledge of the effects of a number of variables.

In the above part of this paper, the main aspects of the problem have been discussed. In the following a brief summary of the essential evidence leading up to the theory is presented, an outline of the theory given, and lastly, detailed experimental proof is presented on a variety of steels and other alloys. It should be noted that much of the data refers to ferritic steels which it will be realised are quite complex in their behaviour. Specific examples of the behaviour of austenitic steel and non-ferrous metals are, however, included to indicate that the theory, with perhaps some modification in individual cases, is of quite general application.

BASIC EVIDENCE

Pure Metal:

It is obviously difficult, if not impossible, to attempt a solution of the creep behaviour of a complex alloy without at least having some idea of the basic creep law for pure metal. Many investigators have suggested equations for the creep curve of a pure metal derived from limited experimental work but all of these have failed to withstand more rigorous testing. In was left to Dorn and his associates[11] to put forward a correlation between strain and time which considerably clarifies the problem. From X-ray analysis and plastic properties they concluded that the same sub-microscopic structure is developed at the same value of $te^{\frac{-\Delta H}{RT}}$ following creep at the same stress.

Also $\epsilon = f\left(te^{\frac{-\Delta H}{RT}}\right)$ for the same stress.

where ϵ = creep strain

t = time

T = absolute temperature

ΔH = activation energy

At high temperatures of testing, i.e. where rapid crystal recovery can occur, the activation energy ΔH approximates to that for self diffusion of the particular metal being tested. At intermediate temperatures of testing the activation energy is also found to be constant[12] but in this case is considerably smaller than the activation energy of self diffusion.

The above relationship pre-supposes a thermally stable material. If grain growth occurs during the test as noted by Guard & Hibbard[13] or recrystallisation such as was found in tests on lead by Greenwood[14] then obviously the relationship no longer holds. However, in commercial alloys such effects seldom occur and so the above complications need not be further discussed.

In the appropriate range of temperatures, therefore, but with the above reservations, it is a simple matter from the result of a creep test on a pure /

pure metal at one temperature and stress to deduce the creep curve at the same stress for any other temperature. Using this method of analysis it is not necessary to know the actual equation of the creep curve which, in any case, might be different for each material and in each condition tested. An important conclusion from this work is evident in Fig. 3 which shows creep curves at temperatures T_2, T_3, T_4 derived from an arbitrary creep curve at temperature T_1. It will be noted that the end of primary creep and the onset of tertiary creep occurs at identical strain values no matter the temperature of testing.

Two other important observations were made by Dorn[11]. He confirmed McLean's conclusion[15] that the ratio of the fraction of the creep strain arising from grain boundary shearing to the total creep strain remains essentially constant in a constant stress test but that the ratio increases as the stress decreases. He also noted that the same values of grain boundary strain were obtained at the same values of total strain independent of test temperature proving that the same creep law applies to grain boundary shear as applies to the total creep. To quote Dorn - "This suggests that grain boundary shearing might be attributed to local crystallographic mechanisms of deformation in the vicinity of the grain boundary rather than to a process such as viscous shearing". The fact that the strain arising from the grain boundaries increases as the stress decreases suggests that the strain damage becomes concentrated more and more in the vicinity of the grain boundary as the stress is reduced. The other important observation by Dorn concerned the strain at rupture. He found that at constant stress the strain at rupture was the same no matter the testing temperature.

In most mathematical theories of creep the stress σ enters the proposed function as $\frac{\sigma}{T}$ whereas Dorn considers that stress alone should be used. Using/

Using his method of analysis stress must be taken into account by carryiing out tests at various stresses, i.e. the stress dependence must be obtained by experiment. The actual stress law for creep of pure metals has still to be resolved and it is still impossible to deduce the creep curve at one stress from a given creep curve at a different stress. This suggests that even in the case of a complex alloy there is a better chance of obtaining a correlation between tests at constant stress and varying temperature rather than constant temperature and varying stress. In the field of practical creep testing two schools of thought exist one of which favours constant stress tests and the other constant temperature tests. However, although the above evidence supports the constant stress approach to the problem an important reservation must be made. Since the activation energy for creep of a pure metal changes on moving from intermediate temperatures to higher temperatures where rapid crystal recovery can occur, it is to be expected that a similar phenomenon will be encountered in testing a complex alloy. As will be shown later this effect explains some of the anomalies of short time creep testing, and what is more important, can be shown to be one source of error in the extrapolation of iso-strain curves from higher temperatures and shorter times of testing to lower temperatures and very long times of testing.

Simple Solid Solutions:

The relative importance of solid solution hardening and precipitation hardening as regards creep resistance is still the subject of controversy. It can be stated, however, that the addition of an element which goes into true substitutional solid solution increases the strength properties compared with those of a pure metal and no doubt, therefore, increases the creep resistance. The principal effect of such an addition is to/

to produce lattice distortion, the degree of distortion depending on the difference in atomic radius between solute and solvent[16]. This distortion increases the rate of strain hardening and therefore increases the creep resistance.

Another important effect of solute atoms is to increase the recovery and recrystallisation temperatures. Thus, there is a temperature range from the lowest temperature at which pure metal recovers rapidly up to the temperature where a solid solution begins to recover within which a very marked difference in creep resistance is to be expected between the pure metal and the solid solution. Although there is a difference in creep resistance there seems no reason to suppose from the evidence available that the creep behaviour of a simple solid solution is otherwise different from that of a pure metal, and thus the method of analysis proposed by Dorn for a pure metal should also apply to a simple solid solution, as indeed he himself has suggested[17].

Solid Solution with Stable Precipitation:

It is, of course, well known that the presence of a precipitate in fine dispersion increases the strength properties at ordinary temperatures, and that even a coarse dispersion can have a considerable effect as has been clearly shown by Gensamer for steel[18].

At temperatures where creep can occur the precipitate may not be stable and due to continual solution and reprecipitation the particles become fewer but larger in size. Alternatively the rate of spheroidisation may be so small that the precipitate remains unaffected during the progress of the creep tests. Giedt and others[19] have investigated the creep properties of such a thermally stable alloy containing a/

a precipitate and have concluded that the activation energy for creep is the same as that of a pure metal. Here again, therefore, the method of analysis proposed by Dorn for pure metal can be applied.

In the above three cases there is evidence to believe that creep at constant stress can be correlated by a simple functional relationship. This might have been expected since in all these cases the only structural change in the material during creep is that resulting from strain hardening. This increases during primary creep and becomes constant during secondary creep. It is perhaps more correct to say that the removal and replenishment of dislocations become equal in the secondary creep stage[20].

Unfortunately in most if not all cases involving practical materials the creep curves do not obey a simple functional relationship of the above type. As evidence of this it suffices to state that such materials when tested at constant stress do not reach the end of primary creep or the beginning of tertiary creep at the same strain values over a range of testing temperatures.

It can be concluded, therefore, that changes in structures occur during the creep testing of most alloys which are more complex than that resulting from simple strain hardening. It is essential, therefore, to know most of the effects of alloying elements and to determine, if possible, which are of major importance. Rotherham & Tottle[21] have reviewed this subject and detailed many of the possible effects of alloying elements so that there is no need to detail them here. Four effects must, however, be mentioned since they are of fundamental importance as regards this paper. These are as follows:- /

1) Solution Hardening.

2) Formation of dislocation precipitates (or atmospheres)
   a) Before testing.
   b) During creep testing.

3) Precipitation and spheroidisation.
   a) Before testing.
   b) During creep testing.

4) Grain boundary changes.

As already indicated solution hardening can greatly increase the creep resistance of a pure metal and no doubt it also contributes to the creep resistance of complex alloys. In such cases, however, it is very difficult to separate solution hardening from other phenomena so that its relative importance is difficult to assess. As will be made clear, however, its effect in complex alloys is possibly not as great as is generally supposed.

Strain-age-hardening:

When a metal is deformed plastically, multiplications of dislocations takes place and the metal is hardened because of mutual interference between these dislocations. This simple form of hardening, which has been called strain hardening, is the only method of hardening pure metals. If steel is deformed at room temperature it is strain-hardened but after a considerable time at room temperature or a short time at a higher temperature, further hardening takes place. This has been termed strain-age-hardening or strain-ageing. By deforming plastically at temperatures above room temperature strain-age-hardening can take place during the process of straining.

The author has already shown[10] that strain-age-hardening at high temperatures is a complex phenomenon. By carrying out short time tensile tests at various temperatures on a simple iron-carbide alloy, it was found that a maximum in stress occurred at about 200°C. At/

At this temperature, therefore, strain-age-hardening was a maximum under the conditions of test used. At a lower temperature less than the maximum strain-ageing is obtained, whereas at a temperature higher than 200°C. over-ageing or recovery takes place. The author has suggested that the increased strength results from some form of coherent precipitation of carbide and nitride in the dislocations formed during straining rather than by the formation of 'atmospheres' as postulated by Cottrell[22].

With the addition of a carbide forming element such as manganese, chromium, etc., a second maximum in strength is obtained at a temperature higher than 200°C. In these cases the second strain-age-hardening effect appears to be associated with the precipitation of a particular alloy carbide in the dislocations during straining.

Strain-age-hardening is obviously taking place when a maximum is obtained in tensile strength as is the case for example with mild steel at about 200°C., but the absence of a true maximum in stress does not mean that strain-age-hardening is absent. A retardation in the normal decrease in strength with increasing temperature is a sufficient indication of this phenomenon. Fig. 4 shows typical examples of stress-temperature curves for various materials from which it can be concluded that strain-age-hardening at high temperature is quite a common occurrence even for austenitic steel and non-ferrous metals.

During creep testing a test piece is strain-hardened to a greater or less extent depending on the stress and temperature of testing. It is to be expected, therefore, that one or more strain-age-hardening effects may operate to increase the creep resistance while the test is in progress.

Two aspects of strain-age-hardening have been noted which appear to be of fundamental importance in relation to creep testing. From the tensile tests carried out in simple iron carbon alloys with alloy additions[10] /

it was found that a powerful carbide former modifies the effect of a less powerful one. For example, at 200°C. chromium carbide cannot precipitate since chromium atoms diffuse too slowly at that temperature. The addition of chromium, however, reduces the intensity of hardening at about 200°C. and shifts the temperature at which maximum ageing occurs at a somewhat higher temperature. It would appear that the carbon and nitrogen atoms in interstitial solution tend to segregate to interstices adjacent to chromium atoms, and thus have difficulty in migrating into the dislocations formed during straining. Such effects appear quite common and thus it is to be expected that the intensity of each strain-age-hardening effect which may operate during creep testing, will alter depending on the amount of each alloying element present. This phenomenon makes it difficult to assess the relative importance of solution hardening in such alloys since what is thought to be solution hardening may, in some cases, simply be the effect of one alloying element on the strain-age-hardening behaviour of another alloying element.

The second aspect of strain-ageing which is important in practice relates to the effect of tempering (or ageing). This behaviour is dealt with below when considering the effect of precipitation.

Precipitation and Spheroidisation:

Although an immense literature exists on the subject of precipitation hardening, comparatively little is known about its effect on high temperature mechanical properties. In the solution treated condition the "yield stress" of a precipitation hardening type alloy is low so that in a creep test even at a low stress the creep rate in the initial stages of testing will be high. In the precipitation hardened condition the "yield stress" is high so that the creep rate in the initial stages will be low. Even in the over-aged condition, this may still apply /

apply since the yield stress is usually higher than that of the solution treated condition.

In longer time creep tests, however, a simple relationship between yield stress and creep resistance, such as is indicated above, cannot apply, since structural changes occur during the progress of a creep test which may result in an increase or decrease in the creep rate. For example, in creep testing at a temperature near to the ageing temperature a precipitation hardened sample, although initially good will over-age rapidly and fail quickly. The solution treated test piece, however, will harden before it over-ages, so that failure will be delayed. Fortunately such effects usually occur at temperatures considerably in excess of those which are of practical importance.

As already indicated, strain-age-hardening manifests itself when excess dislocations are present. Strain induced precipitation can, however, also occur with precipitation hardening alloys. Indeed, as shown by the author[9, 10], there is little or no difference between the two types of alloy in high temperature tensile tests. This is illustrated in Fig. 5 for two molybdenum steels in the normalised and normalised and tempered condition. It will be noted that the 1.5% molybdenum steel is definitely of the precipitation hardening type, since the tempered test piece is harder at room temperature than the normalised one, but that in both conditions it shows a strain-age-hardening effect at about 500°C. similar to that of the lower molybdenum steel. It can be tentatively concluded, therefore, that during tests such as high temperature tensile or creep tests, precipitation hardening alloys will behave in a similar manner to strain-age-hardening alloys. As a result of this simplification the effect of tempering before creep testing is more easily discussed. The results obtained on/

on 0.5% molybdenum steel will be taken as typical. Examples for other steels have already been given by the author(4, 23). When a 0.5% molybdenum steel is normalised the equilibrium amount of alloy carbide has not time to form so that the dislocations present in the unstrained state will contain iron carbide and nitride. On testing at 550°C. (Fig. 6) the creep rate initially is high, since iron carbide and nitride are not stable. As testing proceeds, however, alloy carbide tends to replace iron carbide in the dislocations so that the creep resistance is maintained over a considerable period of time. If the steel is tempered at 650°C. some of the more stable molybdenum carbide is present in the dislocations before testing so that the creep rate initially is low. But since some of the molybdenum has precipitated in the dislocations and has also been transferred to the massive carbide phase during tempering, less remains to be precipitated during further straining. Thus, although the tempered test piece takes a longer time to reach 0.1% creep compared with the normalised steel, it takes a shorter time to reach a deformation of 0.5%. With prolonged tempering most of the molybdenum carbide is spheroidised and the creep resistance deteriorates. High carbon steels tend to deteriorate more rapidly on tempering. This is to be expected, since the more carbon present the greater the loss of alloying element from solution to the massive carbide phase during tempering. At the same time the massive carbide phase tends to spheroidise so that the strengthening effect of its presence decreases.

At lower temperatures precipitation may be so sluggish that the test piece deforms considerably before a particular constituent has time to precipitate. In such cases tempering before testing may be beneficial at low stresses and temperatures of testing. Obviously/

Obviously, however, no prediction of long time properties can be made from comparatively short time tests unless a complete knowledge of the strain-age-hardening behaviour of an alloy is known. Enough has been said, however, to indicate that structural changes, quite apart from simple strain hardening, are quite widespread in commercial alloys.

In many non-ferrous precipitation hardening alloys the structural changes on tempering (ageing) are more complex than for steel. The loss of alloying elements from solution may be high and grain boundary precipitation may occur with consequent weakening of adjacent areas. Recrystallisation of the matrix can also take place during the process of precipitation. As with steel, however, strain-age-hardening occurs in high temperature tensile tests indicating that precipitation in the dislocations can occur during straining. The fact that cold work accelerates the rate of ageing during tempering supports this suggestion. It will be assumed meantime, therefore, that the mechanism of creep in such alloys is essentially the same as for steel with the possibility of further complications arising from such phenomena as recrystallisation of the matrix during creep testing.

Grain Boundary Changes:

During creep testing at high temperatures, the metal as a whole deforms but at the same time, the creep rate is more rapid in the grain boundary regions since they are weaker than the body of the crystals. As a result stresses are set up in the crystal lattice adjacent to the boundary and more intense stress concentrations at triple points, i.e. where three grains meet. As the stresses build up, an increasing distortion is set up in the regions of the crystals adjacent to the boundaries so that eventually recovery takes place, by some process such as polygonisation. In this way, fast creep in /

in the grain boundary regions continues indefinitely. However, as shown by the work of McLean & Dorn, already mentioned, the fraction of the total creep strain arising from grain boundary creep is constant so that the functional relationship suggested by Dorn for pure metals, holds even for samples of different grain size.

The importance of grain boundary creep is evident when testing pure metals and simple alloys, since with a fine grain size the creep resistance decreases[24] as might be expected. In more complex alloys grain size per se is probably of lesser importance. Indeed for 80% nickel 20% chromium alloys Pfeil[25] reports no change in creep resistance despite a large change in grain size.

The grain boundaries also play a part in the mechanism of fracture. This aspect of the subject is dealt with in more detail later in this paper but it might be mentioned at this stage that strain-age-hardening appears to be important as regards intercrystalline fracture with low elongation in creep rupture tests. For each strain-age-hardening maximum in stress obtained in high temperature tensile tests due to precipitation of an alloy carbide there corresponds a minimum in ductility at a somewhat higher temperature[10]. Since the mode of failure in the high temperature tensile tests was intercrystalline, it seems reasonable to conclude that intercrystalline cracking with low ductility after prolonged creep testing is also a consequence of strain-age-hardening.

To sum up, therefore, at this stage it can be said that a simple functional relationship appears to fit the creep of pure metals or simple solid solutions with or without stable precipitates, because the only structural change that takes place during creep testing such metals is strain hardening. Since the same relationship does not hold for a more complex alloy, it is inferred that other sub-microscopic structural/

structural changes occur during the creep testing of such alloys. Evidence is also presented to suggest that whether the alloy is of a simple strain-age-hardening or precipitation hardening type the structural change responsible is the precipitation of a second phase in the dislocations. The precipitation of this second phase may also lead to intercrystalline failure with low elongation after prolonged creep testing.

SHAPE OF CREEP CURVES

It has been concluded that the creep resistance and mode of failure of most commercial alloys largely depends on sub-microscopic structural changes which take place while a creep test is in progress. If this conclusion is correct it is to be expected that as the structural change takes place, some reflection of it will be noted in the shape of the creep curves obtained.

In a pure metal or a simple alloy where strain hardening alone predominates, the shape of the creep curve is as shown in Fig. 3. The creep rate at zero time is infinitely fast but slows down with time at first rapidly and then more slowly until after a certain time the creep rate is virtually constant. At a later stage the creep rate increases at first slowly and then more rapidly until fracture occurs. Various theories have been put forward to account for this shape but need not be detailed here. Sully[26] has recently made a survey of this aspect of the problem.

The effect of strain-ageing during creep testing has received very little attention. Dumbleton[27] carried out creep tests on single crystals of zinc containing nitrogen and reported that after a certain time the creep rate decreased abruptly and later ceased altogether. Since the crystal, after testing, showed the yield phenomena, it was concluded that strain-ageing had occurred. Tapsell and others[28] have/

have reported tests at 12 tons/sq.inch and various temperatures on a cast 0.5% molybdenum steel in the annealed condition. The results at the lower temperatures are shown in Fig. 7. It will be noted that creep practically ceases at 250°C. after less than 10 hours. At 200°C. the creep rate is fairly slow after about 200 hours, but at 150°C. the creep rate is quite fast even after 1500 hours. This is somewhat surprising since after straining at room temperature only a few minutes ageing is required at 150°C. or 200°C. to cause maximum ageing as measured by a tensile test at room temperature.

It would appear, therefore, as if moving dislocations behave in a rather different manner to fixed dislocations. Cottrell[29] has discussed this in relation to Dumbleton's test on zinc and states that a dislocation may be moving too fast for strain-ageing to occur. If, however, a dislocation stops at an obstacle or at another dislocation, strain-ageing will occur if the time at rest is long enough. Once strain-ageing begins the creep rate should decrease abruptly and eventually creep ceases entirely. The tests at 150°C. 200°C. and 250°C. on 0.5% molybdenum steel in Fig. 7 do not entirely support the above hypothesis, since, although the creep rate decreases rapidly in the tests at 200°C. and 250°C. there is no abrupt change in creep rate and creep does not cease entirely even after 500 or more hours of testing.

The author has shown the importance of manganese as regards the creep resistance of simple carbon manganese steels[30] but it is not generally realised that manganese is also essential to ensure high creep resistance in 0.5% molybdenum steel. This is illustrated in Fig. 8 showing creep tests at a temperature of 550°C. and a stress of 6 tons per sq.inch on two 0.5% molybdenum steels, one containing 0.11% manganese and the other 0.6% manganese.

It will be noted that molybdenum obviously does not exert its/

its full effect quickly at 550°C., manganese being necessary to maintain the creep resistance until the molybdenum has time to come into play. It is to be expected, therefore, that the effect of molybdenum will take an even longer time to develop at a lower temperature. A test by Ridley[7] on 0.5% molybdenum steel lasting 70,000 hours at a stress of 6 tons per sq.inch and a temperature of 500°C. illustrates this point. As shown in Fig. 9 after about 10,000 hours of test the creep rate began to increase, but after 40,000 hours this trend was reversed. The creep rate then began to decrease and was still decreasing after 70,000 hours. A similar effect was noted by Kirkby[31] on an 18% chromium, 9% nickel, 1% niobium austenitic steel at a much higher temperature. This is also shown in Fig. 9.

The above phenomenon is noticeable because it occurred after the third stage of creep had begun. There is no reason to suppose, however, that a similar phenomenon cannot also take place during the primary or secondary stages of creep. In the latter case the creep rate should again begin to decrease after it has reached a more or less steady value. During the primary stage any change should consist of an acceleration in the rate of decrease of creep rate with time. It remains, therefore, to demonstrate experimentally the existence of such effects.

To prove the above contention a large body of creep test data on various steels was re-examined carefully. By plotting the individual strain readings for given times on a very large scale (but within the experimental accuracy of the extensometers used) and drawing a smooth curve through the plotted points, it was found that the shape of the creep curves differed from that of a pure metal. In most cases, however, the variations, although obvious on a large scale graph, lose /

lose their identity when the graphs are reduced to a size which could be used for reproduction. For this and other reasons a different method of presentation was used as is made clear later. For the moment, therefore, it suffices to show the effect diagrammatically and give one example where the result is sufficiently marked to be shown clearly.

When the load is applied to a creep test piece it can be assumed that some of the pre-existing dislocations can move easily since they are unfavourably sited. For this reason the creep rate initially is infinitely fast no matter how quickly it slows down thereafter. In the absence of strain-age-hardening the creep curve will have the shape ABC as shown in Fig. 10. In an alloy where one strain-age-hardening CONSTITUENT can precipitate, for example a pure iron-carbon alloy, the creep curve must follow the curve ABC for some time before ageing begins One simple reason for this, of course, is the fact that some increase in dislocation density by strain hardening must take place before strain-age-hardening is possible. At the point B precipitation of the strain-age-hardening constituent in the dislocations begins resulting in a transition in creep rate so that the creep curve follows the curve BDEF. In most commercial alloys more than one such effect can occur. In the diagram (Fig. 10) the effect of a second strain-age-hardening constituent is illustrated. In this case the creep test had entered the third stage of creep before strain-age-hardening was initiated. The complete curve thus has the shape ABDEGH.

As already mentioned the above phenomenon is easily detected when it occurs during the third stage of creep and two examples of this kind have already been given in Fig. 9. A test showing a transition occurring at an earlier stage of creep is illustrated in Fig. 11. In this /

this figure part of a creep curve on a 1.0% manganese steel at a stress of 5 tons per sq.inch and a temperature of 450°C. is shown the transition in creep rate beginning after about 550 hours of testing.

A number of steels containing about 0.1% carbon and 1.0% manganese were prepared in a small high frequency furnace. Half of these steels were heavily deoxidised with aluminium so that in the normalised condition much of the nitrogen was present as aluminium nitride and was not available to cause strain-age-hardening. These steels thus approximate to iron-carbon alloys. The other steels can be considered as iron-carbon-nitrogen alloys, the nitrogen being about 0.005%.

On testing at a stress of 8 tons per sq.inch and a temperature of 450°C. it was found that each of the iron-carbon alloys showed only one transition in creep rate, whereas the iron-carbon-nitrogen alloys showed two.

Similar steels containing manganese were also prepared and tested. In this case the iron-carbon-manganese alloys show two transitions and the iron-carbon-nitrogen-manganese alloys three transitions. (Fig. 11 shows the third transition in one such alloy). In the same way chromium and molybdenum add further transitions to the creep curves. There is thus very strong evidence to show that the number of transitions in a creep curve from a particular steel is equal to the number of constituents which cause strain-age-hardening in high temperature tensile tests. In can be concluded, therefore, that the postulated strain-age-hardening phenomena do operate during creep testing.

When plotting creep data, particularly with extrapolation in view, the logarithm of the time to reach a particular stage is often used as one axis of the plot. As already shown (Figs. 1 and 2) this can be misleading. From the evidence presented it is obvious that strain-age- /

strain-age-hardening can alter the trend of such curves. To accentuate any such changes it would be better to make more use of creep rate since this is a measure of the "strength" at any point on a creep curve. Creep rate, however, must be combined with strain so that in any attempt to extrapolate, the time of test for a given strain can be evaluated by integration. An approach to creep based on these ideas is outlined below.

OUTLINE OF THEORY

Pure Metals:

Dorn and his Co-workers postulate[11] that in a pure metal or a stable alloy where no change in the precipitate occurs during testing, the same submicroscopic structures are developed at the same values of $t e^{\frac{-\Delta H}{RT}}$ following creep at the same stress to any given value of strain.

where

t = time

ΔH = activation energy for creep

T = absolute temperature

It follows that

$$\dot{\epsilon}_1 e^{\frac{\Delta H}{RT_1}} = \dot{\epsilon}_2 e^{\frac{\Delta H}{RT_2}} \qquad \text{(strain constant)}$$

where $\dot{\epsilon}_1$ and $\dot{\epsilon}_2$ are the creep rates obtained at any given strain on testing at temperature T_1 and T_2 respectively. Taking logarithms and transposing

$$\Delta H = \text{Constant} \left(\log \dot{\epsilon}_1 - \log \dot{\epsilon}_2\right)\left(\frac{T_1 \times T_2}{T_1 - T_2}\right)$$

Thus from creep curves such as in Fig. 3, if the logarithm of creep rate is plotted against total plastic strain (or rather the logarithm of the strain since this is more convenient) then curves such as in Fig. 12 are obtained. As shown for any two temperatures the distance X (i.e. $\log \dot{\epsilon}_1 - \log \dot{\epsilon}_2$) is constant for all strains since X is proportioned to ΔH. The curve for any other temperature in the range where ΔH is constant is easily obtained by calculation from the above equation. By integration the ordinary creep-time curve is also obtained.

A pure metal is strain-hardened during creep testing but this statement is not sufficient for the present purpose. A reasonable assumption is that the strain-hardening (or rather the dislocation density) increases during primary creep and becomes constant during secondary creep when /

when there is a balance between the rate of removal and replenishment of dislocations. The ultimate dislocation density is small at lower stresses and increases with increase in the stress of testing. It is very large at high stresses provided rapid recovery or recrystallisation do not interfere. (At very low stresses exhaustion creep may occur but need not be discussed). Also it follows from the work of Dorn[11] that in constant stress creep tests on pure metal the dislocation density at any given strain is constant no matter the temperature of test.

The above statement simplifies the strain-hardening aspect of the creep of pure metals but does not explain the basic mechanism of the creep process. Although this is not important as regards this paper it is convenient to assume a "dislocation climb" mechanism. This implies that dislocations arrested at a barrier undertake a climb process by self diffusion until they reach a plane where they can move under the applied stress.

Metal with One Strain-Age-Hardening Constituent:

In an alloy where one strain-age-hardening precipitate can form during straining, both strain-hardening and strain-age-hardening must be considered. In the condition before testing the metal has a certain dislocation density. Some, if not all of these dislocations will contain a dislocation precipitate. (The Cottrell theory would state that each dislocation is surrounded by an atmosphere of solute atoms). Thus, the creep resistance in the initial stages of testing is better than that of a pure metal. Strain hardening, is, however, the most important factor in reducing the creep rate since the movement and multiplication of dislocations is so rapid that strain-age-hardening cannot take place. Initially, therefore, the strain-rate curves (curves AB in Figs. 13 and 14) are similar in shape to those of a pure metal (Fig. 12). As the creep rate decreases, however, the rate of formation of dislocations decreases and because of strain-hardening the number/

number stationary at any instant increases. This allows time for strain-age-hardening. The strain-age-hardening constituent will tend to precipitate at points of maximum lattice strain, i.e. points where dislocation climb is most liable to occur. By preventing this climb in a few places more dislocations can pile up and become stationary giving more time for further ageing. As a result the strain-age-hardening process is self accelerating in a manner similar to that suggested by Cottrell(29) for a nitrogen bearing zinc single crystal. Because of this the creep rate slows down rapidly and a transition in creep rate is obtained over the interval BD (Figs. 13 and 14). When this transition is complete the strain rate curve for low stress (Fig. 13) follows a new path which has the same shape as the curve for a pure metal.

In Fig. 13 (low testing stress) a dotted curve BC has been added to show the probable strain-rate curve ABC which would be obtained if strain-age-hardening could be prevented. At some strain X on this curve the creep rate becomes constant as also does the dislocation density. But because the stress is low this dislocation density is small and there is more than sufficient of the strain ageing constituent to provide precipitates for all the available dislocations. Thus, the creep rate becomes constant at a strain Y which is greater than X. At a much higher strain the creep rate again accelerates because of grain boundary creep(11), but more usually because in ordinary testing the stress increases as the test piece thins down. The complete strain-rate curve thus has the shape ABDYEF.

At higher stresses (Fig. 14) a similar argument applies to the initial stages of testing. Because of the higher stress the ultimate dislocatiom density is greater and the strain at which ageing begins is greater compared with the test at a low stress discussed above. /

Once strain-age-hardening begins there are many more points at which a precipitate can form. As a result, all the strain-age-hardening constituent is used up at a strain Z which is less than X. Thus, as testing continues beyond a strain Z more dislocations form so that a continual re-distribution of the strain-ageing constituent is necessary to give the maximum strength in the presence of an increasing number of dislocations. The creep rate therefore is a minimum at a strain Z and thereafter it increases and later levels off at some higher value as the dislocation density becomes constant. The complate strain rate curve has the shape ABDZEFG (Fig. 14). It is important to note that in this case, as indeed in most practical creep curves, there is no real secondary creep stage as is the case for pure metal. By definition, however, secondary creep in a pure metal begins when the dislocation density becomes constant. On this basis secondary creep (if it can be so called) begins at the point E in the curve shown in Fig. 14.

With increasing temperature of testing the probability of any particular dislocation climbing increases since the creep rate is much faster. Because of this the strain-age-hardening constituent has less tendency to precipitate. Thus, the transition in creep rate occurs at a greater strain value the higher the temperature of testing. For the same reason the effect of strain-ageing on the shape of the strain-rate curve becomes less pronounced with increasing temperature until above a certain temperature the transition in creep rate disappears.

A family of strain-rate curves at a constant low stress and various temperatures similar to the curve in Fig. 13 could be constructued but since this case is seldom encountered in practice discussion will be confined to the type of curves obtained at higher stress levels. /

A family of such curves is shown in Fig. 15. At the lowest temperature of testing T_1 a transition in creep rate occurs as before over the interval B_1 C_1. At the next higher temperature of testing T_2 strain-age-hardening is about to begin at the point B_2. But at this strain in the lower temperature test, strain-age-hardening is approaching completeness. Thus, the submicroscopic structure in the two tests at this level of strain is different even assuming that the dislocation density itself is the same. This means that the activation energy for creep must be greater for the lower temperature of test at this level of strain so that the strain-rate curves diverge as the strain increases. In a pure metal, as already shown (Fig. 12) the distance between the curves remains constant.

At all but the highest temperature of testing in Fig. 15 the creep rate reaches a minimum value and thereafter increases and later reaches a constant value as the dislocation density becomes constant. In pure metals the dislocation density at high strains is constant for all temperatures of testing and probably this is also true in the present case. For this reason the submicroscopic structures at all temperatures will tend to reach similarity at high strain levels. They can never become identical since there is no strain-age-hardening in the test at the highest temperature. The activation energy is a maximum for any particular test at the point of minimum creep rate. It then decreases as the strain increases and tends to approach a constant value for all temperatures of test as was the case for pure metals at all strains levels. For the reason given above, however, the activation energy at high strains never reaches equality. It is always somewhat greater the lower the temperature of test.

In the above example representing a metal with one strain-age- /

strain-age-hardening constituent such as carbon in iron, it is impossible to obtain a transition in creep rate if it has not occurred by the time the density of dislocations has become constant. (Temperature T_4 or higher in Fig. 15). If the strain-ageing constituent is more complex as is the case with "manganese carbide" or molybdenum carbide in iron[10], strain-age-hardening is dependent on the rate of diffusion of the alloying element as well as that of carbon. Even in the absence of strain-hardening, precipitation of alloy carbide by replacement of iron carbide may occur in such cases if sufficient time is given. With a high alloy content this gives rise to precipitation hardening as ordinarily defined. Thus, in such alloys a transition in creep rate is possible evem when the creep rate is accelerating.

<u>Metal with Two Strain-age-Hardening Constituents</u>:

The rate of strain-age-hardening is accelerated if the degree of strain-hardening is increases, i.e. if the testing stress is increased, since there are more dislocations in which a precipitate can form. Similarly in precipitation hardening alloys the rate of precipitation is increased by increasing the strain-hardening since this increases the rate of diffusion of the alloying element. In this case, therefore, the strain-age-hardening might be called strain induces precipitation. Since this form of strain-age-hardening is accelerated by strain-hardening it will occur quite quickly at low temperatures if the testing stress is high enough. For example strain-ageing due to precipitation of manganese carbide occurs during the half hour taken to break a short time tensile test at 300°C. since it gives a maximum in stress at about that temperature[10]. Under creep conditions, however, at a temperature higher than 300°C. strain-ageing due to /

to manganese carbide may take a long time to come into play, the time being longer the lower the stress of testing. For example, as shown in Fig. 11 it requires no less than 550 hours at 450°C. with a stress of 5 tons per sq.inch before strain-ageing due to manganese begins.

Strain-rate curves at constant stress and various temperatures for an alloy with two strain-age-hardening constituents are shown in Fig. 16, the second being of the complex type, i.e. like molybdenum carbide in steel. The first transition in creep rate has similar characteristics to the one shown in Fig. 15. At the lowest temperature of testing T_1 in Fig. 16 strain-ageing due to the first constituent is still proceeding at the point D_1. At this strain, however, strain-ageing due to the second constituent begins giving rise to a second transition in creep rate over the interval D_1E_1. As the strain increases strain-age-hardening due to the two constituents proceeds ~~proceeds~~ simultaneously. At some higher strain, however, strain-ageing due to the first constituent is complete and with further straining its effect is less effective since it has to redistribute itself over a larger number of dislocations. The creep rate, however, still decreases for some time due to strain-ageing of the second constituent but because of the decreasing effect of the first constituent it will reach a minimum value at a point F_1 before strain ageing due to the second constituent is complete. The creep rate then increases with further strain, but as before, will tend later to level off to a constant value.

At the next higher temperature of testing T_2 conditions are similar except that strain-ageing due to the first constituent has nearly reached completion when the second transition in creep rate begins. At the next temperature T_3 the creep rate is increasing when the second transition begins. Even so, the effect of the second strain-ageing/

strain-ageing constituent is strong enough to halt the acceleration of creep rate and again to decrease it for some time before it again begins to accelerate. At the highest temperature of testing T_4 the creep rate once it begins to accelerate continues to accelerate until it eventually reaches a constant value. The strain-age-hardening effect of the second constituent, however, is still detectable.

In a similar manner three or more strain-age-hardening constituents may operate during creep testing by adding additional transitions in creep rate and further complicating the shape of the strain-rate curves.

<u>Effect of Heat Treatment</u>:

The microstructure of an alloy can vary widely depending on the heat treatment given to the test material. Close control of the final heat treatment, however, is not a guarantee of identical microstructure, even in the case if two batches of material from the same cast. The 'heat treatment' effect of prior processing must also be considered, particularly in the case of complex alloys. Wood & Rait[32] for example showed that a high temperature treatment given to a 3% Cr-Mo-V-W steel before the normal oil quenching and tempering treatment, changed the resulting microstructure considerably.

Cold working or 'warm working' after heat treatment also influence the submicroscopic structure. Both these processes will alter the dislocation density but will also result in a varying degree of strain-age-hardening or strain-induced precipitation.

In general, however, it can be stated that no matter the original microstructural condition of an alloy the same number of transitions in creep rate will be obtained. The amount of each strain-age-hardening constituent available for precipitation during creep testing may, however, vary greatly so that the creep resistance also varies depending on the/

the original microstructure of the alloy. No attempt is made in the paper to investigate this aspect of the problem in detail but some observations on the effect of tempering are given later.

Effect of Varying Stress:

In the foregoing the emphasis has been placed on tests at constant stress and various temperatures. In general such tests are more suitable for extrapolation purposes than tests at constant temperature because in the former the strain-hardening at any given strain is reasonably constant. For applications involving small deformations the constant stress method of testing is particularly useful.

In tests at constant stress the strain-rate curves obtained are approximately parallel at low strain levels. In constant temperature tests this no longer holds since the initial plastic strain increases rapidly as the stress of testing is increased. Thus, the strain-hardening in the initial stages of testing increases rapidly with increased stress of testing, as also does the dislocation density. Since strain-hardening accelerates ageing the shape of the strain-rate curves will vary considerably with stress. In extreme cases at high sttess some strain-ageing constituents are attenuated so greatly due to the high density of dislocations as to contribute little or nothing to the strength even although they may have a big effect below some lower stress. For example in short time tensile tests at temperatures above 300°C. an aluminium killed steel is as strong as an ordinary steel of the same composition, even although at 200°C. it is inferior[10].

This means that at high temperature strain-ageing due to nitrogen in the ordinary steel is attenuated due to the high density of dislocations and has little effect. On the other hand under creep conditions where the dislocation density is low, the ordinary steel is better since the nitrogen has a powerful effect on the small number of dislocations present.

Although by testing at constant temperature and various stresses other factors such as variation in strain-hardening are introduced, the pattern of strain-rate curves obtained clearly shows these variations. In practice, therefore, both methods of testing have their own advantages depending on the alloy being tested and the information required.

Extrapolation:

If a series of creep tests at constant stress and various temperatures are carried out on a particular alloy and the strain-rate curves calculated it is possible from the pattern of curves so obtained to estimate the strain-rate curve for the same stress at several lower temperatures. The initial part of the curves may be checked by experiment. From these strain-rate curves it is then possible by graphical methods to calculate the ordinary strain-time curves. Finally by plotting iso-strain curves of temperature against log time as in Fig. 1, the temperature for any given strain, in say, 100,000 hours can be obtained.

Similarly from strain-rate curves of tests at constant temperature and various stresses estimates of the strain-rate curves for lower stresses can be made and similarly treated. Examples of such extrapolations are given later in the paper.

In the above a simple mechanism for the creep resistance of /

of complex alloys has been postulated. No doubt this mechanism will need modification as the theories of strain-hardening, strain-age-hardening and precipitation hardening are further developed. It should be appreciated also that in some cases other structural changes may interfere, particularly if the temperature of testing is high. These will include recovery, recrystallisation, discontinuous precipitation and ordinary time induced precipitation. In such cases the discrepancies will be obvious from the strain-rate curves obtained and can be further investigated. The difficulties of testing should also be realised. Since strain hardening induces strain-ageing a slight degree of non-axial loading will alter the results obtained, this effect being more important the smaller the permissible deformation. The time at temperature before testing may also be important in some cases. Ideally, heating to temperature and loading should be instantaneous, though this is obviously impossible. By holding at a testing temperature above the service temperature for any appreciable time the alloy is "tempered" and precipitation of some constituent may occur and alter the shape of the strain-rate curves. At a lower temperature the same time at temperature before testing may have little or no effect.

The above outline details the essential features of the proposed theory. This could have been elaborated but it was considered advisable to await more detailed experimental evidence before doing so.

Having given a theoretical explanation of the expected behaviour of commercial alloys, all that remains is to give a detailed experimental proof using log strain, log creep curves to present the data. This evidence is given in the remaining part of this paper.

EXPERIMENTAL EVIDENCE

A search of the literature reveals that only a limited number of steels and other alloys have been systematically creep tested either at constant stress or constant temperature. Even in such cases and where individual test results are reported, the data is often given in summary form so that an accurate estimate of creep rate at various strains cannot be obtained. For the most part, therefore, the author has made use of his own results. Fortunately it has been possible to obtain creep data from other sources to augment the results and to provide evidence of the behaviour of more complex alloys. Such results also provide independent confirmation of the observed phenomena.

It should be realised also that the theoretical implications were not fully understood at the time many of the tests were being carried out so that some tests were discontinued before all the ageing effects were complete. An endeavour has been made, however, to select data to illustrate the following main points:

1) The effect of alloying elements.
2) The influence of heat treatment.
3) The relationship between stress, temperature and time of testing.

Pure Metals:

Most of the creep data published by Dorn and his associates is reproduced in the form of curves relating strain to compensated time ($te^{\frac{-\Delta H}{RT}}$). Obviously such data must give curves of log strain versus log creep rate similar to those in Fig. 12, that is strain-rate curves showing no transitions in creep rate. In/

In very pure metals, however, grain growth or recrystallisation can take place readily during creep testing resulting in a distortion of the strain-rate curves. A discussion on this aspect of creep testing is, however, outwith the scope of this paper.

Low Manganese Steel:

A number of low manganese steels were tested at a stress of 8 tons per sq.inch and a temperature of 450°C. The results obtained on four of these steels can be taken as typical. The analyses and other details are as follows:

Steel	Type	Al. Lbs/Ton	C.%	Si.%	Mn.%	Ni.%	Cr.%	Mo.%	Cu.%
1	18 lb. H.F.	3	.09	.05	.12	.01	.01	.01	.01
2	"	½	.095	.06	.15	"	"	"	"
3	"	3	.08	.07	.14	.09	.03	.02	.15
4	"	½	.08	.06	.17	"	"	"	"

The ingots were forged to 1 inch square bar and normalised at 950°C.

The creep data was plotted on a large scale and a smooth curve drawn through the plotted points. From the smooth curves thus obtained the creep rate was measured at small intervals of strain. This data was then plotted in the form of strain-rate curves as shown in Fig. 17. As already mentioned the transitions in creep rate found on testing these steels are not obvious from visual examination of the ordinary creep curves. It will be noted, however, that the transitions are clearly depicted by the strain-rate method of plotting. It will also be noted that the aluminium killed steels 1 and 3 containing very little active nitrogen show only one transition in creep rate which can be attributed to strain-age-hardening by precipitation of iron carbide in the dislocations. The other two steels 2 and 4 show /

show two transitions, one resulting from carbon as above and the other due to precipitation of iron nitride. Since nitrogen is a more mobile atom than carbon the first transition is due to the presence of this element and the second due to carbon. Steels 1 and 2, which were made with very pure armco iron, are inferior in creep resistance to steels 3 and 4 respectively. Evidently residual elements have a powerful effect by solution hardening and by preventing rapid recovery. They may also, to some extent, intensify the strain-age-hardening characteristics.

Effect of Manganese:

The results of the tests at a stress of 8 tons per sq.inch and a temperature of 450°C. on a large series of manganese steels made from pure armco iron were available(30), a few of which have been selected as typical. Details of these steels are given below:-

Steel	Type	Al. Lbs/Ton	C.%	Si.%	Mn.%
1	H.F.	3	.09	.05	.12
2	H.F.	½	.09	.06	.15
5	H.F.	2	.11	.16	.40
6	H.F.	0	.13	.16	.35
7	H.F.	2	.10	.10	.59
8	H.F.	0	.11	.12	.56
9	H.F.	2	.10	.12	1.06
10	H.F.	0	.09	.09	1.04
11	H.F.	2	.10	.16	1.64
12	H.F.	0	.11	.11	1.58

These steels were forged to 1 inch square bar and normalised at 950°C.

Strain-rate curves for the aluminium killed steels are/

are shown in Fig. 18. Hatched lines have been added in some cases to indicate the probable trend in longer testing times. Manganese introduces a second transition in the curves but it also reduces the creep rate even in the initial stages of testing. This is partly due to solution hardening and prevention of recovery but since the rate of cooling on normalising is not fast enough to inhibit completely the formation of manganese carbide some of the improvement in the initial stages is probably due to the presence of coherent manganese carbide in the dislocations before testing.

The strain-rate curves for the steels containing active nitrogen are shown in Fig. 19 and these, of course, show three transitions in creep rate. Nitrogen greatly improves the creep resistance of these steels, as is shown by comparison with Fig. 18. The aluminium killed steel 7 was also tested after normalising at 1050°C. and also after very slow cooling from the same temperature. At 1050°C. aluminium nitride has dissolved and does not reform on normalising. All the nitrogen is in the active form so that as shown in Fig. 19 this steel compares favourably with a similar steel not deoxidised with aluminium. By slow cooling from 1050°C. aluminium nitride has time to reform so that the creep resistance is no better than that of the same steel normalised at 950°C., (Fig. 18) despite the fact that the high temperature treatment results in a much coarser structure.

The above data serve to show the importance of manganese and proves that such steels show transitions in the creep curves, but gives no indication of the effect of manganese at different stresses and temperatures. Some basic open hearth steels have, however, been tested more extensively. Tests on an acid open hearth carbon steel reported by Tapsell & Ridley[6, 7] have also been used, though the/

the estimates of creep rate in this case are of necessity less accurate. Details of these steels are given below:

Steel	Type	Form	Al. Lb/Ton	C.%	Si.%	S.%	P.%	Mn.%
13	B.O.H.	1" Dia.	$2\frac{1}{2}$	.11	.05	.027	.012	.45
14	"	"	1	.12	.06	.039	.015	.52
15	"	1" Plate	2	.13	.15	.028	.025	1.02
16	"	"	0	.13	.04	.018	.026	1.00
Ref.(7)	A.O.H.Header		0	.24	.10	.032	.040	.61

All the above steels were tested in the normalised condition.

Tests at a temperature of 450°C. and various stresses were carried out on steels 13 to 16, some of these tests being continued to fracture. The results on the 0.5% manganese aluminium killed steel 13 are shown in Fig. 20. At the highest stress the effect of manganese is not apparent. At 12 tons per sw.inch the effect of manganese is slight and occurs during the third stage of creep. At 10 tons per sq.inch it is more obvious and causes a reversal of the creep rate. It is also clear at a stress of 8 tons per sq.inch. The important point to note, however, is that the shape of the curves changes rapidly as the stress of testing is decreased because of the smaller density of dislocations.

Fig. 21 shows similar curves for the 1.0% manganese aluminium killed steel 15. In this case the change in shape is more pronounced as the stress is decreased. It is also clear that the 1.0% manganese steel has a greater creep and rupture strength than the 0.5% manganese steel. In particular it should be noted that during the third stage of creep, the creep rate accelerates more slowly with the higher manganese steel.

The results on the other two steels (14 and 16) are shown in Figs. 22 and 23. In these tests an additional transition due to nitrogen is introduced. With increasing stress, however, the effect of nitrogen is/

is less marked and at about 14 tons per sq.inch and above it seems to disappear. This agrees with the fact that the tensile strength of aluminium killed and ordinary steel is identical at this temperature[10]. This is attributed to the attenuation of the effect of nitrogen as the dislocation density increases.

In Figs. 20 - 23 a tentative extrapolation of the strain-rate curves at the lower stresses are shown by hatched lines. It is believed that these curves represent, if anything, a conservative estimate of the properties of these steels. Obviously, however, much longer time tests would be necessary before a precise estimate of the complete shape of these curves could be made. Further reference to this aspect is made at the end of this paper.

Two of the above steels have also been tested at a constant stress of 5 tons per sq.inch and various temperatures. The results on the aluminium killed steel 15 are shown in Fig. 24. Tests were carried out on this steel at three temperatures, 450°C. 475°C. and 500°C. Since this steel has rather poor creep properties a more extensive series of tests was not carried out. Assuming a constant activation energy for the initial parts of the curves and a slightly increasing activation energy at later stages, an estimate was made of the strain-rate curves for several lower temperatures. These are indicated by hatched curves. At a later date the results of a creep test at 380°C. on a very similar type of material were obtained[33] and the strain-rate curve was found to be in reasonably good agreement with that predicted.

A more detailed series of tests carried out on steel 16 is shown in Fig. 25. In this case tests were carried out at 25°C. intervals from 425°C. to 550°C. From this series it is relatively easy to predict the strain-rate curve for a temperature of 400°C. or lower. An important /

important point to note is that the symmetry of the curves changes markedly above about 500°C. This implies that, at this stress, rapid recovery begins in tests above about 500°C. Such tests are of little use in any attempt to predict the long time properties at temperatures where this form of recovery is absent.

Tapsell & Ridley[7] have reported extensive results of tests on a carbon steel header. From the published information it is impossible to calculate the creep rates with great accuracy, particularly at short testing times. However, even although the strain-rate curves lack this fine detail, they are still very informative as shown in Fig. 26. It seems clear that as in Figs. 24 and 25 the third stage of creep begins at lower strain values the lower the temperature of testing, and that the creep rate therefore accelerates rapidly before levelling off later to a nearly constant creep rate. The shape of the strain-rate curves in Fig. 26 are rather different from those in Fig. 25. As is shown later tempering of a carbon steel results in strain-rate curves similar to those in Fig. 26. It seems likely, therefore, that the superheater header steel was either given a stress relieving treatment akin to tempering, or alternatively, normalising a heavy section results in a microstructure similar to that obtained by normalising and tempering a smaller section of a similar steel.

Effect of Molybdenum:

Although the results of many rupture tests at various stresses and temperatures on 0.5% molybdenum steel were available[4], information of the elongation during the time to rupture was not obtained so that strain-rate curves could not be produced. Also, since Tapsell & Ridley[6, 7] had already reported extensive long time creep tests on 0.5% molybdenum steel the author considered it unnecessary to repeat this work, and thus had /

had confined his attention to checking cast to cast variation by means of tests at a few different stresses and temperature[4]. Attention will therefore be confined to one steel which is very similar in analyses and mode of manufacture to the one reported by Ridley. Details of these steels are given below.

Steel	Type	Al. Lbs/Ton	Form	C.%	Si.%	S.%	P.%	Mn.%	Mo.%
17	B.O.H.	1/4	1" Dia.	.09	.22	.032	.018	.48	.50
Ref.(7)	"	-	Steampipe	.09	.22	.034	.023	.52	.50

Both these steels were tested in the normalised condition.

The results of tests on steel 17 at a temperature of 550°C. and stresses of 9, 6 and 4 tons per sq.inch are shown in Fig. 27. In each strain-rate curve three transitions in creep rate were obtained due to the presence of nitrogen, carbon and manganese. A fourth transition seems to have been beginning in the test at 6 tons per sq.inch when the test was stopped. From the evidence considered below it is known that a fourth transition does occur at a higher strain level. For completeness, therefore, the approximate position of this transition is shown by hatched curves.

Tapsell & Ridley's tests[6, 7] were carried out at stresses of 6, 4, 2 and 1 tons per sq.inch. The reported creep data was used to calculate the strain-rate curves as shown in Fig. 28, but of course, the fine detail is missing. However, the results of two tests on steel 17 (Fig. 27) have been superimposed on the curves to show the initial transitions in creep rate.

The test at a stress of 6 tons per sq.inch and a temperature of 500°C. lasted 70,000 hours and clearly shows the transition in creep rate due to molybdenum beginning at about 0.3% strain (see also Fig. 9).

-/-

The remaining tests were not of sufficient duration to show the effect of molybdenum, but the probable trend has been indicated by hatched curves. At 4 tons/sq.inch the effect of molybdenum is not very clear from the test results but seems to occur at a lower strain value than for the corresponding tests at 6 tons per sq.inch. At 2 and 1 ton per sq.inch the effect of molybdenum is very noticeable since it occurs at still lower strain values well within the limit of 0.5% strain which was the maximum extension measured. The transitions due to manganese etc., will of course also retreat to lower strain values as the stress of testing is reduced.

There seems no doubt that at still lower temperatures of testing similarly shaped curves will be obtained. Hatched curves have therefore been inserted to indicate the probable trend of such curves.

Although only two samples of 0.5% molybdenum steel of almost identical composition and treatment have been considered, it seems justifiable to conclude that on testing any type of 0.5% molybdenum steel, creep curves showing transitions in creep rate would be obtained.

Effect of Chromium:

From the evidence of high temperature tensile tests[10] it was known that the effect of chromium on the strain-ageing behaviour of a steel was similar to that of manganese except that the effect of chromium was a maximum at a slightly higher temperature. It is to be expected, therefore, that an additional transition in creep rate due to chromium will be obtained on testing chromium steels and that this transition will begin at a somewhat greater strain than the transition due to manganese.

One test carried out many years ago on a 1% chromium steel with a very low manganese content seems to prove this point, but/

but the test was of doubtful accuracy. It has not yet been possible to repeat this work but fortunately test data on 0.5% molybdenum steels containing chromium were available which suffices for the present purpose. Details of one of these steels and a comparable steel without chromium are given below. The analyses of a steel tested by Robinson(34) is also included.

Steel	Type	Form	Al. Lb/Ton	C.%	Si.%	Mn.%	Cr.%	Mo.%	Ni.%
17	B.O.H.	1" Dia.	$\frac{1}{4}$	.09	.22	.48	-	.50	-
18	A.O.H.	"	$\frac{1}{4}$	.10	.14	.51	.82	.54	-
Ref.(34)		Bar	-	.31	.11	.54	.83	.54	2.05

Fig. 29 shows the results of tests on steel 18 at a temperature of 550°C. and stresses of 4, 6 and 9 tons per sq. inch. The results on steel 17 are included for comparison. In these tests a transition in creep rate begins at about 0.04% strain at 4 tons per sq.inch, 0.09% strain at 6 tons per sq.inch and 0.4% strain at 9 tons per sq.inch. The curves for steel 17 do not show this transition so that it is obviously related to the presence of chromium. The transition in creep rate due to molybdenum occurs at still higher strains as was shown in Fig. 28. This transition was just beginning in the test at 6 tons per sq.inch when the test was stopped. The approximate position of the molybdenum transition is shown by hatched curves.

Three samples of steel 18 were also tested after normalising at different temperatures. The results are shown in Fig. 30. As in Fig. 29 the transition in creep rate due to chromium begins at about 0.09% strain. It is also interesting to note that with a low normalising temperature the creep rate in the initial stage of testing is less than that of a test normalised at a higher temperature. At about 0.02% strain /

strain, however, the strain-rate curves cross so that the steel normalised at the highest temperature becomes the best.

Before concluding the discussion on the effect of chromium, mention must be made of four 100,000 hour creep tests reported by Robinson[34]. These tests at four different stresses and at a temperature of 450°C. were carried out in a nickel-chromium-molybdenum steel. By careful measurement and plotting of the points given on the original curves and drawing smooth curves through these points, it was possible to obtain the strain-rate curves shown in Fig. 31. As before, the fine detail at the beginning of the tests is missing but there seems little doubt that the final transition shown results from the presence of chromium. At this temperature the effect of molybdenum would only be apparent after still longer testing times. The similarity of behaviour of the four tests does show the remarkable accuracy achieved many years ago by measuring equipment which would now be considered as archaic.

Effect of Vanadium and Titanium:

A study of the literature suggests that the addition of vanadium or titanium to a simple carbon-manganese steel does not result in as good a creep resistance as the addition of molybdenum. Some support for this conclusion is given by the results of high temperature tensile tests[10] which show that, unlike molybdenum, vanadium or titanium in steel do not give rise to strain-age-hardening at least at temperatures up to 650°C. On the other hand the presence of these elements in solution modifies the strain-age-hardening behaviour at temperatures below 650°C. For example the author has suggested[10] that strain-ageing due to manganese is effective up to a higher /

higher temperature when vanadium or titanium is present in the steel.

As will be shown the effect of vanadium and titanium may vary depending on the heat treatment given to the steel. Two extremes of treatment were thus chosen for the tests reported in this paper. With one the steels were spheroidised and in the softest possible condition. The other treatment used resulted in a higher tensile strength as well as giving good creep resistance. Three vanadium steels, a molybdenum-vanadium steel and a molybdenum-vanadium-titanium steel were tested. The analyses and other details are given below together with those of two 0.5% molybdenum steels and a carbon-manganese steel which were used for comparison.

Steel	Type	Al. Lb/Ton	Treatment	C.%	Si.%	Mn.%	Mo.%	V.%	Ti.%
7	H.F.	2	N.950°C.	.10	.10	.59	-	-	-
17	B.O.H.	¼	A) N.950°C. B) N. + 5 Hrs. 690°C.	.09	.22	.48	.50	-	-
19	H.F.	2	N.950°C.	.11	.08	.53	.51	-	-
20	H.F.	0	N.950°C. + 5 Hrs. 690°C.	.12	.04	.60	-	.23	-
21	H.F.	0	N.950°C. + 5 Hrs. 690°C.	.115	.05	.54	-	.465	-
22	H.F.	0	N.950°C. + 5 Hrs. 690°C.	.135	.09	.48	-	1.14	-
23	B.O.H.	2	A) N.950°C. + 5 Hrs. 750°C. B) W.Q.1150°C. + 5 Hrs. 750°C.	.075	.08	.35	-	-	.27
24	E.F.	¼	N.1050°C. + 5 Hrs. 690°C.	.10	.12	.56	.53	.22	-
25	H.F.	2	N.1150°C. + 5 Hrs. 750°C.	.10	.31	.49	.52	.25	.28

All the above steels were in the form of 1" dia. bar.

The results of tests on the vanadium steels 20, 21 and 22 at a stress of 4 tons per sq.inch and a temperature of 600°C. are shown in Fig. 32. With the treatment given to these steels most of the nitrogen is present as vanadium nitride. The steels should thus behave like aluminium killed steel. As expected only two transitions in creep rate were obtained, as was also the case with the aluminium killed manganese steels (Fig. 18). The rate of diffusion of vanadium is so slow that an enormous time of testing would be necessary before a third transition due to vanadium could be expected.

The results of tests on the 0.5% molybdenum steels 17 and 19 are also shown in Fig. 32. Comparison with the aluminium killed 0.5% molybdenum steel shows that the vanadium steels have the better creep resistance. Steel 17, which has an extra transition due to nitrogen, is however better than the vanadium steels when tested in the normalised condition. After tempering for 5 hours at 690°C. the creep properties are greatly reduced. On the other hand, it was known from the results of other tests that tempering makes little or no difference to the properties of vanadium steels. Vanadium is thus a much more powerful alloying element as regards creep than is generally realised.

As is also shown in Fig. 32, the creep rate of the vanadium steels is about one thousand times slower than that of the aluminium killed carbon-manganese steel 7 tested under the same conditions. It might be said that this is simply the solution hardening effect of vanadium. What appears more likely is that because vanadium increases the recovery temperature and because of its affinity for carbon in solution, the strain-age-hardening due to carbon and manganese is more effective in reducing the creep rate. Such an explanation would also explain /

explain the results of high temperature tensile tests. It will be appreciated from the data already presented that manganese, chromium and molybdenum have a similar though less effective influence on the strain-age-hardening due to other constituents.

The results obtained on the titanium steel 23 are shown in Fig. 33. These tests were also carried out at a stress of 4 tons per sq.inch and a temperature of 600°C. Titanium has a very powerful affinity for nitrogen so that even if the steel had not been deoxidised with aluminium, the steel, when tested in the normalised and tempered condition, would have behaved like an aluminium killed steel. As is evident the conclusions reached from the results on the vanadium steels apply equally to the titanium steel in the above condition of heat treatment.

By heating the titanium steel 23 to 1150°C. titanium carbide and nitride go into solution. If the steel is water quenched from this temperature titanium carbide and nitride does not reform during cooling. By tempering at 750°C. some precipitation hardening takes place. In this condition a very high creep resistance is obtained, as is shown by the result in Fig. 33. It is interesting to note that in this test an additional transition in creep rate makes its appearance. This can be attributed to the presence of nitrogen. It would appear, therefore, that even after 5 hours tempering at 750°C. all the nitrogen in solution after water quenching is not removed in the form of aluminium or titanium nitride.

Fig. 34 shows the results of tests on the more complex steels at a stress of 4 tons per sq.inch and a temperature of 600°C. The strain-rate curve for the molybdenum-vanadium steel 24 has 4 transitions in creep rate, the same number as was obtained on testing the/

the 0.5% molybdenum steel 17 which steel, however, did not have as good a creep resistance. The fourth transition in steel 24 begins at about 0.2% strain compared with about 0.6% strain for the 0.5% molybdenum steel. In both cases, however, this transition seems due to the presence of molybdenum. Even after 10,000 hours of test only three transitions were detected in the strain-rate curve for the molybdenum-vanadium-titanium steel 25 at the above stress and temperature. A very pronounced fourth transition is indicated by a hatched curve. This extrapolation is probably correct since a test on the same steel at a stress of 6 tons per sq.inch and a temperature of 625°C. shows a pronounced fourth transition at about 0.16% strain. The result of this test is also shown in Fig. 34.

The above data indicates that vanadium and titanium, as alloying elements, are more effective in the presence of other alloying elements and when the treatment of the steel is such as to leave some nitrogen in solution. The strain-age-hardening effect of other elements seems to be intensified by the presence of vanadium or titanium. In the complex steels these elements have also a direct effect on the creep resistance, since with the treatments given a very finely dispersed precipitate is obtained which does not spheroidise readily. The presence of such a precipitate hinders the movement of dislocations so that the creep resistance at low strains is improved and this benefit is maintained throughout the test.

An interesting feature of the results on the above steels is the influence of nitrogen. Depending on the treatment given to the steels nitrogen may have no influence on the creep rate or if there is sufficient in solution the creep rate may be reduced by a/

a factor of 100 or more during the nitrogen transition. Evidence from high temperature tensile tests[10] suggests that a somewhat similar effect may occur in simpler steels, particularly those containing chromium. The total nitrogen content may vary from cast to cast and may also vary depending on the method of manufacture. For example electric furnace steel usually has a higher nitrogen content than open hearth steel. If closer control of the nitrogen content of steel could be maintained as well as the heat treatment, the variation in creep properties found between different casts of the same type of steel would be reduced considerably. The deliberate addition of nitrogen as an alloying element should also be seriously considered.

Effect of Silicon:

So far attention in this paper has been confined to alloying elements which form definite carbides in steel. Silicon is not a carbide forming element and its strengthening effect is usually considered to be due simply to solution hardening. High temperature tensile tests on a 1% silicon steel[10] did not appear to show any strain-age-hardening effect due to silicon although the strain-age-hardening at about 200°C. did seem to increase.

The addition of silicon to a steel may, however, drastically change the shape of the creep curves obtained. This is certainly true for carbon-manganese steel, 0.5% molybdenum steel and 1% chromium 0.5% molybdenum steel[23]. The effect of silicon on these steels is to reduce the creep rate markedly in the initial stages of testing and at a later stage to cause a rapid acceleration of creep. Such behaviour cannot be explained by solution hardening, so one or/

or other of the following reasons must be correct. Either silicon does influence the strain-age-hardening behaviour or it has some other, as yet unknown, effect which has not been included in the theory outlined in this paper.

To illustrate the effect of silicon attention will be confined to two 1% silicon steels one of which was deoxidised with aluminium. Two other low silicon steels made under the same conditions and using the same raw materials are included for comparison. The analyses and other details are shown in the Table below.

Steel	Type	Al. Lbs/Ton	C.%	Si.%	Mn.%
26	H.F.	0	.105	.09	.82
27	H.F.	0	.10	1.02	.74
28	H.F.	2	.11	.05	.72
29	H.F.	2	.12	.96	.75

The steels were in the form of 1" dia. bar and were normalised at 950°C.

Tests were carried out at a temperature of 450°C. using stresses of 6 and 8 tons per sq.inch. The results obtained for steels 26 and 27 are shown in Fig. 35. The low silicon steel gave the result expected and the shape of the curves was similar to those of the carbon-manganese steels in Fig. 19. The high silicon steel also shows three transitions in creep rate the third occurring during the third stage of creep, but the shape of the curves is very different from those of steel 26. The creep rate rapidly falls to a low value and then begins to accelerate after only 0.01% or 0.02% strain. It should be noted that at the point of minimum creep rate the high silicon steel is creeping more than one thousand times slower than the steel of low silicon content.

The results on the aluminium killed steels 28 and 29 show an entirely different picture. It will be noted from Fig. 36 that although the high silicon steel has a slightly better creep resistance the strain rate curves of the two steels have the same shape, which is similar to that of the aluminium killed carbon-manganese steels in Fig. 18. From the results in Fig. 36 it can be concluded that silicon has a small solution hardening effect. Comparison of Figs. 35 and 36, however, leads to the conclusion that silicon has a powerful influence on strain-age-hardening due to nitrogen. A possible explanation is that some silicon nitride exists as a fine dispersion before testing and that silicon nitride precipitates in the dislocations during straining instead of iron nitride. At a later stage, however, silicon nitride spheroidises and becomes ineffective so that at high strains the creep rate is fast and approximates to that of aluminium killed steel. If such is the case it should be possible to destroy the high initial creep resistance by a prior tempering treatment to precipitate and spheroidise the silicon nitride. Some support of the above suggestion is given by the work of Turkdogan & Ignatowicz[35] who found that silicon nitride (Si_3N_4) exists in high silicon steels.

It is obvious that much work remains to be done to obtain a full explanation of the effect of silicon on the creep properties of steel. The above information, however, suffices to prove that silicon does modify the strain-age-hardening behaviour of steel. The effect of this element is therefore not an exception to the theory which has been outlined.

Effect of Other Alloying Elements:

Some work has been done to show that tungsten influences the creep properties in a manner similar to molybdenum. This is to be expected in view of the similarity of behaviour in high temperature tensile tests[10]. Copper is also known to improve the creep properties of steel. However, the amount of data available on these and other less common alloying elements was not sufficient to justify inclusion in this paper. One other element, namely, boron, is worthy of mention. This element tends to combine with nitrogen but it may also act of itself as an interstitial element. There is some evidence in the literature that it may substantially improve the creep properties. The effect of this element is thus worthy of study to see if it results in an additional transition in creep rate.

AUSTENITIC ALLOYS

In the foregoing part of this paper the evidence presented in support of the theory is confined to results on ferritic alloys, that is alloys with a body centred cubic lattice structure. It is well known that such alloys are particularly subject to strain-age-hardening effects because they are detectable at room temperature. Austenitic alloys, on the other hand, which have a face centred cubic structure and a high solubility for carbon and nitrogen, are not considered prone to such effects. The evidence of high temperature tensile tests[10] does, however, suggest that strain-age-hardening at high temperatures is possible. Precipitation from the austenitic solution is quite common, particularly in the more complex alloys. Also since 'warm working' is so advantageous it is evident that strain induced precipitation can occur readily. It might be inferred /

inferred, therefore, that austenitic alloys will show transitions in creep rate during creep testing similar to those obtained in testing ferritic steel.

Very little systematic creep testing appears to have been carried out on simple austenitic alloys and any data available in the literature does not show creep curves from which creep rates can be measured. It was thus impossible to build up a logical chain of evidence similar to that given for the ferritic steels. Data was, however, available on the following alloys. (See Table on Page 55).

Duplicate tests on a simple austenitic steel in the solution treated condition (probably 1050°C. W.Q.) have been reported by Cross & Lowther[36]. These tests, both of more than 10,000 hours duration were carried out at a stress of 8345 lbs. per sq.inch (3.7 tons per sq.inch) and a temperature of 1200°F. (649°C.). Creep rate data was given in tabular form and the results have been plotted to give the strain-rate curves shown in Fig. 37. Since the initial deformation on loading was 0.06% the unknown part of the curves must take the form indicated approximately by the hatched curves. Although the creep rates of the two tests varies somewhat in the early stages, they later approximate quite closely. It will be noted that ~~the~~ two transitions in creep rate are obtained in each test. One of these is probably related to the precipitation of chromium carbide during straining. In view of the few remaining possibilities the other transition may be due to precipitation of iron carbide. If this is correct it should be the first transition which refers to this constituent.

Kirkby & Sykes[31, 37] have carried out creep tests on an 18/9/1 niobium steel. From the result of a 20,000 hour test at/

AUSTENITIC TYPE ALLOYS

Ref.	Alloy	Treatment	C.%	Si.%	Mn.%	Ni.%	Cr.%	Mo.%	W.%	Ti.%	Nb.%	Co.%	Cu.%	Al.%
36	18/8	Sol$^{n.}$ Treated	-	-	-	18	8	-	-	-	-	-	-	-
31,37	18/9/1*	1050°C. A.C.	.11	.50	.41	9.5	17.8	-	-	-	1.22	-	-	-
37	18/12/1*	1050°C. A.C.	.12	.78	1.41	12.1	18.5	-	-	-	1.33	-	-	-
31,37	326	1250°C. W.Q. + Aged	.25	-	3.0	17	17	2.5	-	-	1.8	7		
31,37	327	1250°C. W.Q. + Aged	.20	-	-	17	17	3.0	-	.8	-	7	3	-
39,40,41	G.18B	1320°C. A.C.	.4	1.0	.8	13	13	2.0	2.5	-	3.0	10	-	-
25	'80/20'*	1225°C. 3 hrs. W.Q.	-	-	-	(80)	(20)	-	-	2.74	-	-	-	.31
42	Nimonic 80A	1080°C.8 hrs. A.C. + 16 hrs.700°C.	.1 max.	1.0 max.	1.0 max.	(75)	20	-	-	2.3	-	2.0	-	1.0

* ACTUAL ANALYSIS - Others are typical compositions.

at a stress of 3 tons per sq.inch and a temperature of 650°C. the strain-rate curve shown in Fig. 38 was calculated. Data on a similar steel of higher nickel content was also obtained[37]. The results of tests at a stress of 3 tons per sq.inch and temperatures of 650°C. and 700°C. are also shown in Fig. 38. It is interesting to note that, whereas the 18/8 steel (Fig. 37) shows two transitions in creep rate, the niobium steels show three. By tempering a test piece even for 20,000 hours at 650°C. it is not to be expected that niobium carbide would precipitate[38]. On the other hand, since some strain hardening occurs during creep the possibility of strain induced precipitation of niobium carbide must be considered possible. However, such questions can hardly be finalised by ordinary technique. Electron microscope studies of creep specimens may, however, give a definite answer to many of these problems.

The niobium steels have a substantially better creep resistance than the 18/8 steel. It would appear, therefore, that niobium in austenitic steel has an effect similar to vanadium or titanium in ferritic steel.

Comparing the two tests on the 18/12/1 steel it is evident that apart from the initial stages, the two curves are reasonably parallel. From a family of such curves, therefore, it should be possible to predict the strain-rate curve for a lower temperature from the results at higher temperature in a manner similar to that shown in Figs. 24 or 25. These steels were tested in the solution treated condition so that the time held at temperature before loading the test may cause some precipitation. This may explain the fact that the creep rate initially is better in the 700°C. test than in the test on the same steel at 650°C. At 700°C. more precipitation is liable to occur before testing than at 650°C. with the result that the initial creep rate is/

is better at 700°C. than at 650°C. Many such inversions have been reported in the literature. It is probable that some of these result from the same cause.

Kirkby & Sykes(31,37) have also carried out long time creep tests on two complex austenitic alloys 326 and 327. Oliver & Harris(39,40,41) have tested an alloy now well known as G.18B, and Pfeil, Allen and Conway(25) have improved on the 80/20 nickel chromium type of alloy by the addition of titanium and aluminium. Detailed information on Nimonic 80A was also received(42). Strain-rate curves of individual tests on these alloys are shown in Fig. 39, from which it is obvious that one or more transitions in creep rate occur in each case.

An examination of the results on some of these alloys in more detail is interesting. Fig. 40 shows strain-rate curves on alloy 337 at 6 different stresses, the temperature of testing being 700°C. The curves for stresses of 8, 9 and 12 tons per sq.inch are reasonable symmetrical and show at least three transitions in creep rate. The curve for 14 tons per sq.inch is distorted compared with the others, but this may be due to recovery processes resulting from the higher stress. The curves for 10 and 11 tons per sq.inch appear to be displaced though transitions can still be detected. These results emphasise the great experimental difficulties associated with the testing of such complex alloys. The strain-rate method of presentation does, however, help to indicate which tests are out of line.

Three test results on G.18B were available at a stress of 6 tons per sq.inch and temperatures of 650°C. 750°C. and 850°C. One of these tests was of 65,746 hours duration. The strain-rate curves are shown in Fig. 41. Although there is a large temperature difference between the tests it seems clear that two transitions in creep rate are obtained in each case and that these transitions occur at/

at higher strain levels the higher the temperature of testing. This type of behaviour is similar to that obtained on carbon-manganese steels (Figs. 24 and 25). To show this aspect more clearly, additional curves have been inserted. The strain-rate curve for the test at 650°C. has also been extrapolated. These additions are shown by hatched lines. It should be appreciated that additional transitions in creep rate may occur at strains less than 0.1%. More precise information on the shape of the creep curves during the initial stages of testing would, however, be necessary to check this possibility.

All the above materials can be considered as steels or at least iron base alloys. Detailed information was also obtained on one face centred cubic type of alloy, namely, Nimonic 80A, which has a nickel base, iron only being present as an impurity. This alloy was solution treated for 8 hours at 1080°C. air cooled followed by an ageing treatment of 16 hours at 700°C. air cool. The tests had been carried out at a temperature of 700°C. using stresses of 16, 13, 10, 7 and 4 tons per sq.inch. The longest test (34,053 hours) was at 7 tons per sq.inch. The data was plotted on a large scale and from smooth curves drawn through the plotted points, creep rates were measured and the strain rate curves shown in Fig. 42 obtained. The probable trend of the curve for 4 tons per sq.inch in longer testing times is indicated by a hatched curve. In view of the difficulties of testing such complex material and the higher temperature used it is not surprising to find some discrepancies in the strain-rate curves, particularly at low strain levels. Nevertheless in each of the five tests, three transitions in creep rate were obtained. In all the tests, with/

with the exception of the highest stress, the first transition occurs at about 0.005% strain, the second at about 0.03% strain and the third at approximately 0.3% strain. It should be noted that the third transition becomes more pronounced the lower the stress of testing, as was also found for the simple ferritic steels. This third transition at the higher stress of testing is not very pronounced. It should be appreciated, however, that it begins very near to the onset of fracture. If failure had been delayed it is possible that this transition would be more noticeable.

Even with a stress of 16 tons per sq.inch no initial plastic extension was obtained on loading the test. In previous examples of varying stress tests (Fig. 20 - 23) the initial plastic extension (and the dislocation density) increased as the stress of testing was increased so that the strain-rate curve pattern was distorted. The 0.1% proof stress of Nimonic 80A is about 35 tons per sq.inch at 700°C. so that stresses greater than 16 tons per sq.inch would be necessary to give high initial deformations. Such curves would no doubt have shown curves similar in shape to those in Figs. 20 - 23.

The above tests suffice to show that transitions in creep rate are found on testing austenitic alloys, as was also the case for ferritic alloys. These transitions are clearly revealed by the strain-rate method of plotting. From a family of such curves it should be possible to predict the long time creep properties of austenitic alloys. The particular constituent responsible for each transition is unknown is most cases. Tests carried out on simple austenitic alloys with single alloy additions would be very informative.

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TITANIUM AND TITANIUM ALLOYS

The vast majority of commercial alloys can be classed under one of three main lattice structure types. Alloys of the two cubic types of structure have already been shown to give transitions in the creep curves which can be correlated with strain-ageing or strain induced precipitation. All that remains, therefore, is to show that alloys with a hexagonal close packed structure behave in a similar manner. One of the most important metals of this type at the present time is titanium, and attention, therefore, will be confined to the results of creep tests on titanium and titanium alloys. Summary data on 1000 hour tests have been published by Imperial Chemical Industries Ltd.[43]. Full details of these tests have been obtained[44]. The alloys tested were as follows:

Alloy	Type
199	Commercially pure but not highest purity.
314	Low Aluminium-manganese alloy.
314A	High aluminium-manganese alloy.
371	Aluminium-tin alloy of high creep resistance.

The first three of these alloys were tested in the form of hot rolled bar but alloy 371 was heat treated. This material was quenched from 1100°C. followed by annealing for 1 hour at 800°C.

High temperature tensile results have also been given for these steels[43]. These results are summarised in Fig. 43. It will be noted that strain-age-hardening seems to occur in the three alloys resulting in a retardation of the rate of fall of tensile strength with increasing temperature. Alloy 371 shows a slight maximum in tensile strength at about 500°C. The commercially pure titanium does not appear to strain-age-harden as judged by/

by the tensile strength but the curve of stress to give 0.5% strain does show such an effect. This difference indicates that at high strains rapid recovery or recrystallisation takes place. Similar effects show up during creep testing of the commercially pure titanium at high temperatures.

In ordinary steel the creep rate at room temperature falls rapidly and reaches a zero value even on testing at stresses well above the yield point. On the other hand, some steels can creep for long periods of time on testing at slightly higher temperatures. For example, it has been shown (Fig. 7) that at a stress of 12 tons per sq.inch and a temperature of 150°C., a 0.5% molybdenum steel casting is still creeping after 1,500 hours. It is to be expected, therefore, that this steel will continue to creep for long periods of time even if tested at room temperature. It is likely that the presence or absence of prolonged creep in steel at room temperature is controlled largely by the amount of nitrogen available to cause strain-age-hardening. This aspect of the problem is, however, obscure and worthy of further investigation, in view of its possible importance in the mechanism of fatigue.

As shown in Fig. 43 strain-ageing is not pronounced in commercially pure titanium and occurs at a higher temperature than in steel. It is, therefore, not perhaps surprising to find that titanium creeps quite extensively at room temperature as is shown by the strain-rate curves of tests on commercially pure titanium in Fig. 44. In each of the four tests three transitions in creep rate were obtained. The second of these is not very pronounced but since it occurs in all four tests it cannot be attributed to experimental error. The types of precipitate responsible for these ageing effects are more or less unknown but it is significant that in titanium considerable/

considerable amounts of three impurity alloying elements may be present in interstitial solution. These, of course, are nitrogen, carbon and oxygen. In an iron-carbide alloy the affinity of oxygen for iron is such that below 1000°C. virtually all the oxygen in solution is removed as iron oxide. In steel, of course, manganese and silicon eliminate oxygen from solution even more effectively. One type of "steel", namely, weld metal is, however, exceptional. This is cooled rapidly from a temperature in excess of 2000°C. so that it is probable that some oxygen is retained in interstitial solution. It would be interesting to find out if an all weld metal test piece shows a transition in creep rate due to oxygen. Some such effect must occur since the author has found that a weld metal containing little else than 0.05% carbon and 0.35% manganese has creep and rupture properties as good as ordinary 0.5% molybdenum steel.

The commercially pure titanium was also tested at 350°C. giving the strain-rate curves shown in Fig. 45. The curves for the test at 7 tons per sq.inch shows two transitions in creep rate. The third, if detectable at this temperature, must take place at a very low strain value. On testing at a stress of 9 tons per sq.inch or higher, simple curves of the pure metal type (Fig. 12) are obtained. The curves show no transitions in creep rate because recovery takes place so rapidly that ageing cannot occur. If the test at 7 tons per sq.inch had been continued for a longer time it is possible that recovery would set in at a higher strain value causing anomalies in the shape of the creep curve. As already noted the occurrence of rapid recovery during creep was to be expected in view of the shape of the tensile curve obtained (Fig.43).

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Strain-rate curves for room temperature creep tests on alloy 314 are shown in Fig. 46. This alloy has a better creep resistance than the commercially pure titanium but the results confirm that three transitions in creep rate are obtained in room temperature tests. Similar curves could be shown for alloy 314A which has a still better creep resistance.

Only two test results were available to indicate the trend in tests at constant stress and varying temperature. These tests were carried out on alloy 371 at temperatures of 400°C. and 500°C. using a stress of 25 tons per sq.inch. The results are shown in Fig. 47. There seems no reason to doubt that if more tests had been carried out at say, 25°C. intervals, then a family of curves would have been obtained similar to that obtained with carbon-manganese steel (Figs. 24 and 25). Hatched curves have thus been added to the diagram to show the probable result of such tests.

EFFECT OF TEMPERING OR AGEING

Although heat treatment variables are outwith the scope of this paper it was considered important to give some indication of the effect of tempering (or ageing) treatments. Tempering may be necessary to obtain the required mechanical properties at room temperature. It may be thought desirable because it improves the creep resistance as measured by a short time creep test. In many cases it is mandatory to temper (or stress-relieve) finished components before they go into service. The effect of the latter treatment is particularly important because to ensure complete stress-relief it may be necessary to temper for a very long time or at a very high temperature. It should be realised that material of high creep resistance cannot be fully stress-relieved/

stress-relieved without destroying the high creep resistance. In such cases a residual stress of a few tons per square inch may remain after a moderate tempering treatment but this should not give rise to trouble in service.

In the initial stage of testing a solution treated alloy acts essentially like a simple solid solution. Thus, in a creep test considerable deformation occurs before the creep rate is reduced to a low value due to strain-hardening and later strain induced precipitation. In an alloy such as a normalised alloy steel the equilibrium amount of alloy carbide does not have time to form on cooling but iron carbide and nitride can precipitate in the dislocations. Since such particles are not very effective in preventing creep the deformation in the initial stage of a test is appreciable.

On tempering a solution treated alloy precipitation may occur. With a normalised alloy steel, alloy carbide is precipitated in the dislocations replacing the less stable iron carbide. In both cases the presence of the alloy precipitate before testing ensures that the creep rate in the initial stage of testing will slow down more rapidly than that of the test in the solution treated or normalised condition. Tempering, however, removes alloying elements from solution so that less remains to be precipitated during testing. Thus, at a later stage of testing the creep rate of the tempered test may become faster than that of the solution treated or normalised alloy. This cross over point varies depending on the stress or temperature of testing and on the amount of tempering which has been given.

Although the above is a reasonably satisfactory explanation of the behaviour of alloys on tempering it is possibly not the full explanation. On cooling from the normalising temperature more vacancies and possibly more dislocations will exist than will be the case on cooling at a/

a much slower rate. Tempering may thus tend to reduce the number of vacancies and dislocations present. Such a phenomenon would help to reduce the creep rate in the initial stage of testing. However, all such questions must remain unanswered until more is known about the various types of lattice defects and their influence on the mechanism of creep.

The author has studies the effect of tempering on the creep resistance of various steels(4, 23). For the present purpose the effect of various tempering treatments on the behaviour of a 0.5% molybdenum steel will be taken as typical. The effect of tempering on the creep resistance of two carbon-manganese steels is also discussed. The treatment given in the latter case approximates to that commonly used for stress-relieving. The steels tested were as follows:

Steel	Al. Lbs/Ton	Form	Treatment	C.%	Si.%	S.%	P.%	Mn.%	Mo.%
17	¼	1" Dia.	Various	.09	.22	.032	.018	.48	.50
30	0	1" Plate	A) N. 920°C. B) N. + 2 Hrs. 650°C.	.17	.11	.034	.015	.61	-
31	0	3" Plate	A) N. 920°C. B) N. + 2 Hrs. 650°C.	.135	.14	.036	.020	1.20	-

All the above steels were of the basic open hearth type.

The 0.5% molybdenum steel 17 was tested at a temperature of 550°C. using stresses of 4, 6 and 9 tons per sq.inch. The tests were carried out both in the normalised condition and after tempering for 5 hours at 650°C. Strain-rate curves of these tests are shown in Fig. 48. It will be noted that whereas the normalised test is the better at all strain levels on testing at 9 tons per sq.inch, the tempered test is the better up to a strain of 0.25% on testing at 6 tons per sq.inch, and thereafter it is worse. /

Presumably at 4 tons per sq.inch the tempered material would be better up to a still higher strain.

The effect of varying the tempering time is shown in Fig. 49. These tests were carried out at a stress of 6 tons per sq.inch and a temperature of 550°C. In the normalised condition the creep rate initially is quite fast but slows down rapidly due to the various transitions in creep rate. After tempering for times up to at least 20 hours at 650°C. the creep rate initially is better than that of the normalised test but at a later stage it becomes worse. After 100 hours the creep resistance is poor because much of the molybdenum has been precipitated as carbide and spheroidised so that it is no longer effective in reducing the creep rate.

When the above tests were repeated using the same stress but at a temperature of 600°C. all the tempered tests were worse than the normalised sample even in the initial stage of testing. Even at 4 tons per sq.inch the tempered steel is inferior, as is shown in Fig. 50.

Obviously for applications involving long time service it is difficult to decide in any particular case whether or not tempering would be beneficial. The lower the temperature or the stress the greater the chance that the tempered material will be the better. This is particularly true when the design is such that only a small deformation can be tolerated.

In general a solution treated or normalised alloy will have a better rupture strength than the tempered or aged alloy, since tempering at a high temperature results in more spheroidisation than will occur with maximum ageing at the testing temperature. The tempered alloy may, however, be better if it breaks with a greater extension than the solution treated alloy. Also some precipitation reactions are so slow that the solution treated sample may extend to fracture and little or no precipitation has /

has taken place even under the accelerating influence of strain. In such cases tempering is essential. In still more complex alloys precipitation may be so sluggish that ordinary tempering or ageing is ineffective. In such cases good properties may be obtained by "warm working" at a high temperature to accelerate ageing.

From the above results it is obvious that a stress-relieving treatment of a few hours at 650°C. modifies the creep resistance of 0.5% molybdenum steel. Similar results could be shown for other low alloy steels. Stress-relieving also improves the initial creep resistance of ordinary carbon-manganese steels as shown by tests on the 0.6% manganese steel 30 at a stress of 5 tons per sq.inch and a temperature of 450°C. (Fig. 51). A series of tests at various temperatures on steel 31 in the tempered condition are shown in Fig. 32. The most interesting point about these results is the fact that the third stage of creep begins at very low strains. At 475°C. it occurs at about 0.05% strain, and presumably at lower testing temperatures the third stage of creep will begin at even lower strain values.

There are still many aspects of the effect of tempering which are still obscure. One such problem is the effect of initial structure on the strain at which a particular transition in creep rate may begin. The results in Fig. 32 compared with those in Fig. 25 suggest that by tempering to improve the initial creep resistance all the transitions in creep rate occur at lower strain values compared with tests in the normalised condition. This might have been expected since the probability of a transition beginning depends on the density of dislocations and the number stationary at any instant. The number of dislocations stationary at any instant must increase the slower the creep rate. Thus, because tempering slows down the creep rate rapidly in the initial stages, the/

the transitions in creep rate occur at low strain values. For the same reason the greater the number of transitions in creep rate the lower the strain at which the last one is liable to occur. For example, in Fig. 34 the transition due to molybdenum begins at about 0.1% strain in the complex alloys, whereas it is nearer 1.0% strain for the ordinary 0.5% molybdenum steel. In a molybdenum steel with low manganese (Fig. 8) the transition due to molybdenum must occur at a still higher strain value.

Other problems such as the influence of prior treatment on the behaviour after tempering[4] could also be mentioned but it would serve no useful purpose to discuss the effect of tempering at greater length in the present paper.

EXTRAPOLATION

The study of creep can be intensely interesting but many investigators can devote only a part of their effort to the solution of questions of theoretical importance. Of necessity most of the available creep testing equipment must be used to accumulate data which will be useful to the design engineer. In many applications the components have to withstand the service conditions without failure, or without deforming more than a permissible amount, for 100,000 or even 200,000 hours (11.4 to 22.8 years). Since tests of this duration are impracticable it is of vital importance to be able to estimate the long time properties of any particular material from tests of much shorter duration.

The difficulties and uncertainties of the most commonly used methods of extrapolation have been reviewed by Bailey[5] and it would appear that as yet no method has been devised which can be considered satisfactory. For the present purpose use will be made of the method of plotting of one well-known method of extrapolation which uses the results of a series of/

of constant stress creep tests at different testing temperatures. From the creep curves the times to reach various strains are measured and the results plotted using temperature and log time as ordinates. By drawing smooth curves through these points a family of iso-strain curves can be constructed. It is then usual to extrapolate these curves to longer times but as was discussed in relation to Fig. 1, the shape of the iso-strain curve up to any given time is not necessarily a guarantee that the same trend will continue in longer times. By making use of strain-rate curves, however, this difficulty can be overcome in the following manner.

If the creep test results are used to construct a family of strain-rate curves it is then possible, at least approximately, to deduce the shape and position of the strain-rate curves of tests at lower temperature. One such series of strain-rate curves is shown in Fig. 52 which represents tests on the 1.2% manganese steel 31 in the normalised and tempered condition. Leaving aside for the moment any question of accuracy, the times for various strains up to 1% can be measured from the actual experimental data and plotted on a temperature-log time diagram. From the estimated strain-rate curves it is an easy matter to obtain the ordinary strain-time creep curves by graphical integration. The times for various strains obtained from these curves can also be plotted on the temperature-log time diagram. A family of iso-strain curves can then be constructed. The results obtained on steel 31 are shown in Fig. 53. The full curves cover the region of actual experimental data. Hatched curves have been used at the top of the diagram because as the temperature is increased in this region the time for any given strain decreases rapidly in a manner which is not related to the shape of the curves at lower temperature. This can result from the tempering effect of holding the testpiece for some time at the test temperature before loading. (This phenomenon has already been discussed). It can also/

also occur because of rapid recovery above a certain test temperature or because of a rapid increase in the initial plastic extension with increasing temperature. It may even occur due to scaling or decarburisation of the test pieces at high temperature. The top part of the diagram can thus be ignored though it serves to indicate that short time tests at high temperature are of doubtful accuracy even for comparing materials or as acceptance tests.

The hatched curves in the lower part of Fig. 53 indicate the region of extrapolation. It will be noted that the iso-strain curves show a distinct bulge. This becomes more pronounced and tends to occur at longer times the lower the strain value considered. Since this must relate to the fact that the minimum in creep rate occurs at a lower strain the lower the testing temperature (Fig. 52), and since the latter is a consequence of strain-ageing or strain induced precipitation, it seems clear that the bulge in the iso-strain curves is also due to strain-ageing. Assuming that in the absence of strain-ageing, an iso-strain curve would be of a smooth concave shape, then due to strain-ageing, the curve is displaced to longer times. Considering any given strain, then at a high testing temperature this strain will be reached when the creep rate is still decreasing so that ageing has had less than its maximum effect in increasing the time to reach this strain. At a lower testing temperature the given strain will be reached in a time which coincides with the point of minimum creep rate. In this case ageing has had its maximum effect in increasing the time to reach the given strain. At still lower temperatures the minimum creep rate is reached at a low strain value and the creep rate is accelerating when the given strain is reached. Thus, once again ageing has had less than its maximum effect. From the above it seems evident that ageing distorts the/

the smooth concave iso-strain curves to form the bulges shown in Fig. 53. In cases where the strain-rate curves show more than one minimum in creep rate, more than one bulge will be obtained in the iso-strain curves, although they will not necessarily be so pronounced as those shown in Fig. 53. For example the 0.5% molybdenum steel (Fig. 28) would have shown two minima in creep rate if the tests had been prolonged to higher strains. The iso-strain curve for a given strain should thus show two bulges in different temperature regions.

It can be concluded, therefore, that the iso-strain curves in Fig. 53 are of the correct shape. It is, however, necessary to make some estimate of the accuracy of the extrapolation. A complete review of this aspect of the subject must await more detailed testing and in any case, is so important as to require a separate paper. A few remarks can, however, be made. The accuracy of the extrapolation depends on three factors, namely, material variations, testing errors and errors in the estimated strain-rate curves. Material variations within a cast and between cast to cast can be assesses by a sufficient number of duplicate tests. Testing errors at strains of 0.1% or above should not be great using modern high sensitivity creep testing equipment. It is, however, difficult to obtain precise agreement at low strain values, particularly below 0.01% but this is of little consequence when considering long time tests. For the same reason it is difficult to define the exact position of the first transition in creep rate in Fig. 52 but even if this were neglected it would lead to only a few hours error in the extrapolation. The measurement of creep rate at longer testing times is comparatively easy and should present no difficulty. The author, however, finds a large set of ship curves invaluable when drawing smooth curves through the plotted points of the creep test. If, however, the stress and temperature of testing is low and also of course the creep rate, the/

the experimental variations in the strain points make it impossible to estimate the creep rate accurately over a small range of time. In such cases the trend over many thousands of hours must be used to assess the creep rate. As the creep rate approaches a value of 10^{-8} inch per inch per hour or less it becomes almost impossible to assess the creep rate, simply because it would require tens of thousands of hours of test to show the trend accurately. Such tests, however, are entering the region where it is necessary to estimate the strain-rate curves from the pattern obtained at higher testing temperatures.

As explained previously the estimated strain-rate curves were drawn simply from an examination of the pattern of the known strain-rate curves and using what was considered to be a slight underestimate of their probable position. This seems to be a justifiable procedure since the very long time tests in Fig. 28 seem to show a regular pattern even at low testing temperatures. It is to be hoped, however, that as more information is obtained over a variety of test conditions and with various materials, it will be possible to define more clearly the laws governing the trend of the curves.

In Fig. 53 the estimated time to reach 0.1% strain at a temperature of 425°C. is 116,000 hours and from the graph the temperature to give 0.1% strain in 100,000 hours is 427°C. It is known that the error in the estimated time of 116,000 hours may be considerable but it can be safely assumed that, in the extreme, the true time is not less than half the above nor more than twice as much. This means that the time for 0.1% strain at 425°C. lies between 60,000 and 240,000 hours. Although this variation looks formidable it does not make a very big difference in the estimated temperature for 0.1% creep in 100,000 hours. In fact this temperature will lie between 418°C. and 439°C. Neglecting the odd degree it can be concluded therefore that at a stress of 5 tons per square inch the temperature for/

for 0.1% strain in 100,000 hours will be 427°C. ± 10°C.

By using the strain-rate method of plotting it has been possible to make an accurate estimate of the long time properties of a steel from a few tests of up to 10,000 hours duration. The big advantage of this method lies in the fact that the probable error in the estimate can be calculated. As the method is refined it should be possible to reduce the probable error to still lower values.

Ridley[7] has published creep test results at 5 tons per square inch on a carbon-manganese steel the maximum duration of tests being 70,000 hours. Fig. 54 shows the actual test points through which he drew his iso-strain curves. Since from Fig. 26 it was known that this steel shows strain-rate curves similar to those in Fig. 52, the iso-strain curves were re-drawn to fit the plotted points more closely. The curves so obtained are shown in Fig. 54. The similarity to Fig. 53 is obvious. These results thus serve to confirm that the shape of the iso-strain curves in Fig. 53 is correct.

Rupture tests are generally carried out at constant temperature and and the results plotted using the log of the stress and the log of the time to fracture as ordinates. Since with this method of plotting the test results often lie approximately on a straight line it has been assumed that this straight line relationship holds for still longer times. In this way an estimate of the rupture strength in 100,000 hours at the test temperature can be obtained. Even if such a relationship could be proved to hold for one material it is not necessarily correct for other materials or even for the same material at a different test temperature. The dangers of this method of extrapolation were indicated in Fig. 2 with reference to the elongation at fracture in rupture tests.

In Figs. 20 - 23 four series of strain-rate curves to fracture are shown. These will be used to illustrate the possibilities of this/

this method although it is difficult in this case to estimate the probable accuracy of the extrapolation. The a ccuracy would be improved, however, by longer time tests. It would be helpful also if the position of the estimated strain-rate curves was cross-checked by carrying out tests at constant stress. Such tests are in hand and an improved design of extensometer is being used for these high strains.

It can be assumed, however, that the estimated strain-rate curves at 8 tons per square inch in Figs. 20 - 23 are reasonably accurate since they are only one step away from the known curves. The creep curves to fracture were calculated by graphical integration with the results shown in Fig. 55. Although the elongation at fracture is not known this factor does not greatly influence the estimated time to fracture for steels 13, 14 and 16. With steel 15 an error is introduced which is probably not large since the elongation of this steel should be high. An interesting feature of these results is the prolonged third stage of creep shown by the two 1% manganese steels. This, of course, was self evident from the shape of the strain-rate curves. It should also be noted that, except for steel 13, the beginning of third stage creep does not denote the eminence of failure. The third stage of creep occupies by far the largest proportion of the time to rupture.

The time to rupture at lower stresses was also calculated from the strain-rate curves. All the results are shown in Fig. 56. It seems evident that even allowing for a large error in the estimated time to fracture the rupture curves bend downwards with increasing testing times. While making no guarantee regarding the above tests it would appea. possible with a slightly more elaborate set of results to guarantee the rupture strength to within $\pm$ 0.5 tons per square inch. As with the creep data this permits quite a substantial variation in the estimated times to fracture.

It can be concluded, therefore, that the strain-rate method of plotting is of considerable help in the extrapolation of rupture test data as well as in the extrapolation of creep date. In both cases, however, more detailed investigation will be necessary before the laws relating the shapes of the strain-rate curves can be formulated. A precise estimate of the variation of activation energy at different stresses, temperatures and times should be of considerable help in this connection. The experimental difficulties are, however, rather formidable since very exact techniques are necessary to determine this even for the simple case of pure metals.

RUPTURE DUCTILITY

One very important aspect of the creep problem remains to be considered, namely the elongation obtained in long time rupture tests. At ordinary temperatures most metals and alloys fracture in a transcrystalline manner whatever the rate of straining. At high temperatures failure is usually transcrystalline if the metal is broken quickly but with slow rates of straining failure may be wholly or partly intercrystalline. Under creep conditions failure is almost always intercrystalline.

The amorphous theory of grain boundaries suggested by Rosenhain[45] has now given way to the belief that in the narrow zone between crystals there is a "transition lattice"[46]. At high temperature this zone behaves in a quasi-viscous manner so that when a stress is applied one crystal slides over the other. This has been proved by direct measurement of a bicrystal[47]. The internal friction experiments of Ke[48] also support this viewpoint.

It is difficult to see why this grain boundary movement should lead to intercrystalline failure of a pure metal. On the application of a very small stress the boundaries flow under shear stress until the/

the elastic strains set up in the crystals produce a system of forces in which there are no shear components along the grain boundaries. Under ordinary stresses, however, creep occurs within the body of the crystals so that the grain boundaries continue to flow in an attempt to relieve themselves of stress. Stress is thus concentrated at grain corners and on grain boundaries transverse to the direction of the load. In a pure metal it is to be expected that after such stresses have caused a little plastic deformation the stress concentrations would be dissipated by recovery, recrystallisation or grain boundary migration so that intercrystalline failure would not occur. In some of the tests reported in the literature it is therefore possible that the so called pure metals used were not pure enough and that the intercrystalline failures resulted from the presence of impurities.

The mechanism of failure at high temperature has recently been receiving considerable attention. For example Crussard & Friedel[49] suggest that voids are formed by the condensation of vacancies and that these voids lead to intercrystalline failure. Machlin[50] on the other hand concludes that this is highly improbable but that growth of pre-existing voids by vacancy condensation is probable. It would appear therefore that this aspect of the problem is still in a state of flux and that a satisfactory theoretical solution is still not in sight.

However, intercrystalline failure of itself is not of great practical importance. It is important only if it leads to failure after a low extension. This seldom, if ever happens, with a pure metal or simple solid solution. A brittle network of a second phase round grain boundaries leads to intercrystalline failure with a low extension but this phenomenon is easily detected by a simple tensile test and such alloys can be avoided. In many cases, however, failure with low elongation only occurs in rupture tests of long duration, the alloy being quite ductile in shorter time tests.

This cannot be due solely to grain boundary flow since with tests of still longer duration the elongation again increases(10).

There seems little doubt that intercrystalline failure with a low elongation results when the relative strength of the body of the crystals, compared with that of the grain boundary regions, exceeds a certain critical value. This idea was expressed many years ago by Bailey(51). Such an effect, however, cannot occur unless some structural change takes place during testing which augments the strength of the crystals to a greater extent than that of the boundary regions.

The close correlation between the strain-age-hardening phenomena detected by high temperature tensile tests(10) and the transitions in creep rate in the creep tests is convincing proof that the latter are also related to ageing phenomena. Similarly it was found(10) that for each maximum in stress in the iso-strain curves of the high temperature tensile tests there was a corresponding minimum in reduction of area at about the same temperature. It was concluded, therefore, that these minima in reduction of area were also a consequence of strain-age-hardening. Microscopic examination of the fractures of a large number of tensile specimens showed some interesting features. When the reduction of area was of the order of 10 per cent the fractures were wholly intercrystalline and intercrystalline cracks could be detected a considerable distance from the main fracture. At somewhat higher reductions in area, an apparently normal cup and cone fracture was obtained. However, examination revealed that the flat area of the centre of the test-pieces showed intercrystalline cracks, whereas at the sides failure was by shear. In other words failure was initiated by intercrystalline cracks near the centre of the tests and changed to shear near the periphery.

In view of the intercrystalline nature of the fractures in these tensile tests which were carried out at a very slow rate of straining and the/

the probability of the low reductions in area being related to strain-ageing phenomena, it seems quite possible that the low elongations which can be obtained in long time rupture tests at high temperature are also related to strain-ageing or strain induced precipitation phenomena. If this assumption is correct then the low elongations in rupture tests must be in some way related to the transitions in creep rate obtained in the creep curves of such alloys.

Unfortunately relatively few creep tests are carried out to fracture and even so the extensometer may be removed an an early stage. Rupture tests in general are used simply to measure the time and elongation at fracture. A few strain-rate curves to fracture have been reported in the earlier part of this paper. A few other results were available. Of particular interest are some creep tests made on a steel which because of its treatment proved to be very brittle, failure occurring after less than 1% extension. Details of these materials in the order they will be considered are as follows.

Steel	C.%	Si.%	Mn.%	Cr.%	Mo.%	V.%	Ti.%
14	.12	.06	.52	-	-	-	-
16	.13	.04	1.0	-	-	-	-
32	.14	.08	1.55	-	-	-	-
25	.10	.31	.49	-	.52	.25	.28
17	.09	.22	.48	-	.50	-	-
18	.10	.14	.51	.82	.54	-	-
Nimonic 80A	Nickel base alloy see page 55						
G.18B	Austenitic alloy see page 55						

Strain-rate curves to r~~e~~upture on steels 14 and 16 are shown in Figs. 22 and 23. It will be noted that whereas the 1% manganese steel 16 fractured with an elongation of about 30% the 0.5% manganese steel 14 failed in two tests at about 12% elongation. In no case has the/

the failure any apparent connection with the transitions in creep rate.

In ordinary tensile tests either the elongation or the reduction of area can be taken as a measure of ductility. The author considered the latter to be the better criterion when comparing a variety of materials because the elongation depends largely on the work hardening characteristics. Necking in a tensile test begins at the point when the rate of strain-hardening becomes equal to the rate at which the stress would be increased by further deformation at constant load. Up to this point any section of the gauge length which necks down slightly is stronger than any other section so that the extension remains uniform. Beyond this point a section which necks down slightly is weaker than the rest because work hardening does not quite compensate for the local increase in stress due to the reduction in area. As a result necking once started proceeds at the same section until fracture occurs.

A similar phenomenon must also take place during constant load creep tests to fracture but it will be further complicated by such factors as the change in creep resistance with increasing extension. As the extension exceeds, say, 1% the actual stress on the test piece begins to increase appreciably. As the stress increases the degree of work hardening must also increase. The elongation will remain uniform so long as this factor compensates for the local increase in stress due to slight necking. From the results in Figs. 22 and 23 it is clear that the high manganese steel 16 has a greater work hardening capacity under the test conditions than steel 14 so that the elongation at fracture for steel 16 is also greater. The lower elongation of steel 14 is thus not a sign of embrittlement. In fact if the ductility is taken to be the reduction of area the low manganese steel is the better, having a reduction of about 50% compared with about 40% for the higher manganese steel. In both cases, therefore, fracture can be taken to be the/

the result of the normal mechanism. Although this mechanism is unknown it must be connected with the build up of stress concentrations due to grain boundary flow. These will increase with increasing strain since in this case, and unlike that of pure metals, the recrystallisation temperature is quite high.

If, however, the above steels are compared with the 1.55% manganese steel 32 a different phenomenon makes its appearance. This is apparent from Fig. 57 showing strain-rate curves to fracture at a stress of 14 tons per sq.inch and a temperature of 450°C. From the previous argument it might be expected that the 1.55% manganese steel would fail with a greater extension than the 1% manganese steel. It was also to be expected that this steel would show a longer time to fracture. In actual fact the extension was only 12% and the time to fracture was less than that of the 1% manganese steel. This low elongation and time to fracture is due to some form of embrittlement since the fracture was markedly intercrystalline and the reduction of area was only 15%. The significant feature of the result on steel 32 is that the strain-rate curve does not show a transition due to manganese but fractures quickly at about the strain where the manganese transition should just be beginning. The expected curve showing this transition is shown by the hatched curve in Fig. 57. It can be concluded therefore that when a transition in creep rate begins the chance of intercrystalline failure increases. As the transition in creep rate nears completion the chance of failure decreases and the alloy may extend considerably before failure eventually occurs.

Another example of this type of failure was given by a series of creep tests on the molybdenum-vanadium-titanium steel 25. This steel was heat treated by holding at 1150°C. for 1 hour before air cooling. It was then tempered for 5 hours at 750°C.. Though this treatment gave the maximum creep resistance it was found that the structure had coarsened considerably and the steel was brittle. (It was subsequently found that 10 minutes at 1150°C. was/

was sufficient to prevent this grain coarsening and that a better ductility was obtained without much sacrifice of the creep resistance). Fig. 58 shows part of the creep test curve on this steel at a stress of 7 tons per square inch and a temperature of 625°C. At a time of 2,500 hours the creep rate was accelerating but at about 3,500 hours and at a strain of 0.225% a marked transition in creep rate began and the creep rate decreased rapidly. At 4,600 hours, however, there was an abrupt change in creep rate which accelerated suddenly. Fracture took place at 4,800 hours with virtually no measurably elongation on reduction of area. The fracture was, of course, intercrystalline. Thus, in this case also, fracture took place near the beginning of a transition in creep rate.

The strain-rate curve of the above test and of tests at 6, 8 and 10 tons per square inch are shown in Fig. 59. The test at 6 tons per square inch was removed after 12000 hours because at this temperature scaling was severe and much longer time would probably have been required to cause fracture. It will be seen, however, that at this stress the steel had survived the pronounced transition in creep rate without fracture and it may be that the elongation would have been appreciable. At a stress of 7 tons per square inch the transition should have been nearly as pronounced as at 6 tons per square inch. The probable curve is shown by a hatched line. However, because of intercrystalline failure there was an abrupt increase in creep rate as was indicated in Fig. 58. Similarly at a stress of 8 and 10 tons per square inch there is an abrupt change in creep rate due to intercrystalline failure. In these cases, however, it occurs nearer to the beginning of the transition though at a higher strain because the transition begins at a higher strain with increasing stress. No doubt the manganese steels would have shown a similar pattern if they had been tested under a variety of stress and temperature conditions. It is unlikely, however, that very low/

low extensions could be obtained in such steels since under the conditions that failure during a manganese transition is liable to happen, the transitions themselves occur after quite a high extension.

Although creep tests to fracture were not carried out on the 0.5% molybdenum steel 17 and the 1% chromium 0.5% molybdenum steel 18, some conclusions can be made from the available data. From previous work[4] the elongation at fracture during long time tests at 550°C. was known. From these tests the elongation-log time to fracture curves shown in Fig. 60 were obtained. The stresses to cause fracture have been indicated on these curves. It will be noted that a minimum in ductility occur after about 1000 hours with the 0.5% molybdenum steel but that with the chromium-molybdenum steel the elongation is still decreasing after 70,000 hours of test. It will also be noted that the minimum elongation of the former steel is about 3% whereas after 70,000 hours the extension of the chromium-molybdenum steel is still about 8%.

From the strain-rate curves in Figs. 27 and 28 and from the known times and elongations to fracture it is possible to make an approximate estimate of the shape and position of a series of strain-rate curves to fracture at 550°C. for the 0.5% molybdenum steel. This is shown in Fig. 61. The position at which the transition due to molybdenum begins is indicated by a chain dotted curve. It seems clear that at stresses of 14 tons per square inch and down to 9 tons per square inch failure occurs at or near to the beginning of the molybdenum transition in creep rate. The elongation at the same time falls to 3% because the transition begins at a lower strain the lower the testing stress. At a stress of 8 tons per square inch the steel survives the beginning of the transition without failure and breaks at an extension greater than 3%. The behaviour is thus similar to that of the molybdenum-vanadium-titanium steel although, of course, the elongation at fracture is higher.

Similarly from the results in Fig. 29 and the known times and elongations to fracture a family of strain-rate curves can also be constructed for the chromium-molybdenum steel. This is shown in Fig. 62. In this case fracture at or near the beginning of the molybdenum transition occurs at stresses of 12 tons per square inch or higher. Thus if this steel had behaved like the 0.5% molybdenum steel a minimum in elongation would have been obtained at an elongation of about 20% at a stress of 12 tons per square inch and in a time of 1000 hours. In actual fact of course the elongation of the chromium-molybdenum steel continues to fall slowly as the stress of testing is reduced. Dorn, however, has stated(11) that the strain damage due to grain boundary flow is concentrated more and more in the vicinity of the grain boundaries as the stress of testing is reduced. Thus, the elongation at fracture should decrease as the stress is reduced. This tendency will, however, be reversed if the structure of the steel changes due to softening or spheroidisation. Because of its lower alloy content the 0.5% molybdenum steel is less resistant to spheroidisation than the chromium-molybdenum steel. This factor could account for the difference between the steels. Nevertheless it can be concluded that intercrystalline fracture with low elongation is very probably due to failure during a transition in creep rate. When the elongation is higher failure may still be due to this effect but it can also result from other phenomena. It is clear, however, that all the factors influencing the elongation at fracture require much more investigation.

The interesting feature of the above results is that the addition of about 1% of chromium to a 0.5% molybdenum steel increases the elongation at fracture. Before dealing with this point, however, mention should be made of a few other strain-rate curves to fracture. Some results on Nimonic 80A are shown in Fig. 42. As the stress decreases from 16 to 10 tons per square inch the elongation at fracture decreases. At a stress of 7 tons per square inch the elongation increases and the material shows a pronounced/

pronounced transition in creep rate beginning at about 0.3% strain. It seems probable therefore that at the higher stresses intercrystalline failure has occurred near to the beginning of this transition. If failure could have been avoided a more pronounced transition in creep rate would have been obtained at these stresses. The evidence indicates, therefore, that Nimonic 80A at 700°C. behaves in a similar manner to the molybdenum-vanadium-titanium steel at 625°C. (Figs. 58 and 59).

The results on alloy G.18B shown in Fig. 41 are also interesting. At a stress of 6 tons per square inch and a temperature of 850°C. fracture took place at about 10% elongation and seems to occur just at the beginning of a transition. At 750°C. on the other hand the test had survived the transition without failure which took place eventually at about 15% elongation. It is probable that the elongation will increase at still lower temperatures as is suggested by the hatched curves in Fig. 41. The results on this alloy thus support the conclusions already made.

To explain the effect of chromium on 0.5% molybdenum steel it is necessary to obtain some idea of the mechanism of grain boundary creep. McLean[52,53,54,55] has shown that creep of polycrystalline metals occurs by dislocation migration in the grains resulting in slip and sub-grain tilting as well as by grain boundary shearing. He also shows that, for a given stress, the ratio of the fraction of the total plastic strain arising from grain boundary shear to the total plastic strain, is constant. Dorn and others[56] have confirmed this finding. Furthermore, Dorn[11] has shown that this ratio remains constant for a given stress independent of testing temperature and concludes that the same activation energy and the same creep law holds for grain boundary shear as applies to the total creep of the pure metal. This, Dorn suggests, means that grain boundary shearing can be attributed to localised crystallographic mechanisms of deformation in the vicinity of the grain boundaries rather than to such a process as viscous shearing. Consequently the shape of the curve of/

of grain boundary creep against time is the same as that of the curve of total creep against time.

However, from the point of view of this paper it is more convenient to deal with strain-rate curves rather than ordinary creep curves. From the above it is clear that the shape of the strain-rate curve for grain boundary creep will be the same as that of a pure metal (Fig. 12). Also since the ratio of grain boundary strain to total strain is a constant it can be assumed that the ratio of the creep rate in the boundary zones to the actual creep rate will also be a constant. If these two creep rates are plotted against total plastic strain then, for a pure metal, strain-rate curves such as those in Fig. 63 will be obtained. Since the ratio of these creep rates at any strain is a constant the distance X between the curves is also constant for all strains.

Since Dorn's work suggests that the same creep law applies to grain boundary creep as for total creep of a pure metal it is a reasonable assumption that the strain - grain boundary creep rate curves of a complex alloy will also have the same general shape as the ordinary strain-rate curve of the alloy. It is possible that since the grain boundary creep rate is faster than the ordinary creep rate the transitions in the grain boundary creep rate curve will occur at a higher strain than those of the ordinary creep rate. However, this factor will be neglected. Initially therefore as shown in Fig. 64, the behaviour is similar to that of a pure metal. Also if a simple transition in creep rate occurs such as that of nitrogen or carbon in steel the distance X will not alter since such elements can diffuse very rapidly to the grain boundaries if necessary. This is also shown in Fig. 64.

If, however, a second transition in creep occurs which depends on the diffusion of an alloying element such as molybdenum the conditions are rather different. In the body of the crystals the alloying element does not have far to travel to cause precipitation within a dislocation. Migration to the grain/

grain boundary zones, however, requires more time. Thus, in this case the transition in grain boundary creep rate will occur at a higher strain than the ordinary transition. The strain-rate curves thus diverge as shown in Fig. 64. Since the grain boundary zones become relatively much weaker than the body of the crystal in the region of the second transition the chance of intercrystalline failure is increased.

If, however, an additional transition is introduced and this transition occurs at a strain somewhat less than the second transition in Fig. 64, a new factor is introduced. The effect of this is shown in Fig. 65. In this case the second transition occurs while the creep rate is still decreasing. The strain-rate curves thus do not diverge as much as in the example shown in Fig. 64. The chance of failure during the second transition is thus decreased. Also during the third transition in creep rate the effect of the second transition is still reducing the grain boundary creep rate so that the difference between the strain-rate curves is minimised. The chance of failure during the third transition is thus less likely.

No useful purpose would be served by attempting to extend the above hypothesis to any greater length. Direct proof of any such effect would be given by estimating the grain boundary creep of alloys by the methods used by McLean for pure metal. If the ratio of grain boundary creep to total creep were measured throughout a test the variation of this ratio could be used to prove or disprove the above hypothesis. In any case it is obvious that much more work on the effect of combinations of alloying elements would be necessary before arriving at a definite conclusion. It is interesting to point out however, that some published(9, 10) and many unpublished results of high temperature tensile tests on steels with a combination of alloying elements lend support to the suggestion that the presence of one alloying element may tend to minimise the embrittling effect of another alloying element.

CONCLUSIONS

1) A simple functional relationship[11] appears to fit the creep of pure metals or simple solid solutions with or without stable precipitates since the only submicroscopical change during creep testing is strain-hardening. Recrystallisation or grain growth during testing destroys the relationship.

2) The same relationship does not hold for commercial alloys and it is inferred that other submicroscopical changes must occur during creep testing of such alloys.

3) Evidence is presented to show that the structural change responsible is probably some form of coherent precipitation of a second phase in the dislocations during creep testing, either due to strain-ageing or a strain induced precipitation hardening effect.

4) It was concluded that such ageing effects should influence the shape of the creep curves obtained. In the third stage of creep the creep rate may slow down giving a new "primary" stage of creep before the creep rate again begins to accelerate. A similar effect was found to occur in the early stages of creep. In this case it consists of an acceleration in the rate of decrease of creep rate with time. These phenomena have been called transitions in creep rate.

5) It was realised that time or log time was a difficult criterion to use in assessing creep data and for this reason the variation of creep rate with strain was adopted. A simple theory was developed making use of log strain-log creep rate curves which show the transitions in creep rate more clearly. The main points of this theory are as follows:-

a) In a pure metal strain-hardening or rather the density of dislocations increases during primary creep and becomes constant during secondary creep. A similar effect occurs with alloys but is not related directly to primary or secondary creep. The density of dislocations increases as the stress of testing is increased.

b) Strain-hardening is the most important factor in the initial stage of a creep test, the movement and multiplication of dislocations being so rapid that strain-age-hardening has no time to occur.

c) As the rate of creep decreases and tends towards a constant value, the rate of formation of dislocations decreases and the number stationary at any instant increases. The strain-ageing constituent thus has more time to precipitate at points where it is stable, that is at points of maximum lattice strain. As a few dislocations become fixed in this way other dislocations pile up giving more time for ageing. The process is thus self aggravating and the creep rate slows down rapidly over a short interval of time. A transition in creep rate is thus obtained.

d) With increasing temperature of testing the probability of any dislocation "climbing" increases so that strain-ageing is less liable to occur. The transitions in creep rate thus occur at greater strains the higher the temperature of testing.

e) With a pure metal, tests at constant stress and varying temperature give a series of simply related strain-rate curves, the difference between the logarithm of the

the creep rates at any strain being a measure of the activated energy for creep. The curve for any lower temperature can thus be calculated.

f) When transitions in creep rate occur in a creep test the activation energy is no longer constant but from the results of a series of creep tests at constant stress a family of strain-rate curves can be obtained from which it is possible to obtain a good estimate of the shape of the strain-rate curves for lower temperatures. Thus, by integration it is possible to predict the whole of the creep curve at a low temperature and not just the time for some given strain.

g) In tests at constant stress and varying temperature the variation of dislocation density with increasing temperature of testing is not great so that a reasonably symmetrical series of strain-rate curves is obtained. On the other hand with tests at constant temperature and varying stress the initial dislocation density on loading may increase enormously as the stress of testing is increased. Since strain-ageing and strain induced precipitation are accelerated under such conditions the strain-rate curves obtained are more complex. For this reason it seems clear that constant stress tests are more amenable to extrapolation than tests at constant temperature.

6) From the results of high temperature tensile tests[10] it was known that strain-ageing phenomena exist corresponding to the presence of carbon, nitrogen, manganese, chromium, molybdenum etc., in ordinary ferritic steel (body-centred cubic). In all the many creep tests carried out it was found that the number of transitions /

transitions in creep rate obtained in any test was uniquely related to the number to be expected from the evidence of the tensile tests.

7) It was also shown that the presence of one alloying element modifies the strain-age hardening behaviour of other alloying elements. This results partly from solution hardening and partly from an increase in the recovery temperature but is also due to the affinity of the alloying elements for elements such as carbon and nitrogen in solution. This phenomenon is particularly important with reference to alloying elements such as vanadium and titanium.

8) The addition of 1% silicon to steel greatly improve the creep resistance in the initial stages of testing. It was shown that although the evidence of high temperature tensile tests is not conclusive, silicon does in fact modify the strain-age-hardening behaviour of a steel. It would appear that silicon nitride can precipitate in the dislocations in preference to iron nitride.

9) In face centred cubic type alloys the evidence of tensile tests suggests that high temperature strain ageing or strain induced precipitation is possible in such alloys. The results of creep tests on six austenitic steels, some of which were of very long duration, showed that transitions in creep rate do occur with such alloys. Creep tests on a nickel base alloy (Nimonic 80A) also showed transitions.

10) Creep tests on several titanium alloys were also included as representative of the hexagonal close packed type of lattice. The results of tensile tests in these alloys suggested the possibility of strain-ageing phenomena and this was confirmed by the presence of transitions in creep rate in the creep curves.

11) The possibility of using the strain-rate method of plotting to facilitate extrapolation of creep and rupture data to long testing/

testing times has been briefly considered using data on several carbon steels. It has been shown that quite an accurate estimate of the long time creep and rupture strength can be made even allowing for a considerable error in the predicted strain-rate curves. The accuracy of the estimate can also be assessed.

12) The versatility of the method will be appreciated when it is realised that it has been possible to correlate the results of creep tests for strains less than 0.01% up to strains greater than 1.0% and for creep rates varying by a factor of 10 million. The duration of the tests varied from about 50 hours up to a maximum of 100,000 hours. The total testing time of all the tests dealt with exceeds 900,000 hours which is more than 100 years of testing.

13) It was known from the results of high temperature tensile tests[10] that if the addition of an alloying element to steel produces a maximum in stress in the iso-strain-temperature curves it will also result in a minimum in reduction of area at about the same temperature. Failure was wholly or partly intercrystalline. It was concluded that intercrystalline failure with low elongation in high temperature creep rupture tests might also be the result of strain-ageing phenomena and that if this surmise was correct the failure must be related in some way to the transition in creep rate.

14) From strain-rate curves to fracture on a number of steels it was found that a number of factors must control the elongation in any given rupture test. However, when a low elongation was obtained, intercrystalline failure was initiated during or near to the beginning of a transition in creep rate. If the elongation was high failure may/

may also occur during a transition but it could result from some other mechanism.

15) The addition of an alloying element may minimise the embrittling effect of another alloying element. A tentative hypothesis has been put forward to explain intercrystalline failure during a transition in creep rate and to show why the addition of one alloying element may counteract the embrittling effect of another alloying element.

16) In conclusion it should be emphasised that although the above evidence throws some new light on the phenomenon of creep it also shows up the many deficiencies in our present knowledge of the subject. In particular it should be emphasised that very little is known about the effect of combinations of alloying elements and of heat treatment on the long time properties of alloys. On the theoretical side it would be very helpful if the exact mechanism of strain-ageing could be established. Perhaps the electron microscope may lead to a solution of the vexed question as to whether or not precipitates do exist in the dislocations or whether the term 'atmosphere' would be more appropriate.

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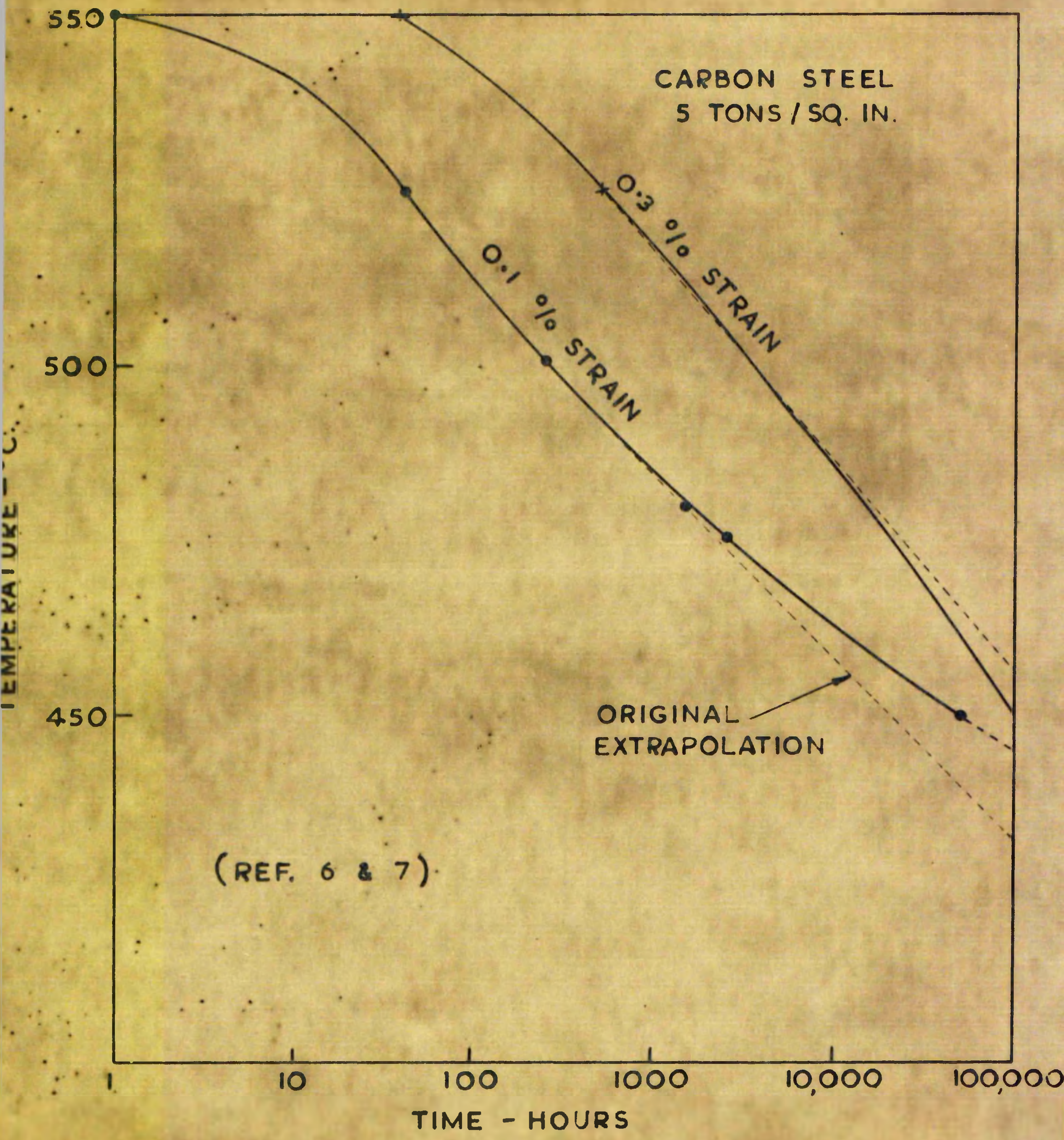
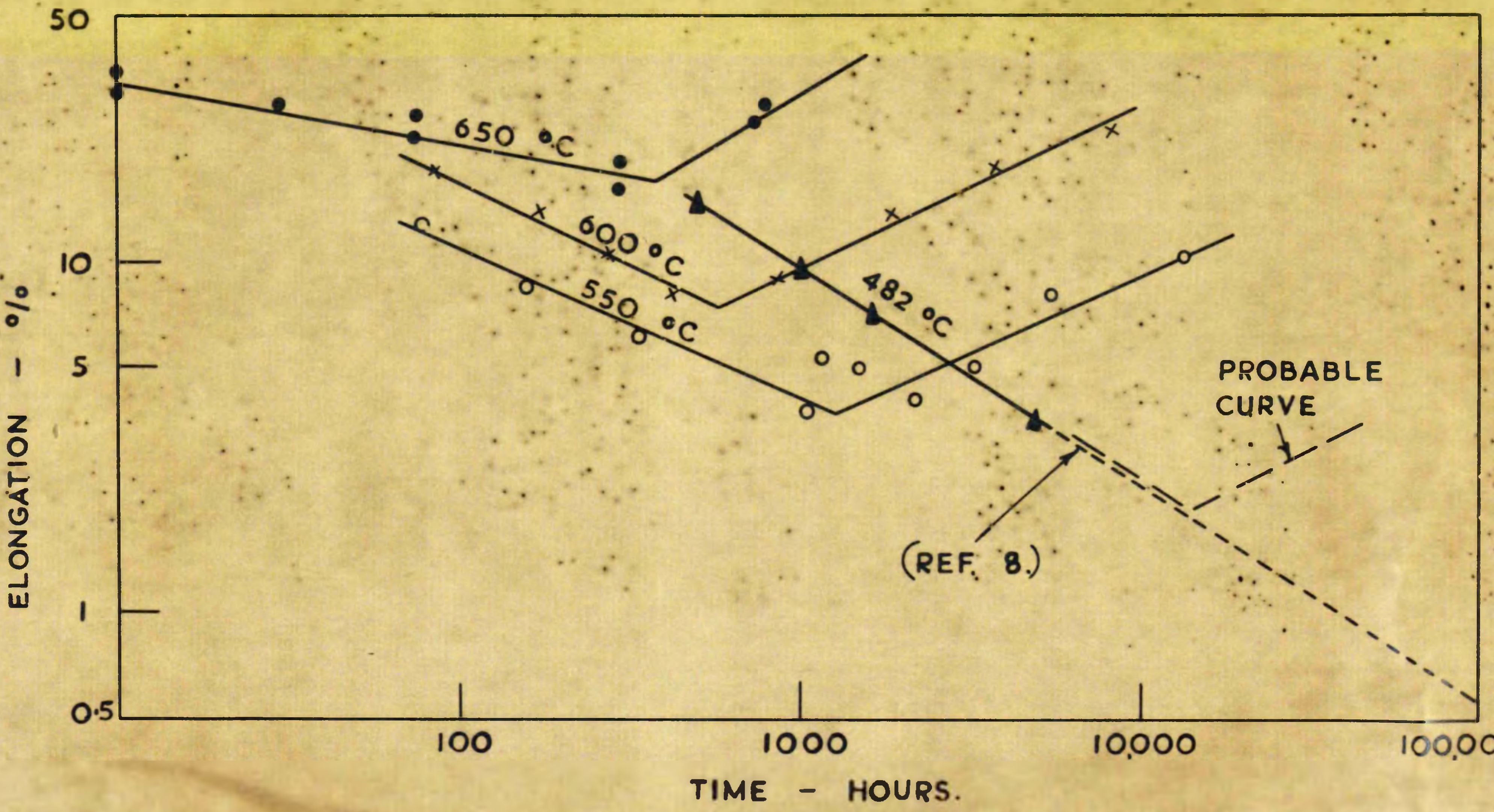


TEMPERATURE - LOG TIME ISO - STRAIN CURVES FOR CARBON STEEL.

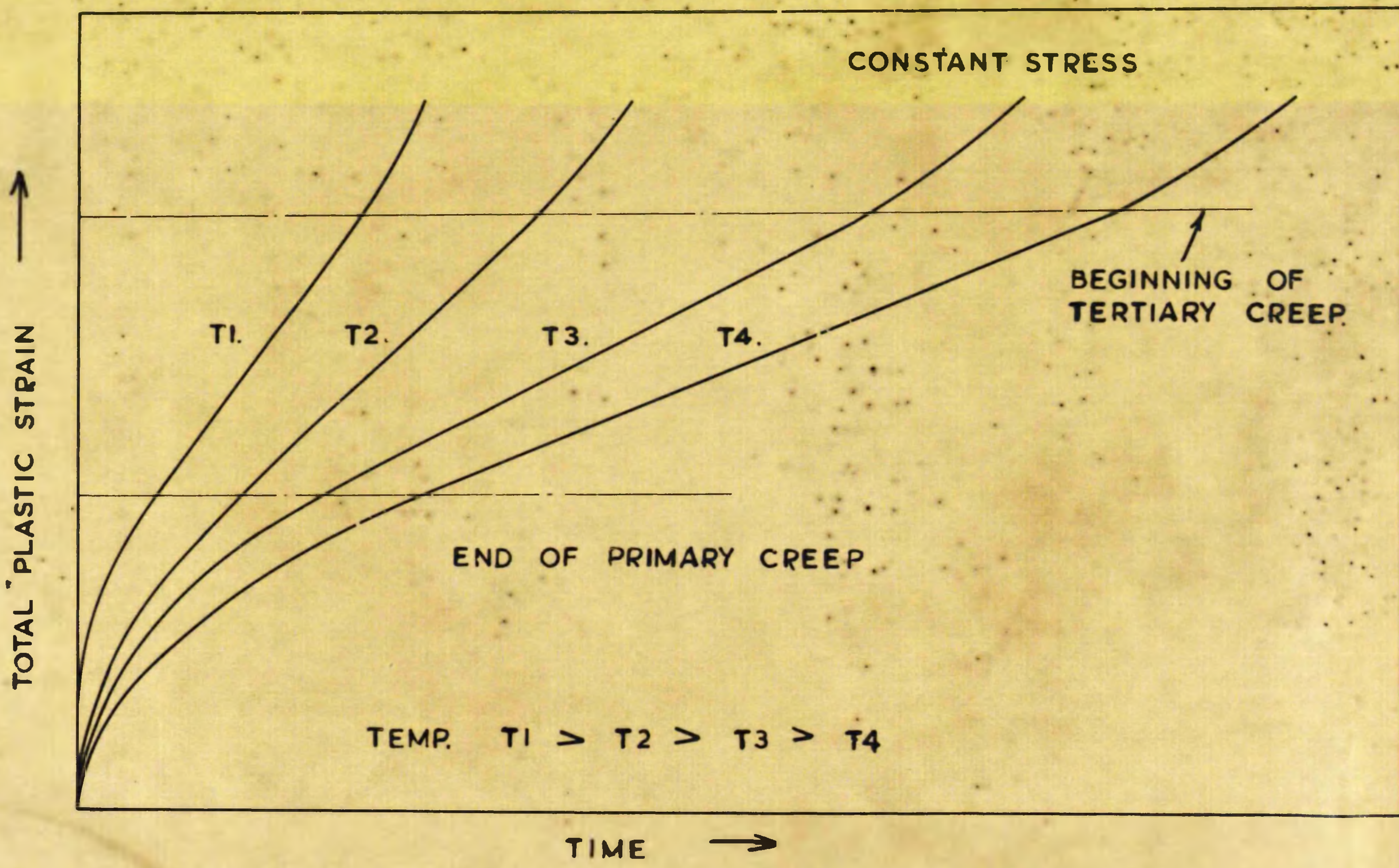
FIG. 1.



ELONGATION - TIME TO FRACTURE CURVES FOR
0·5 % MO, STEEL AT VARIOUS TEMPERATURES.

FIG. 2

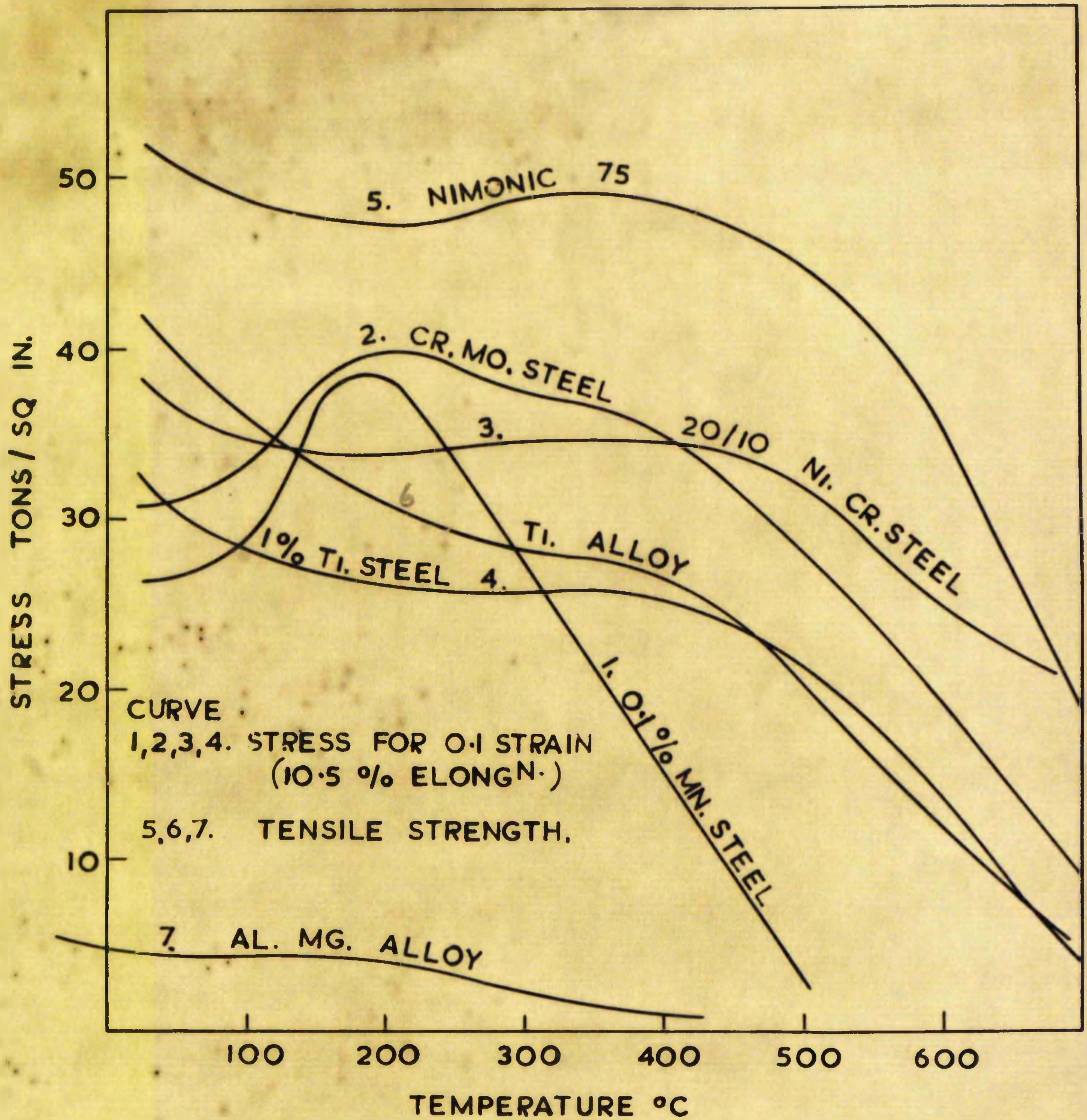
FIG. 2.



IDEAL CREEP CURVES ON PURE METAL.

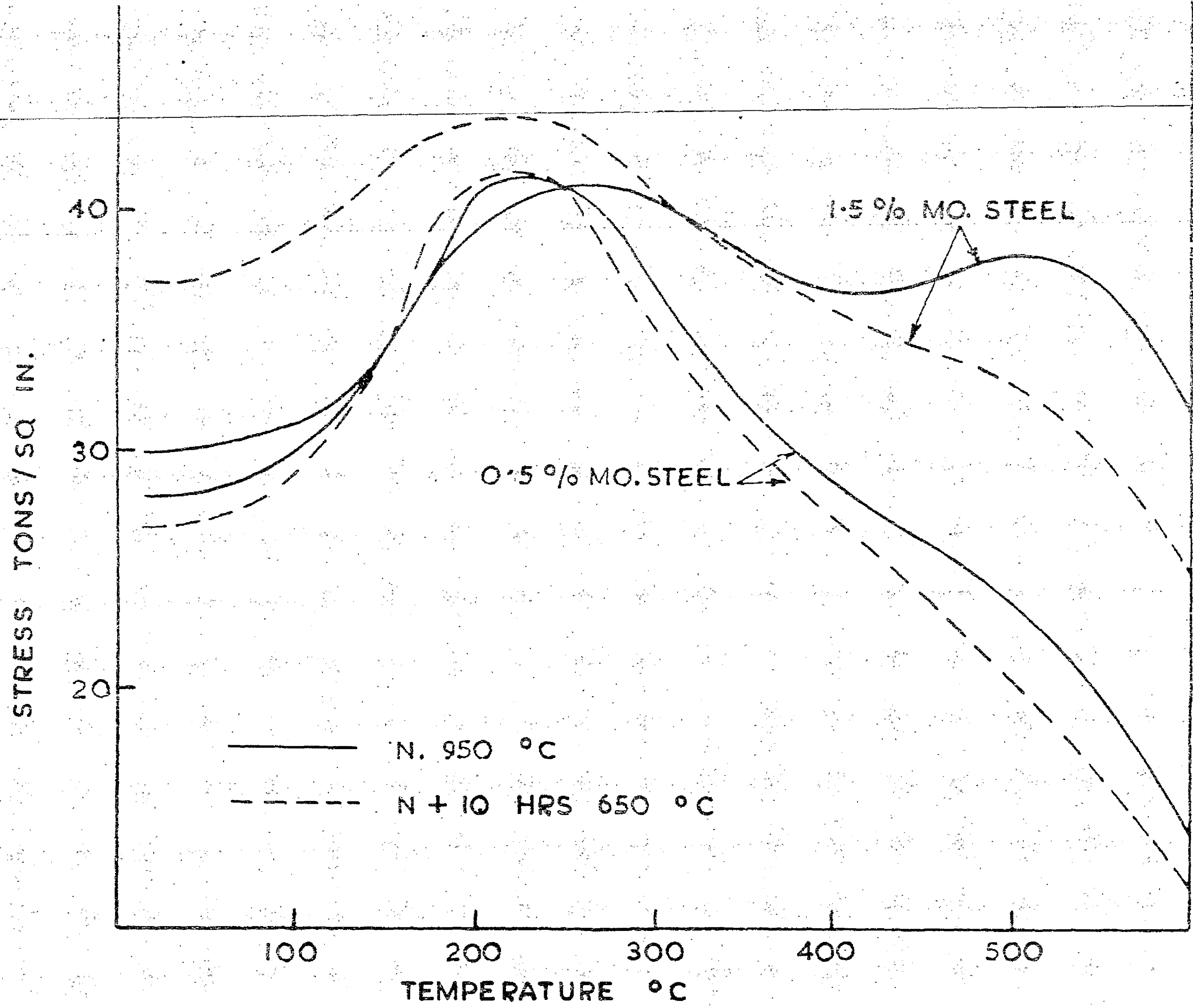
FIG. 3.

FIG. 3.



HIGH TEMPERATURE TENSILE PROPERTIES OF VARIOUS ALLOYS.

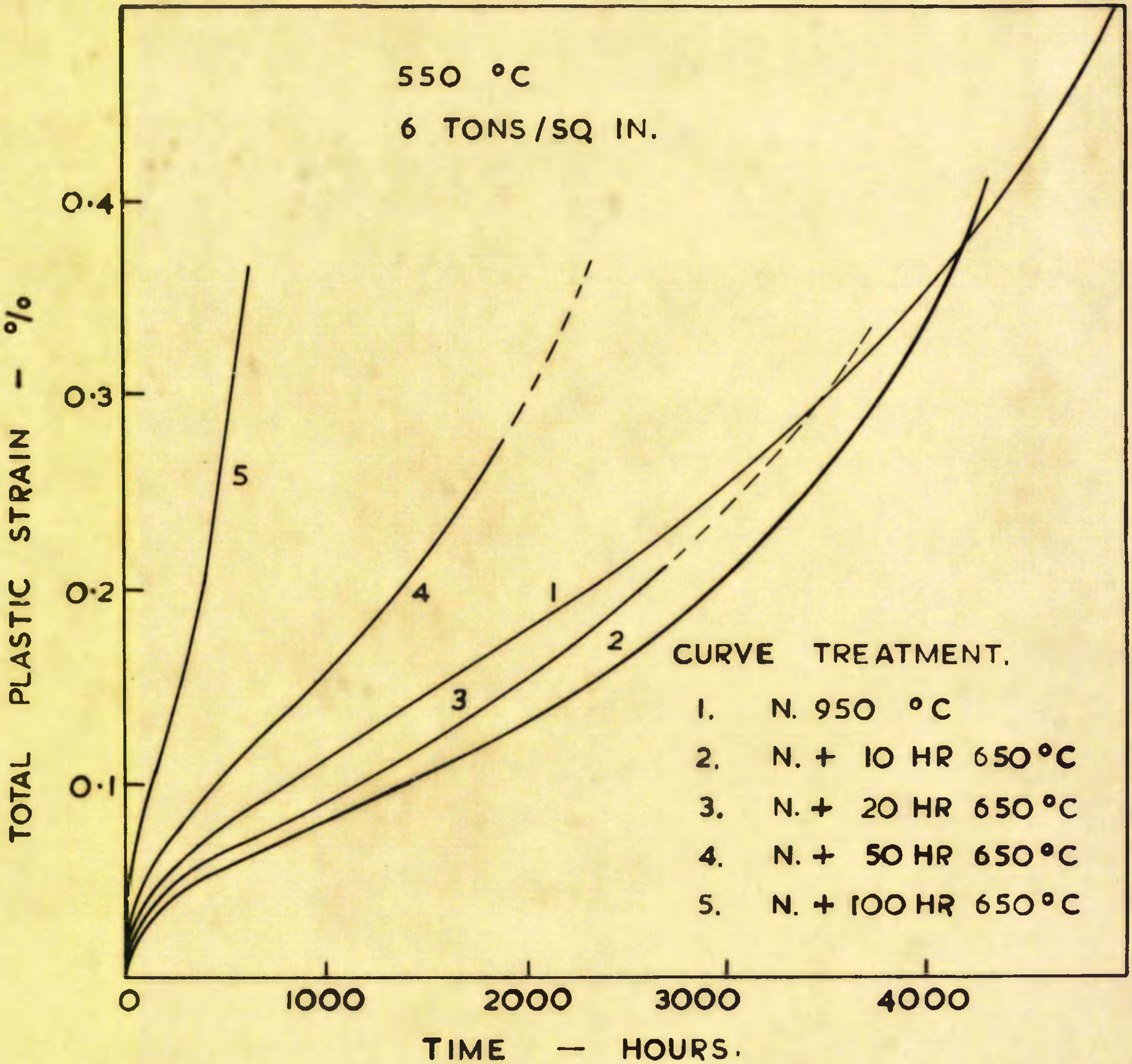
FIG. 4.



STRESS FOR 0·1 STRAIN (0·5 & 1·5 % MO. STEELS)

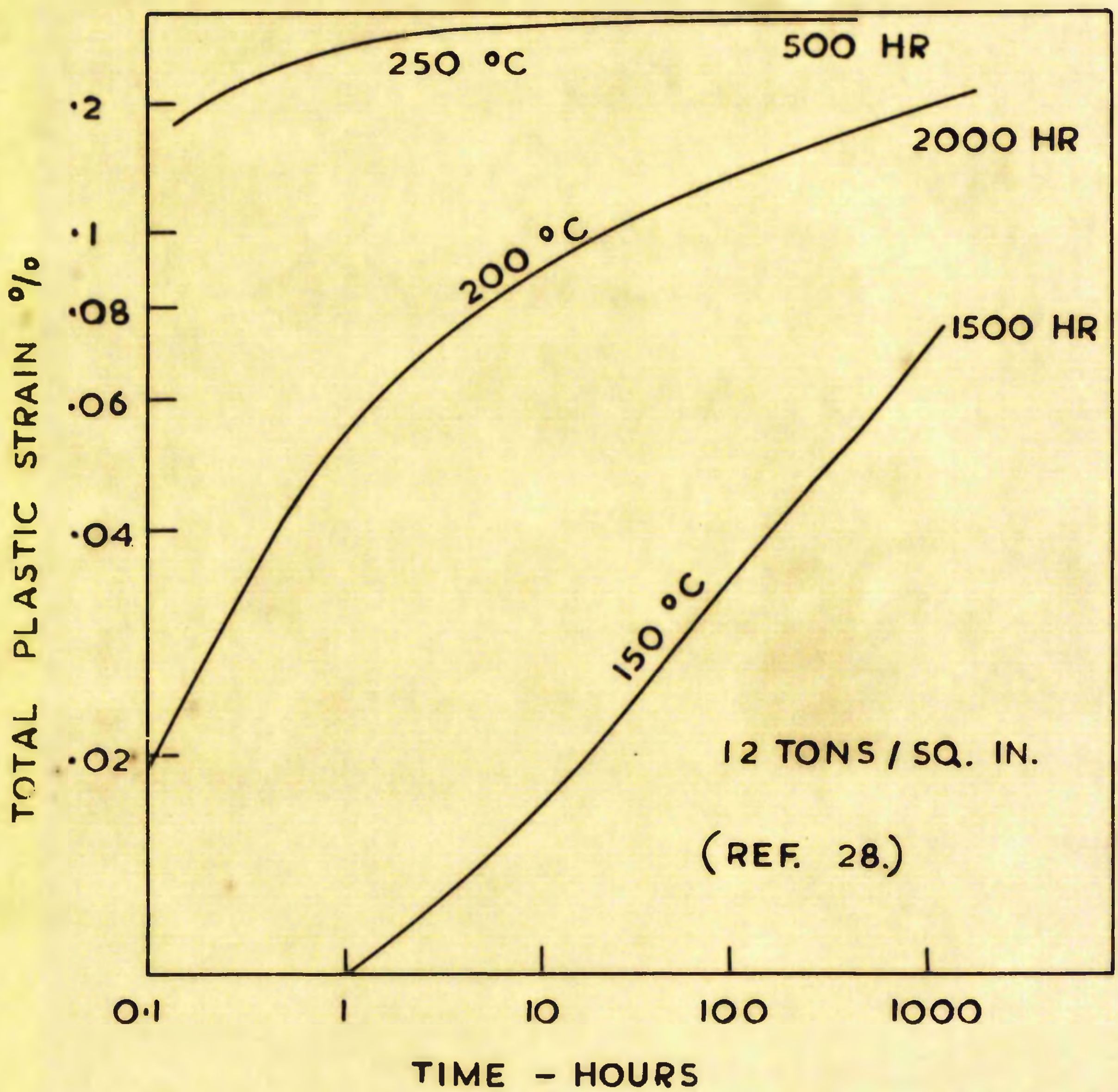
FIG. 5.

FIG. 5.



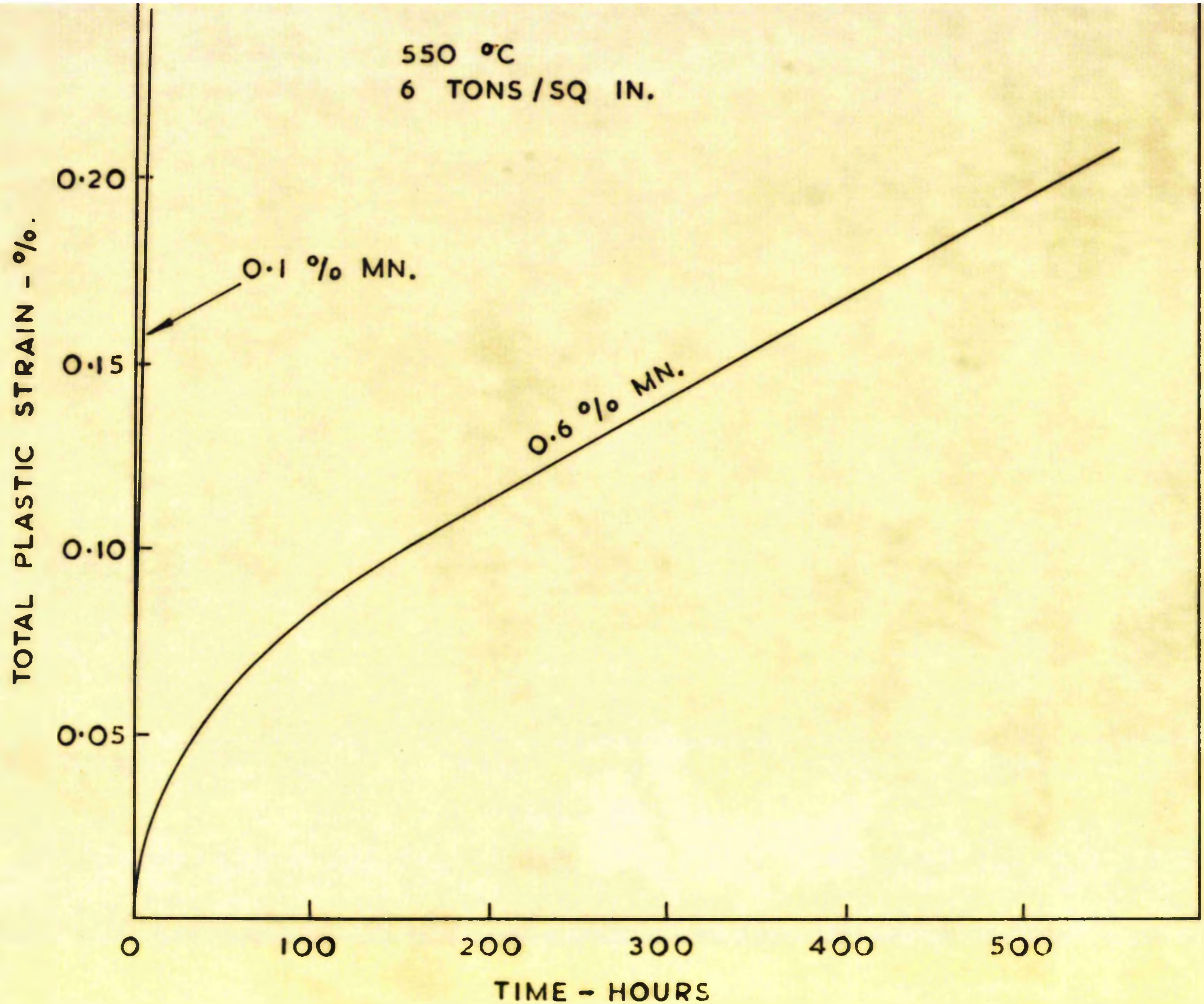
EFFECT OF TEMPERING ON CREEP CURVES FOR 0·5 % MO. STEEL. 17.

FIG. 6.



CREEP CURVES ON 0·5 % MO. STEEL CASTING AT 12 TONS / SQ IN.

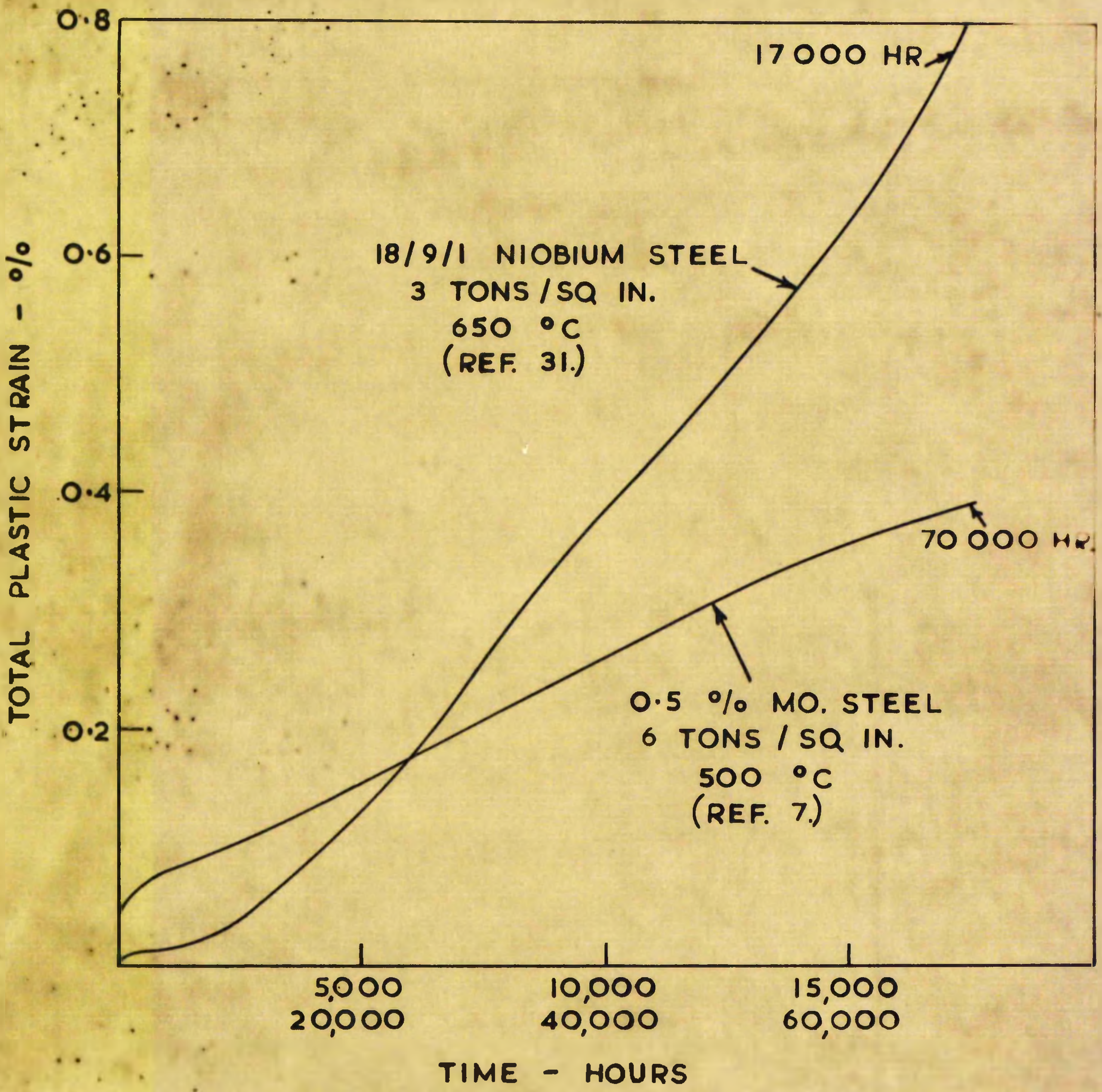
FIG. 7.



CREEP TESTS ON 0·5 % MO. STEEL OF DIFFERENT MN. CONTENT.

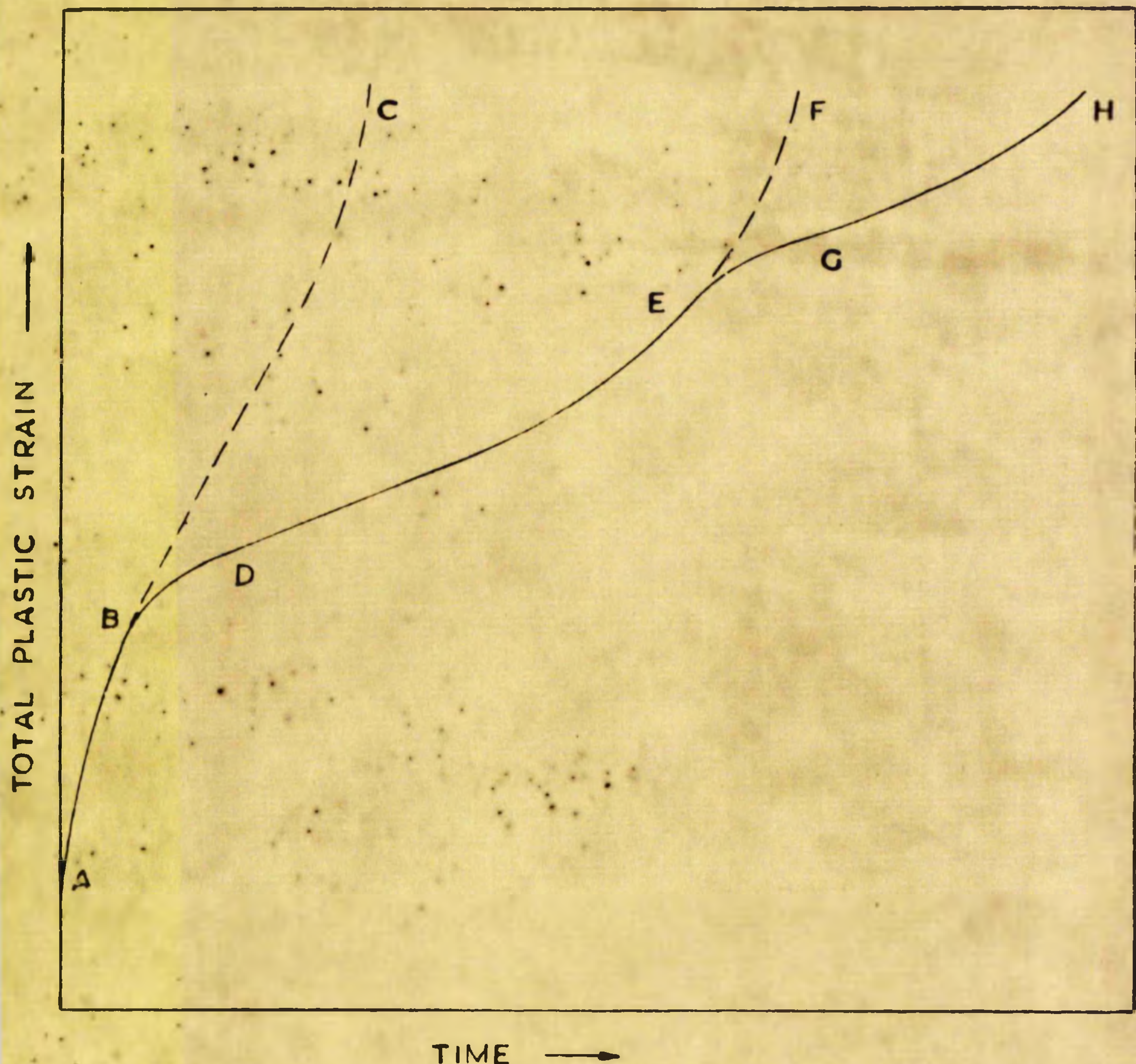
FIG. 8.

FIG. 8.



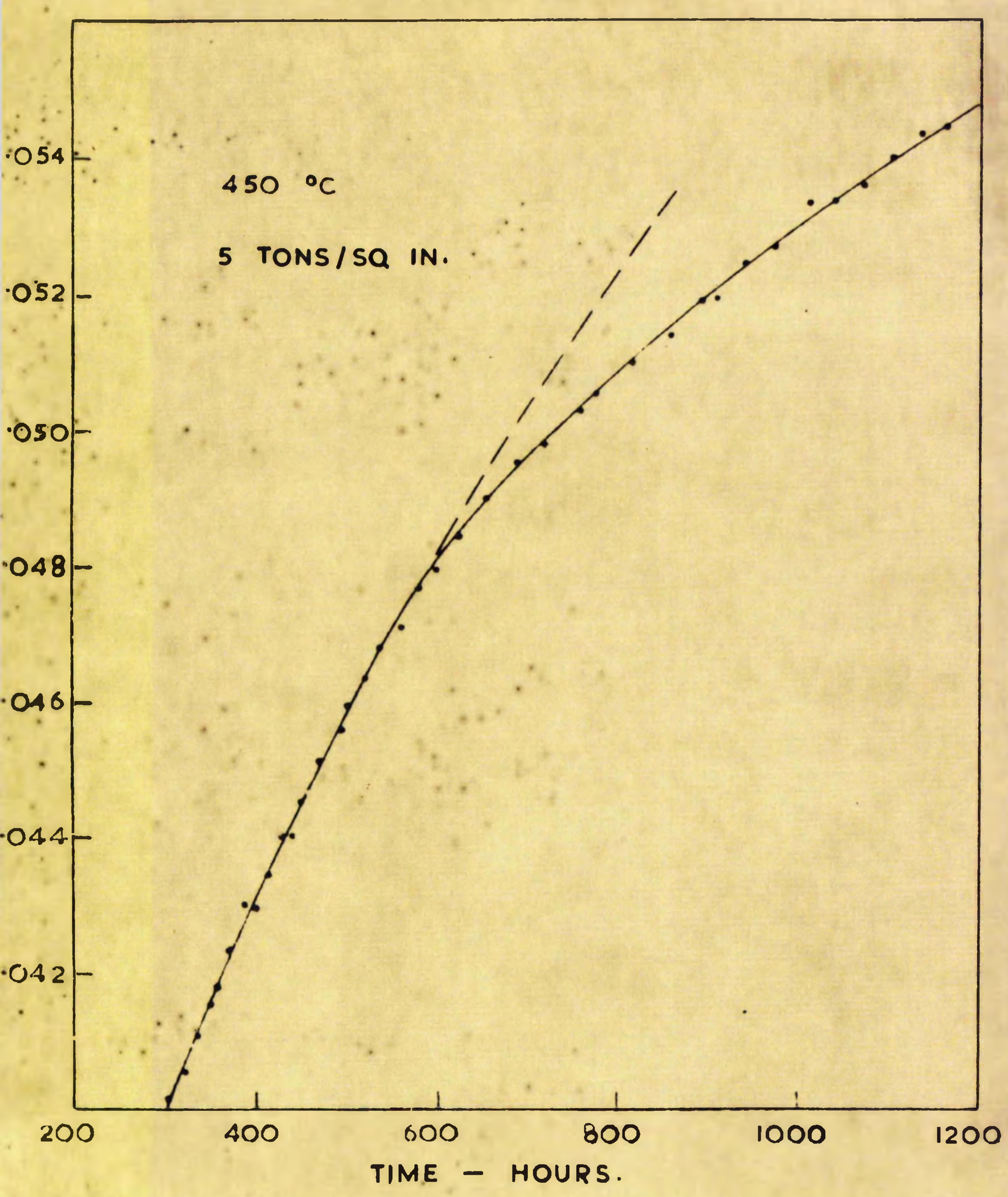
CREEP CURVES SHOWING TRANSITION DURING THIRD STAGE CREEP.

FIG. 9.



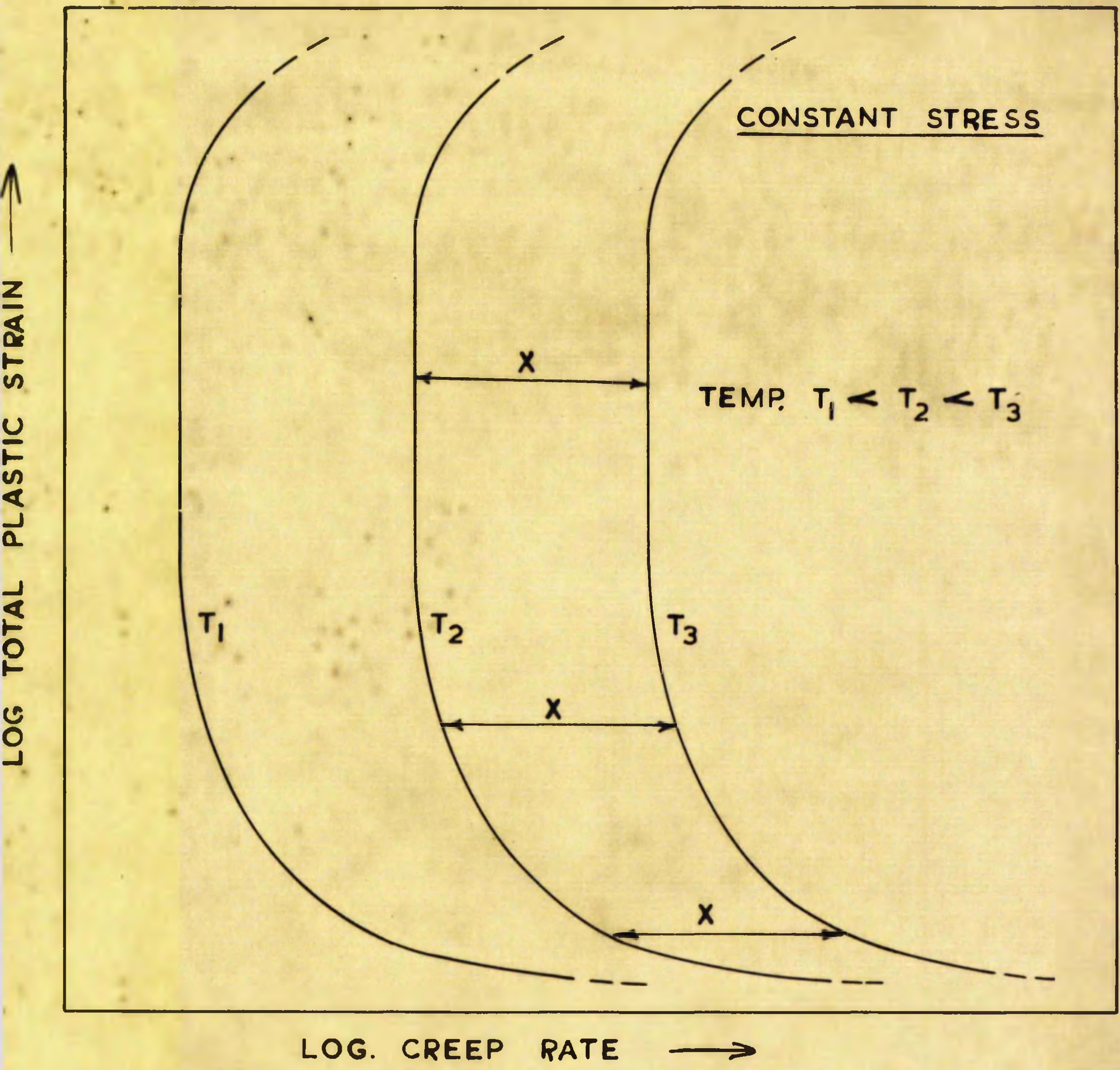
DIAGRAMMATIC CREEP CURVE SHOWING TRANSITIONS.

FIG. 10.



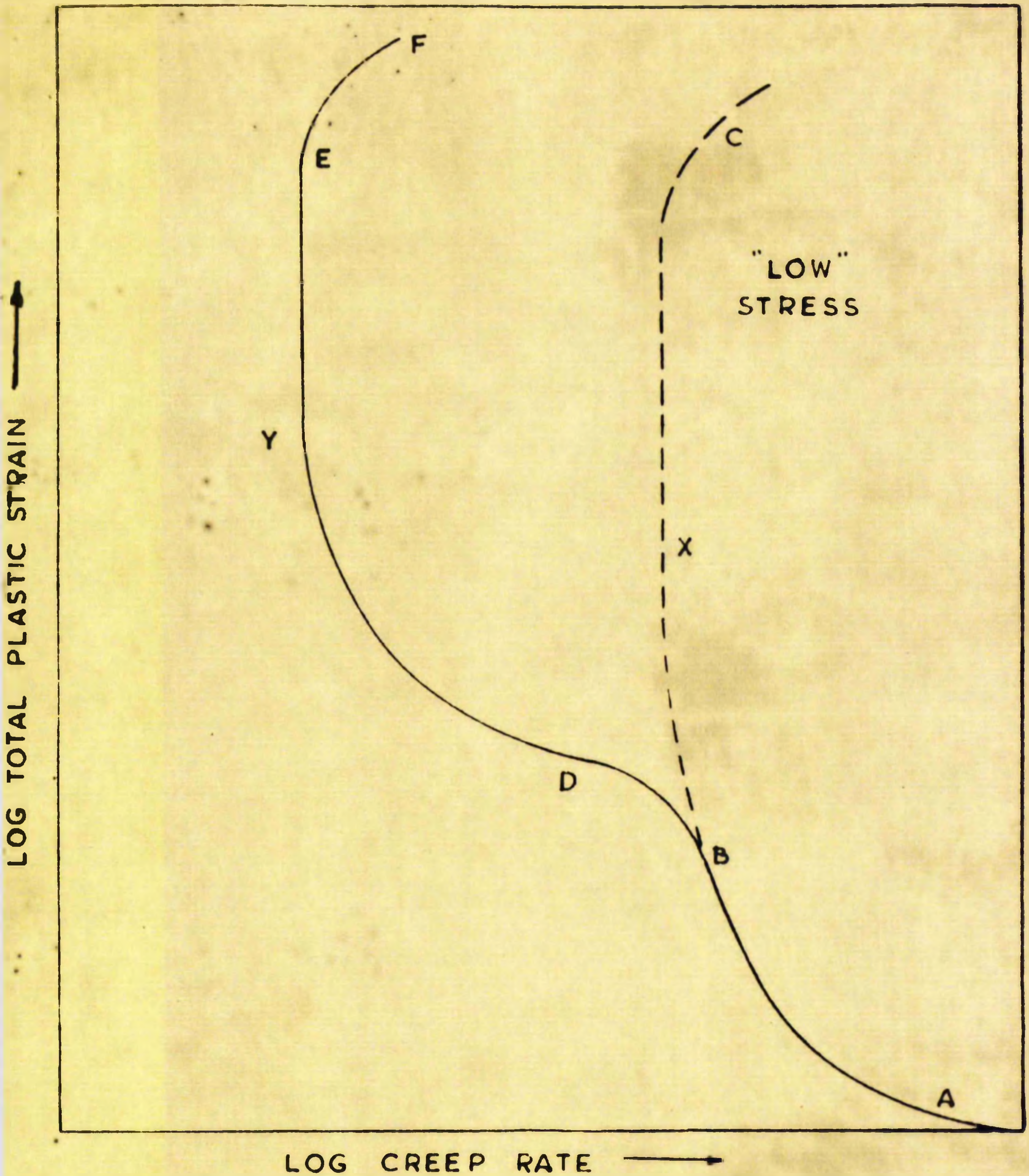
PART CREEP CURVE ON 1 % MN. STEEL.

FIG. 11.



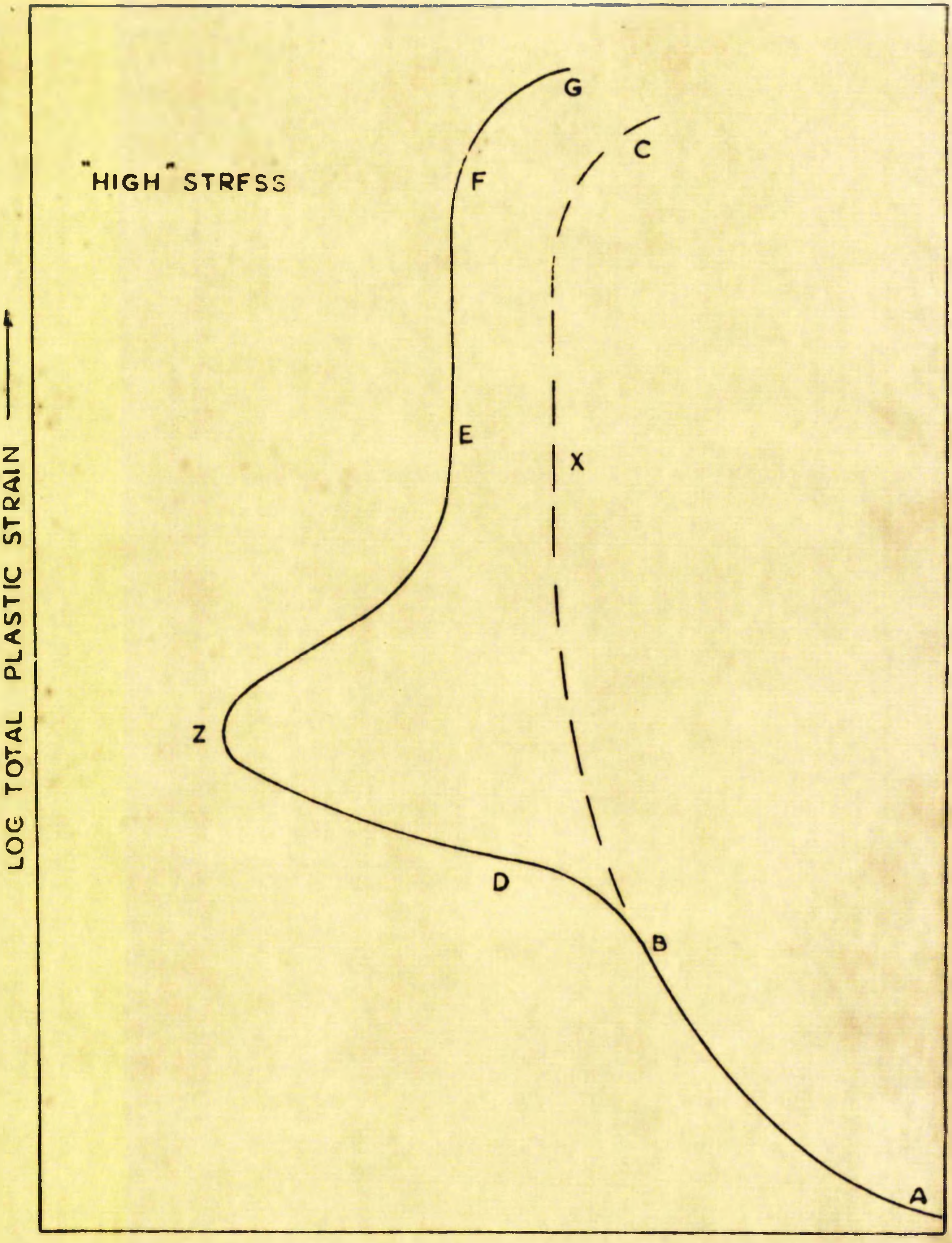
FAMILY OF CONSTANT STRESS STRAIN – RATE CURVES FOR PURE METAL. (DIAGRAMMATIC.)

FIG. 12.



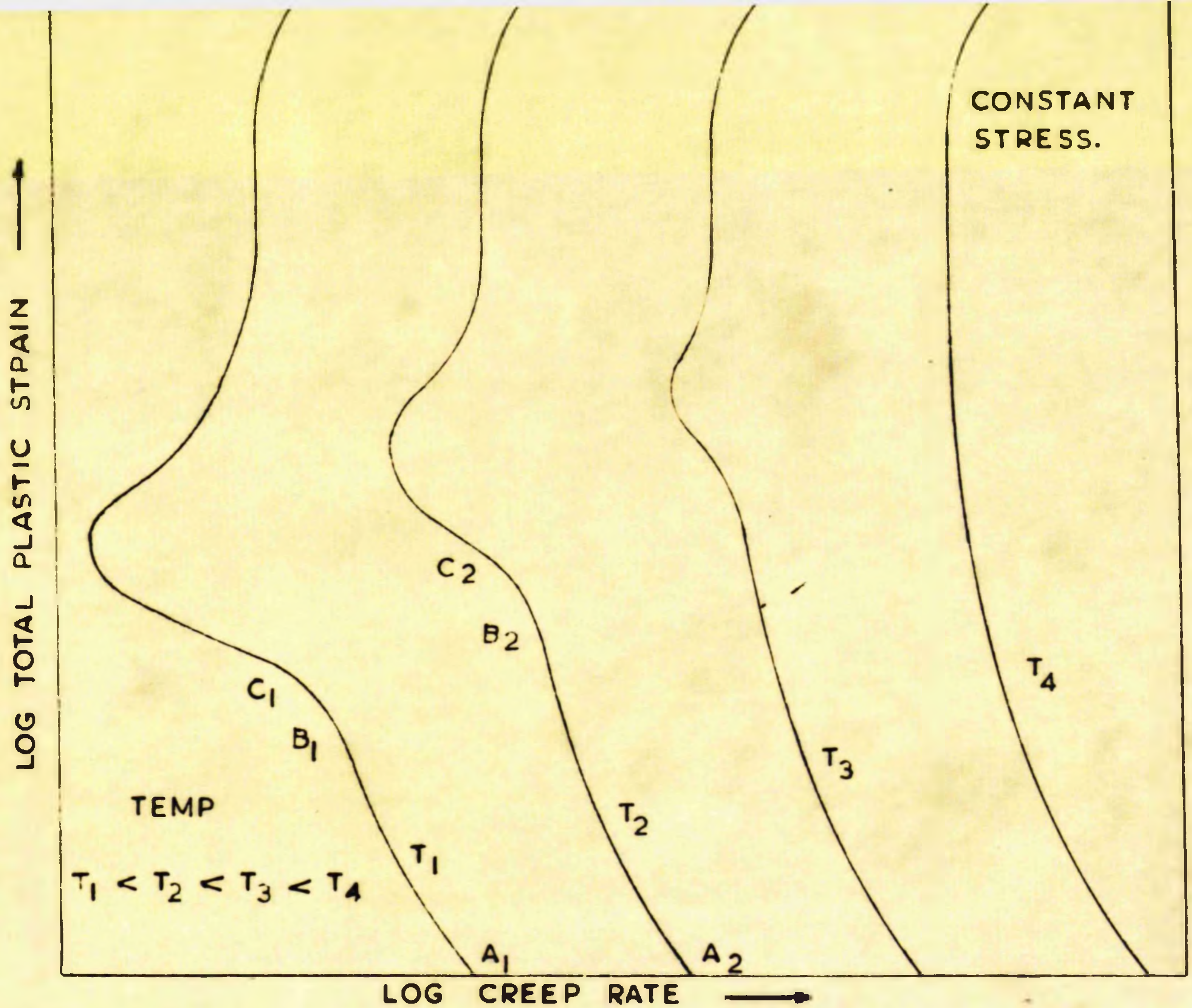
"LOW" STRESS STRAIN - RATE CURVE WITH ONE TRANSITION
(DIAGRAMMATIC)

FIG. 13.



"HIGH" STRESS STRAIN-RATE CURVE WITH ONE TRANSITION
(DIAGRAMMATIC)

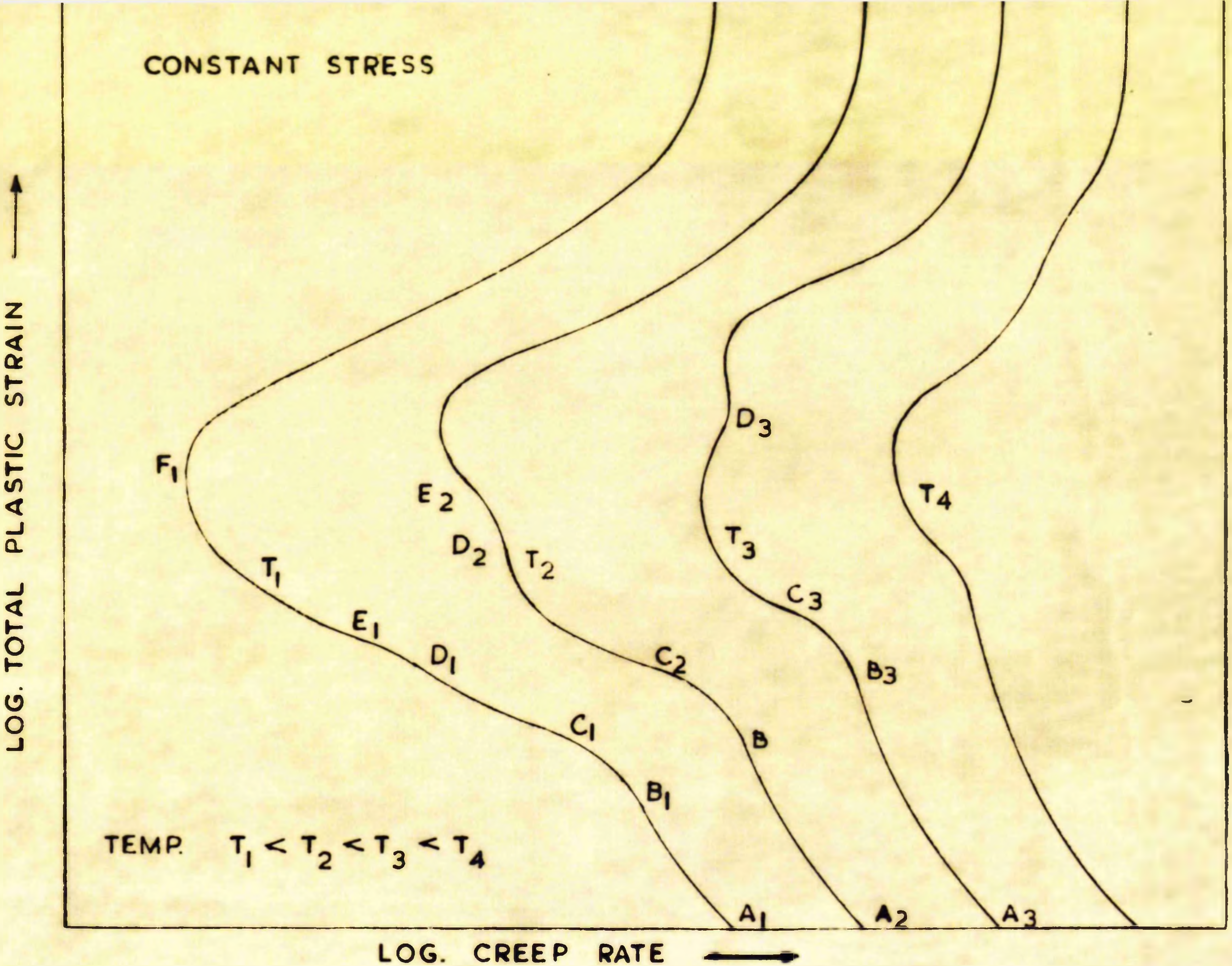
FIG. 14.



FAMILY OF CONSTANT STRESS STRAIN - RATE CURVES
WITH ONE TRANSITION (DIAGRAMMATIC.)

FIG. 15.

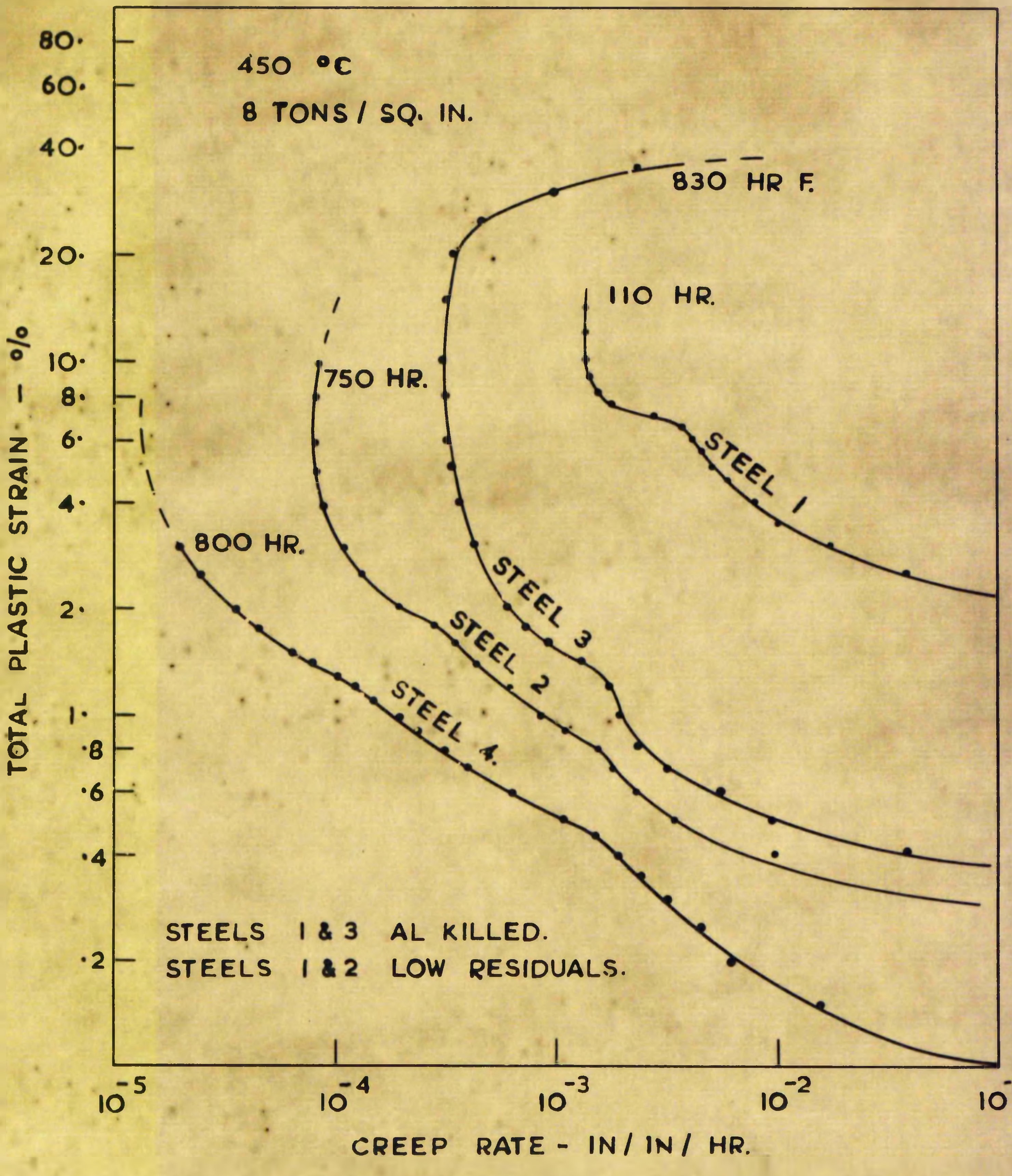
FIG. 1



FAMILY OF CONSTANT STRESS STRAIN - RATE CURVES WITH TWO TRANSITIONS (DIAGRAMMATIC.)

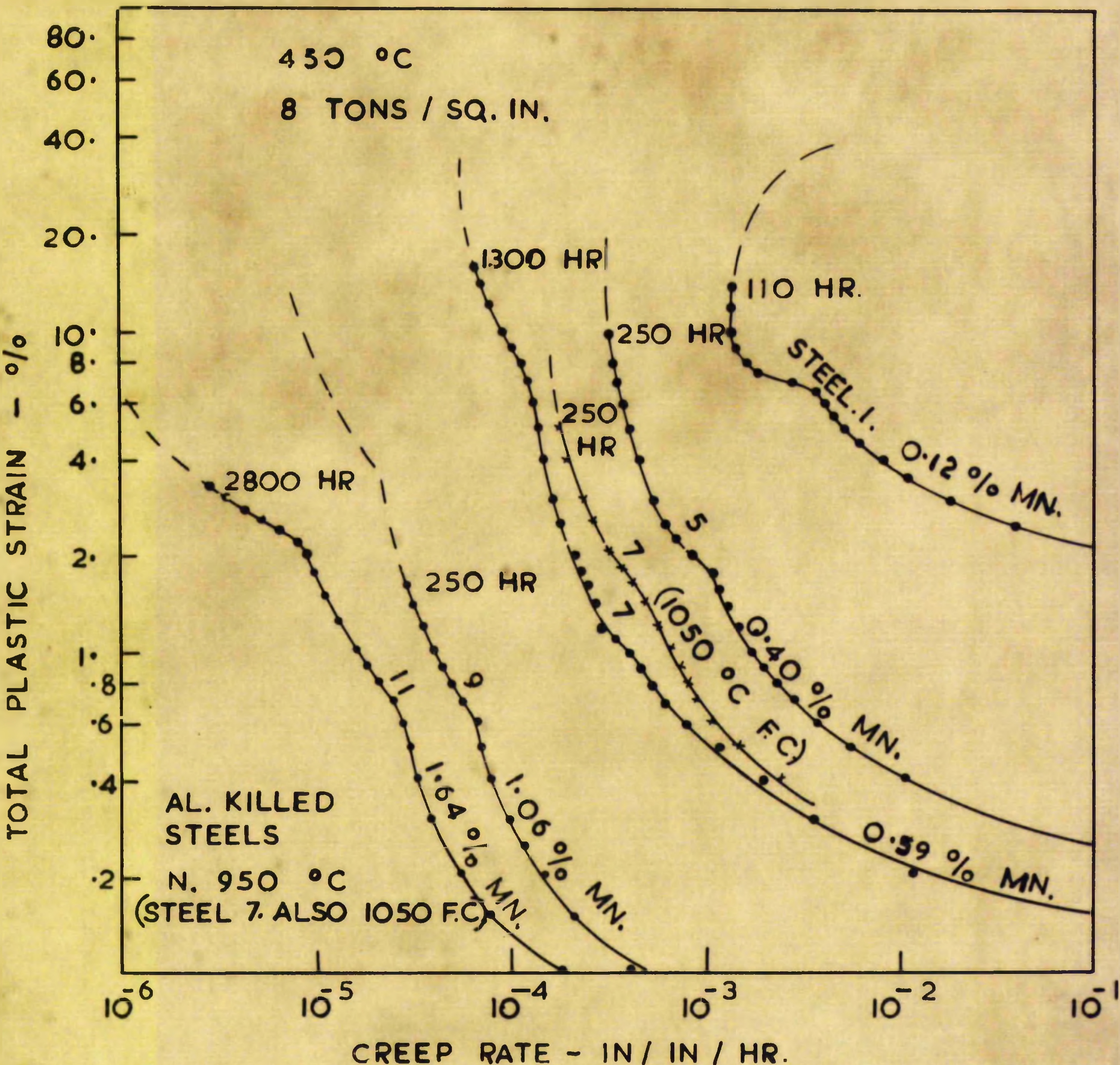
FIG 16.

FIG. 16.



STRAIN - RATE CURVES ON LOW MN. STEELS.

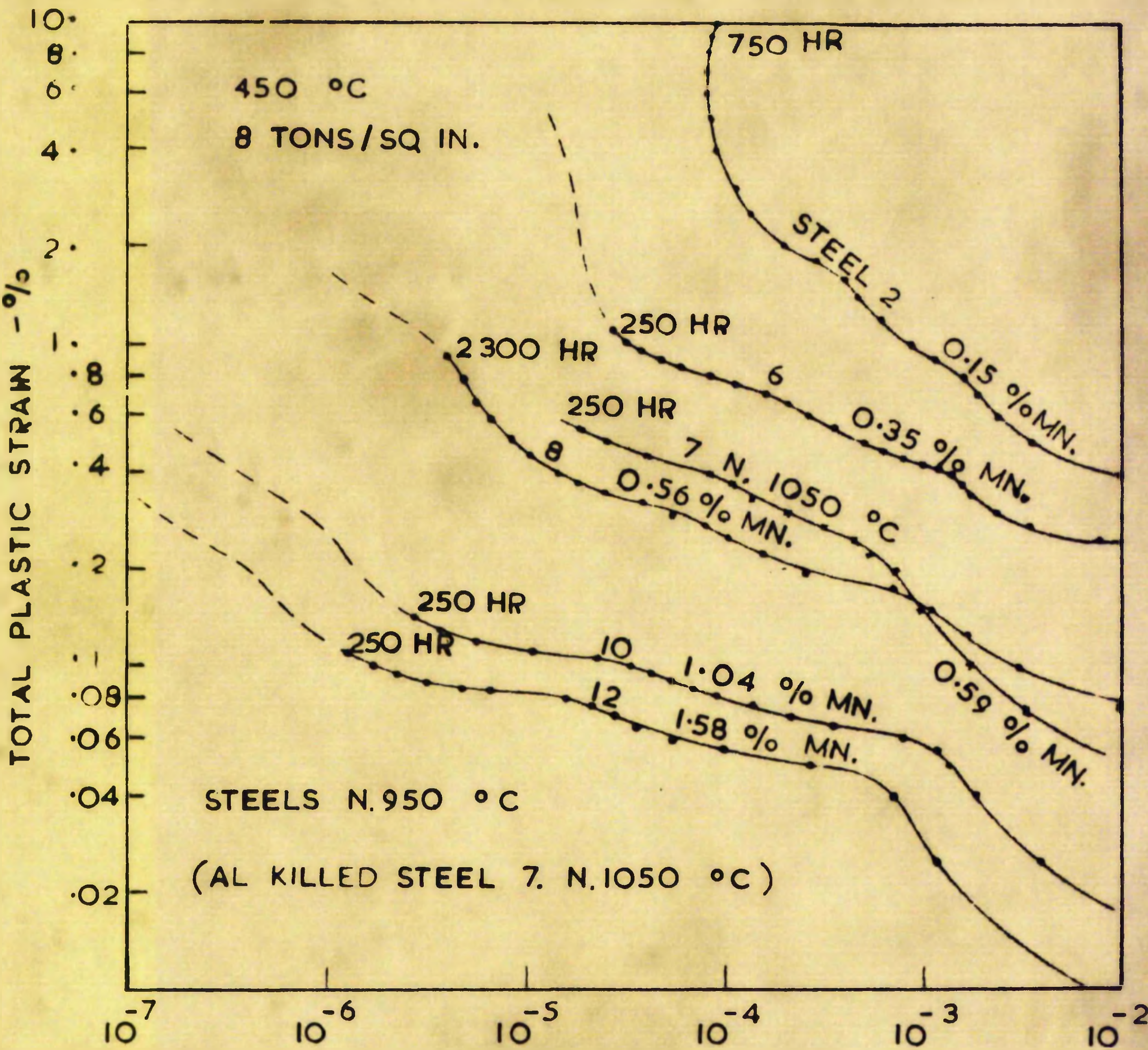
FIG. 17.



STRAIN RATE CURVES ON AL. KILLED STEELS.

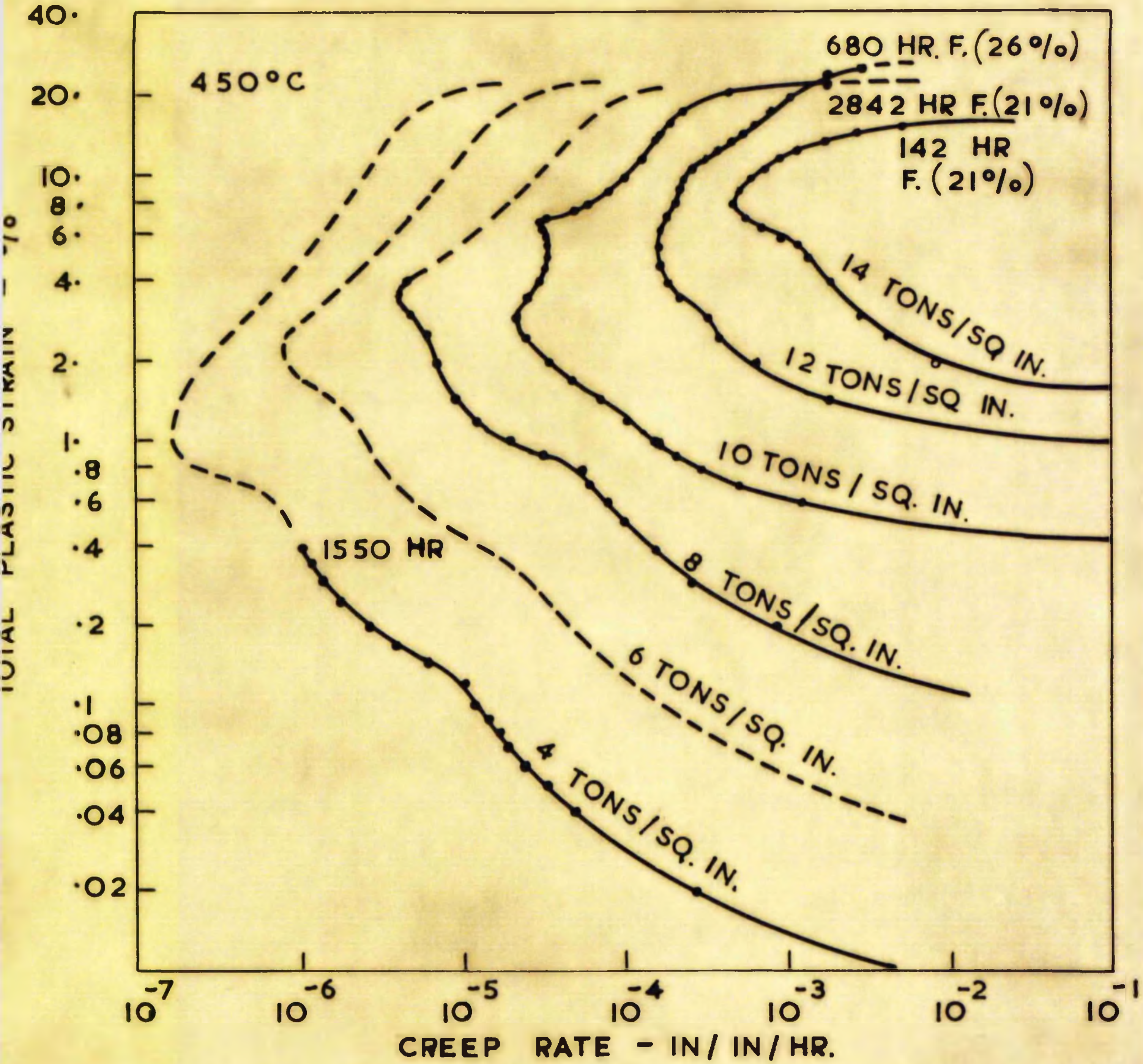
WITH VARIOUS MN. CONTENTS.

FIG. 18.



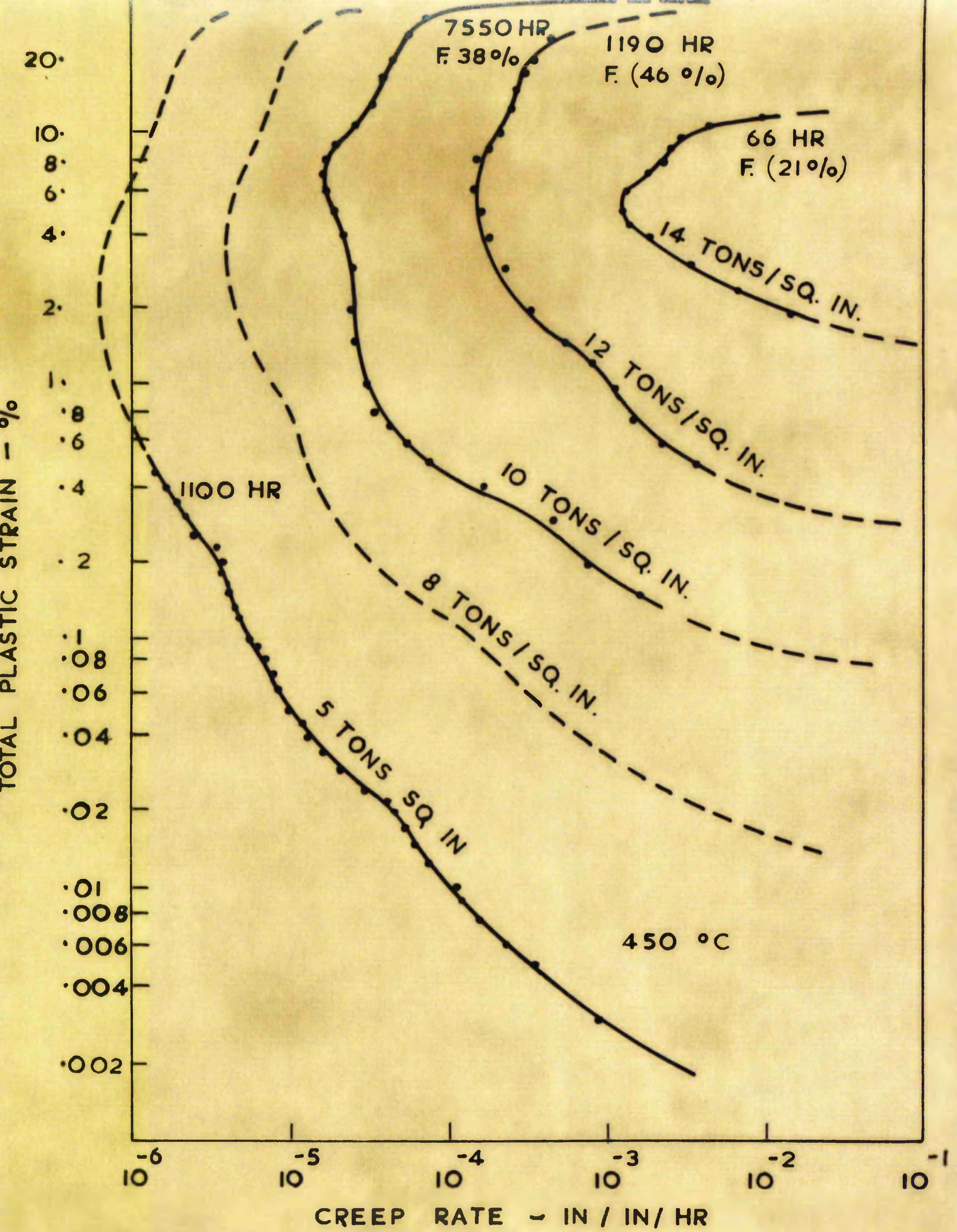
STRAIN - RATE CURVES ON STEELS
WITH VARIOUS MN. CONTENTS.

FIG. 19.



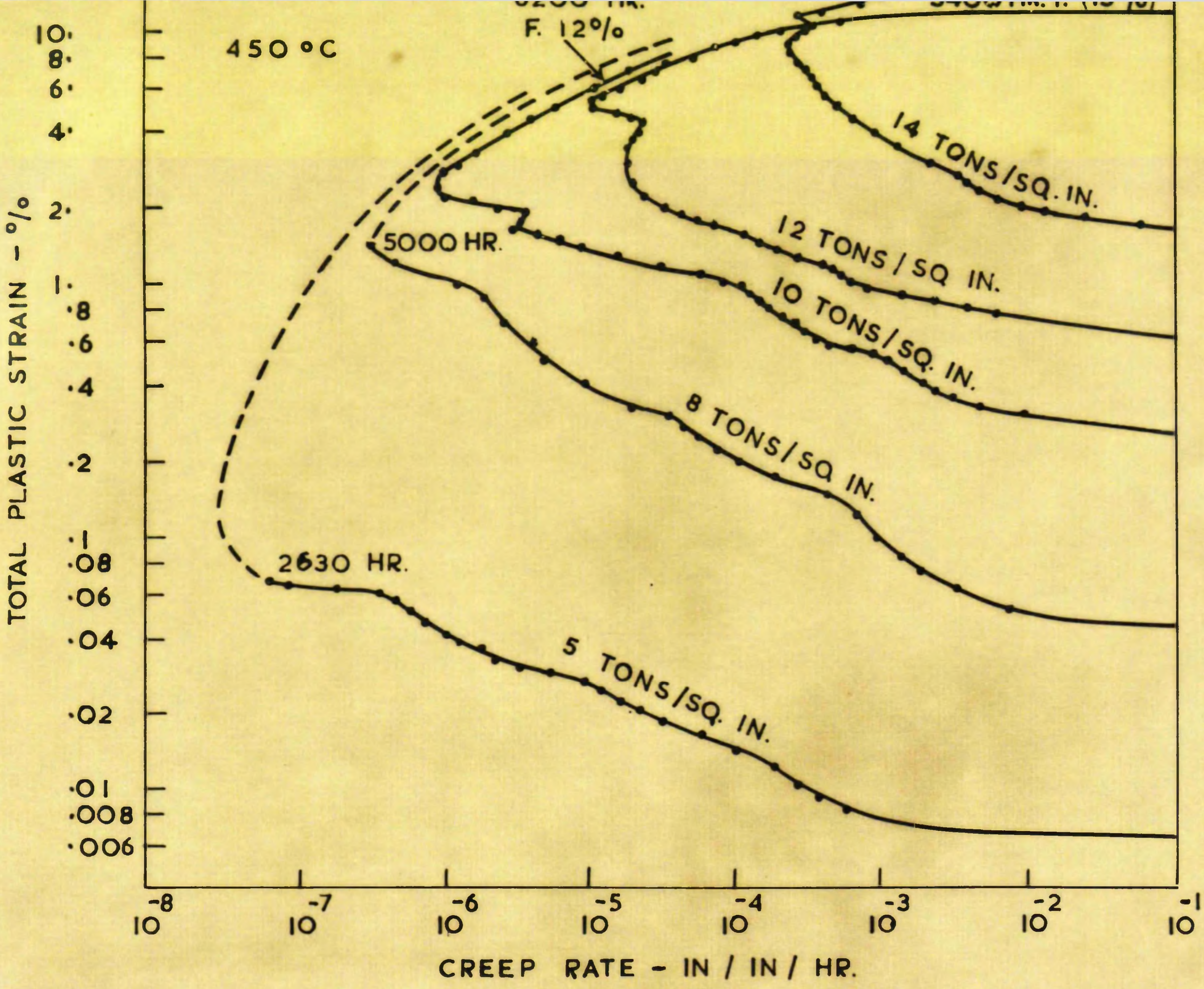
FAMILY OF STRAIN-RATE CURVES 0·5% MN. STEEL 13. (AL KILLED)

FIG. 20.



FAMILY OF STRAIN RATE CURVES 1 % MN. STEEL 15. (AL. KILLED)

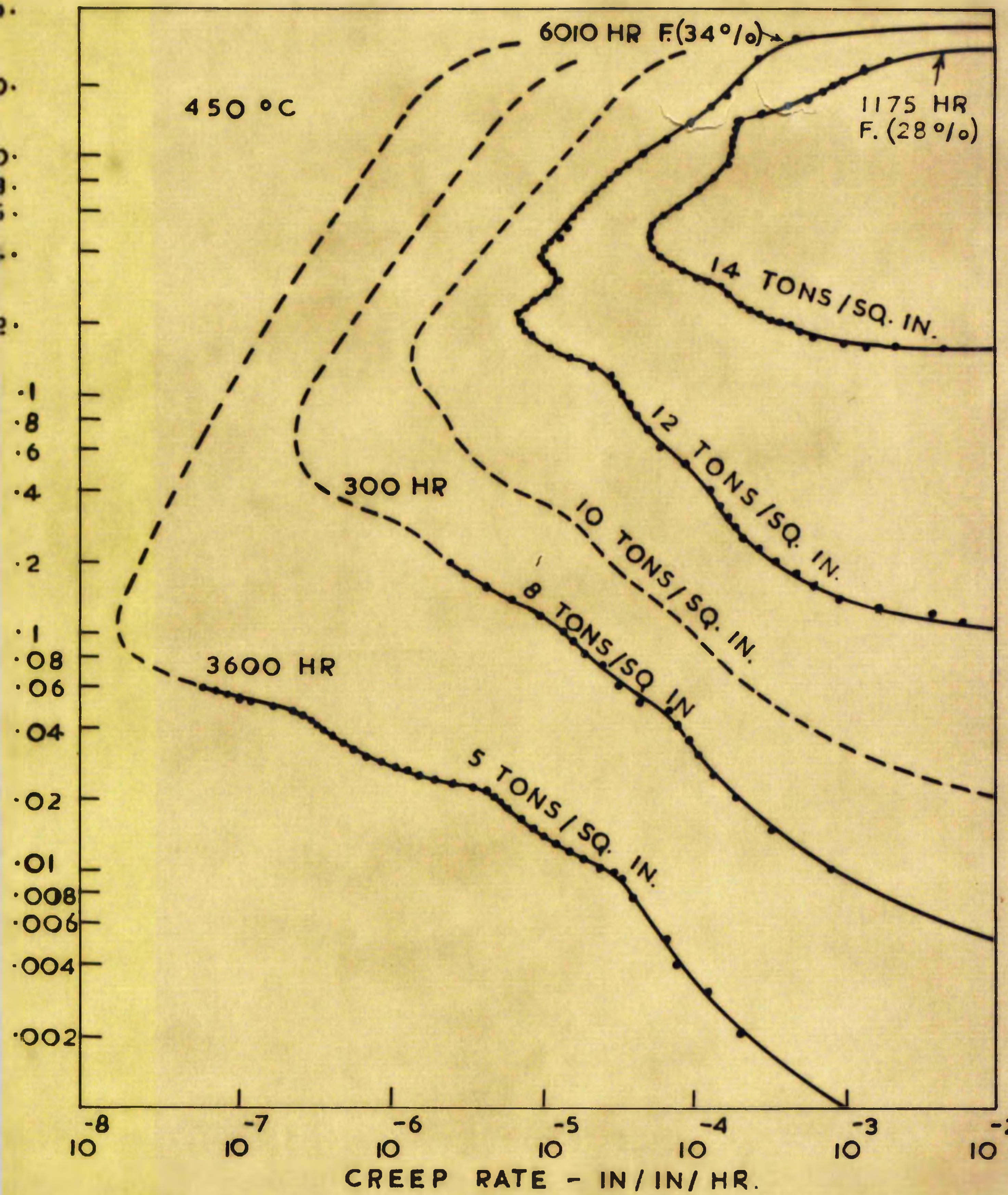
FIG. 21



FAMILY OF STRAIN-RATE CURVES 0·5% MN. STEEL. 14.

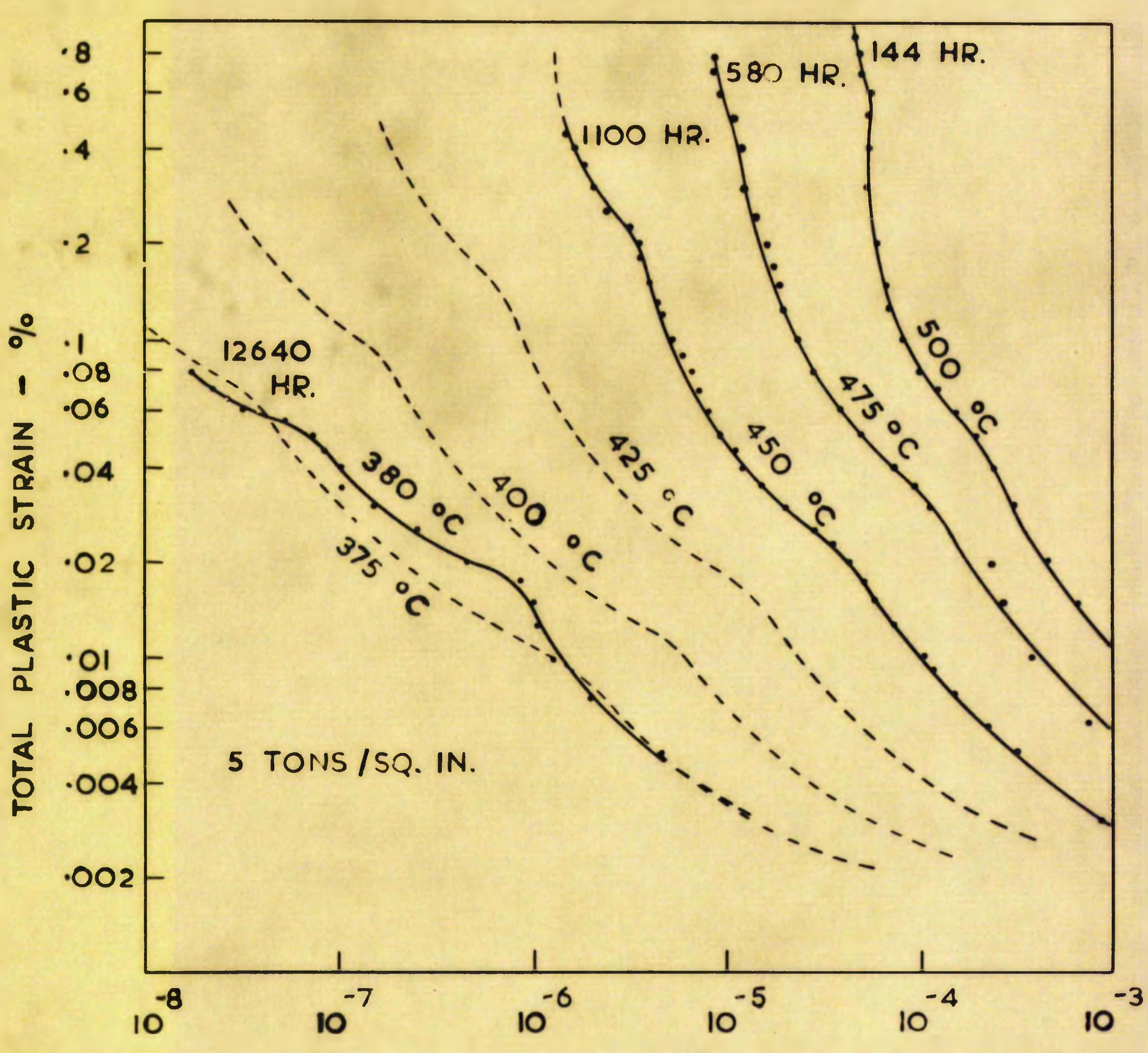
FIG. 2

FIG. 22.



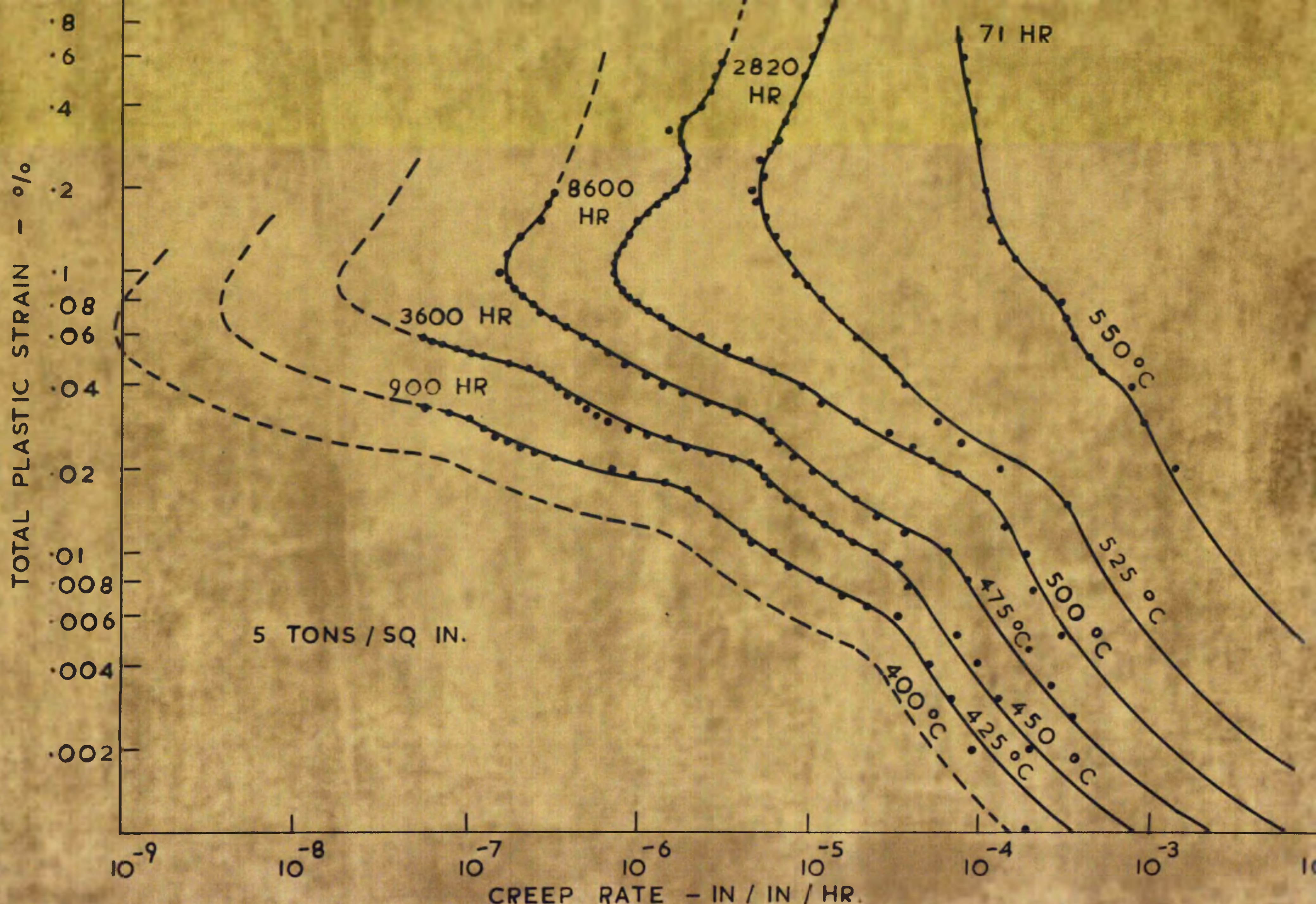
FAMILY OF STRAIN - RATE CURVES 1% MN. STEEL. 16.

FIG. 23.



FAMILY OF STRAIN - RATE CURVES AT 5 TONS / SQ. IN. ON 1% MN. STEEL. 15. (AL. KILLED)

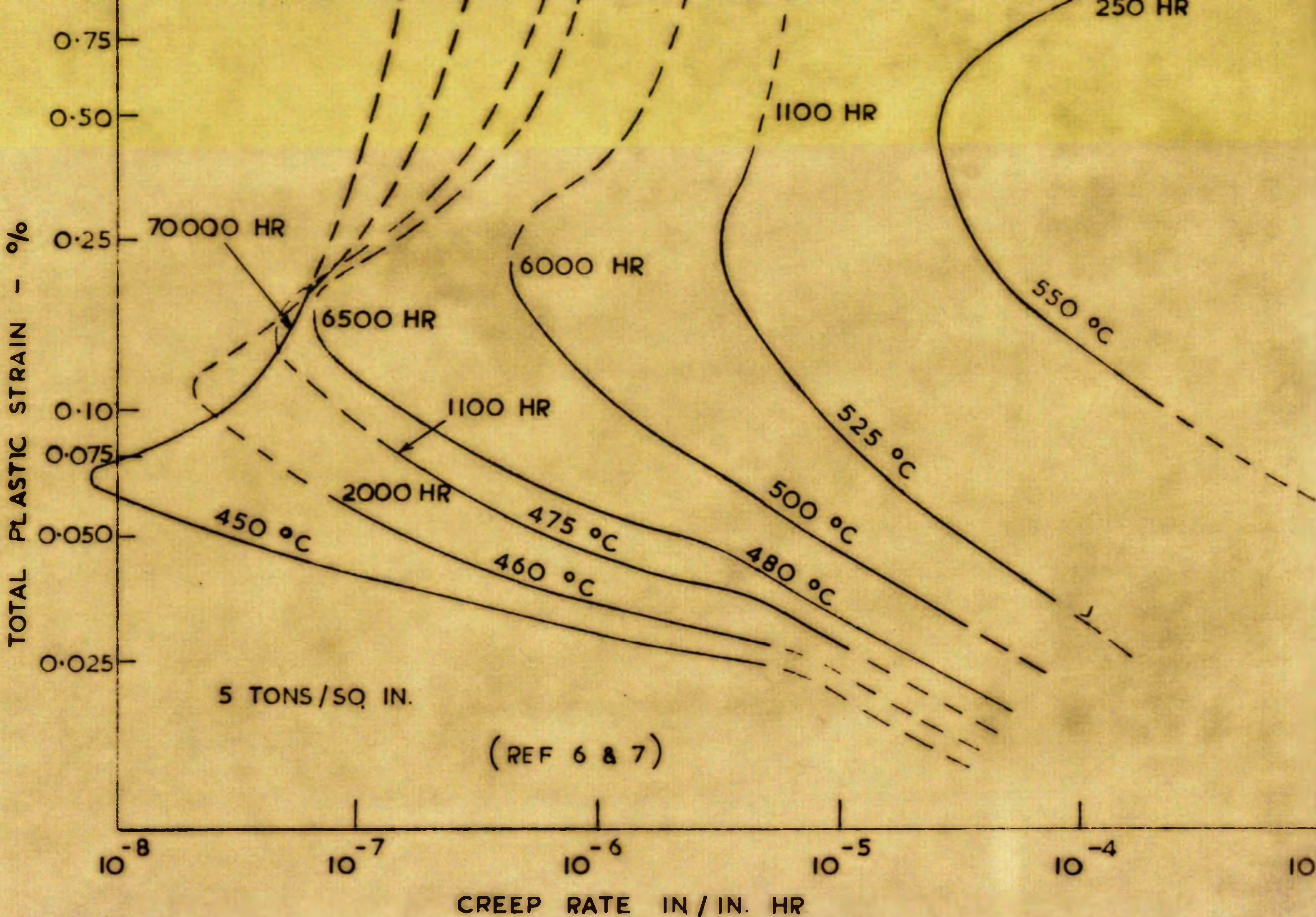
FIG. 24.



FAMILY OF STRAIN - RATE CURVES AT 5 TONS / SQ. IN. ON 1 % MN. STEEL. 16.

FIG. 25.

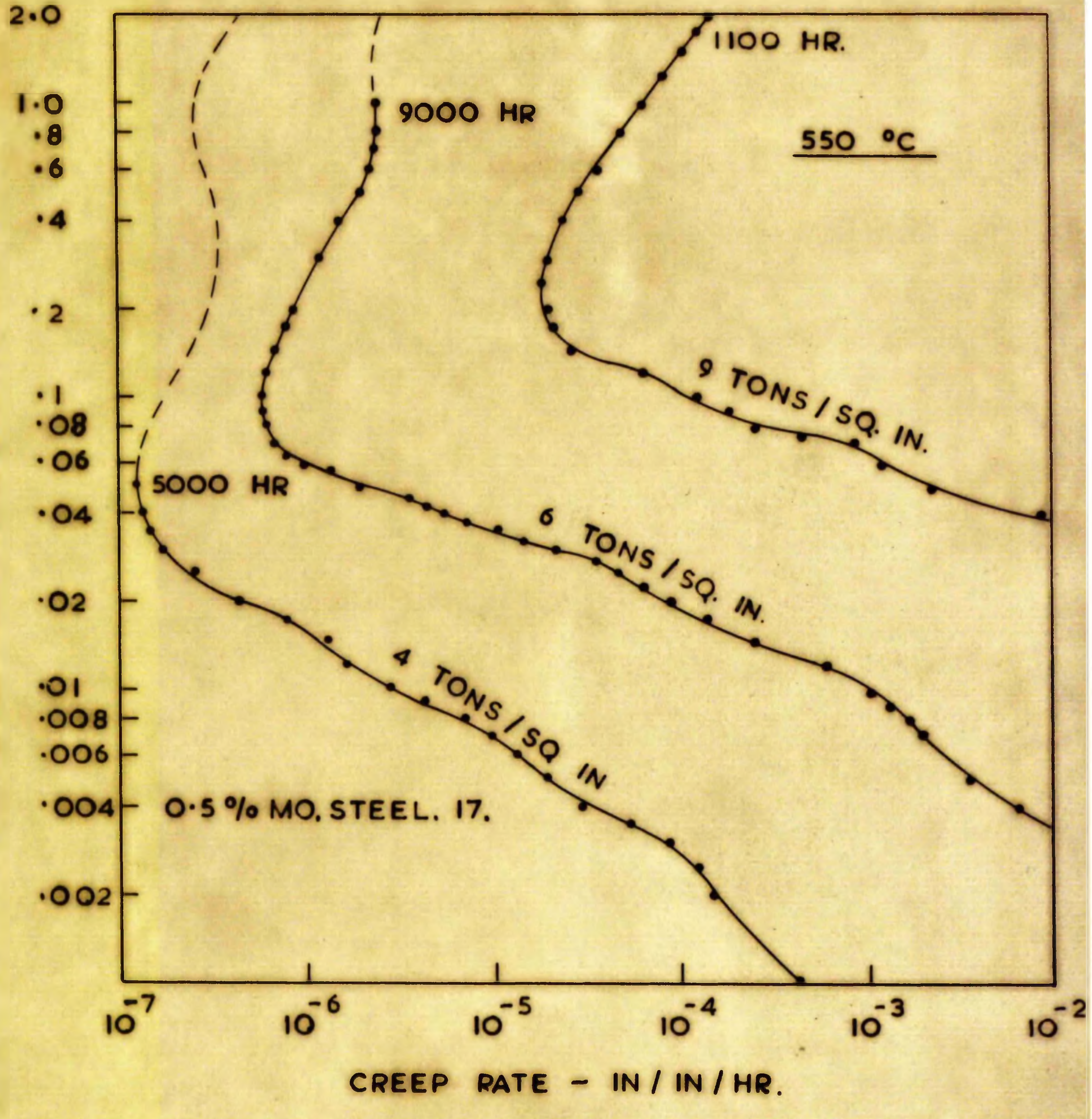
FIG. 25



FAMILY OF STRAIN-RATE CURVES FOR CARBON STEEL AT 5 TONS/SQ. IN.

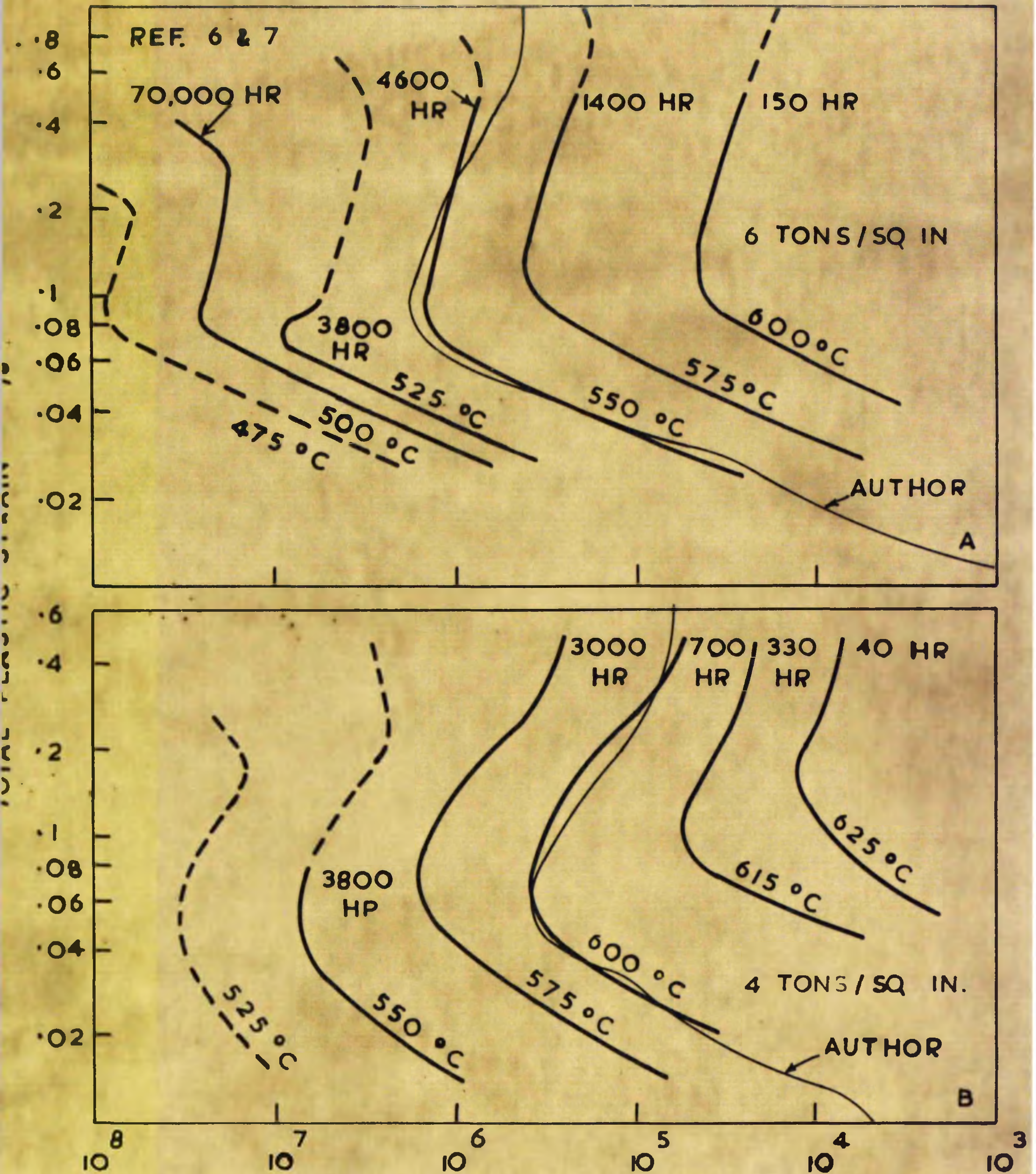
FIG. 2

FIG 26



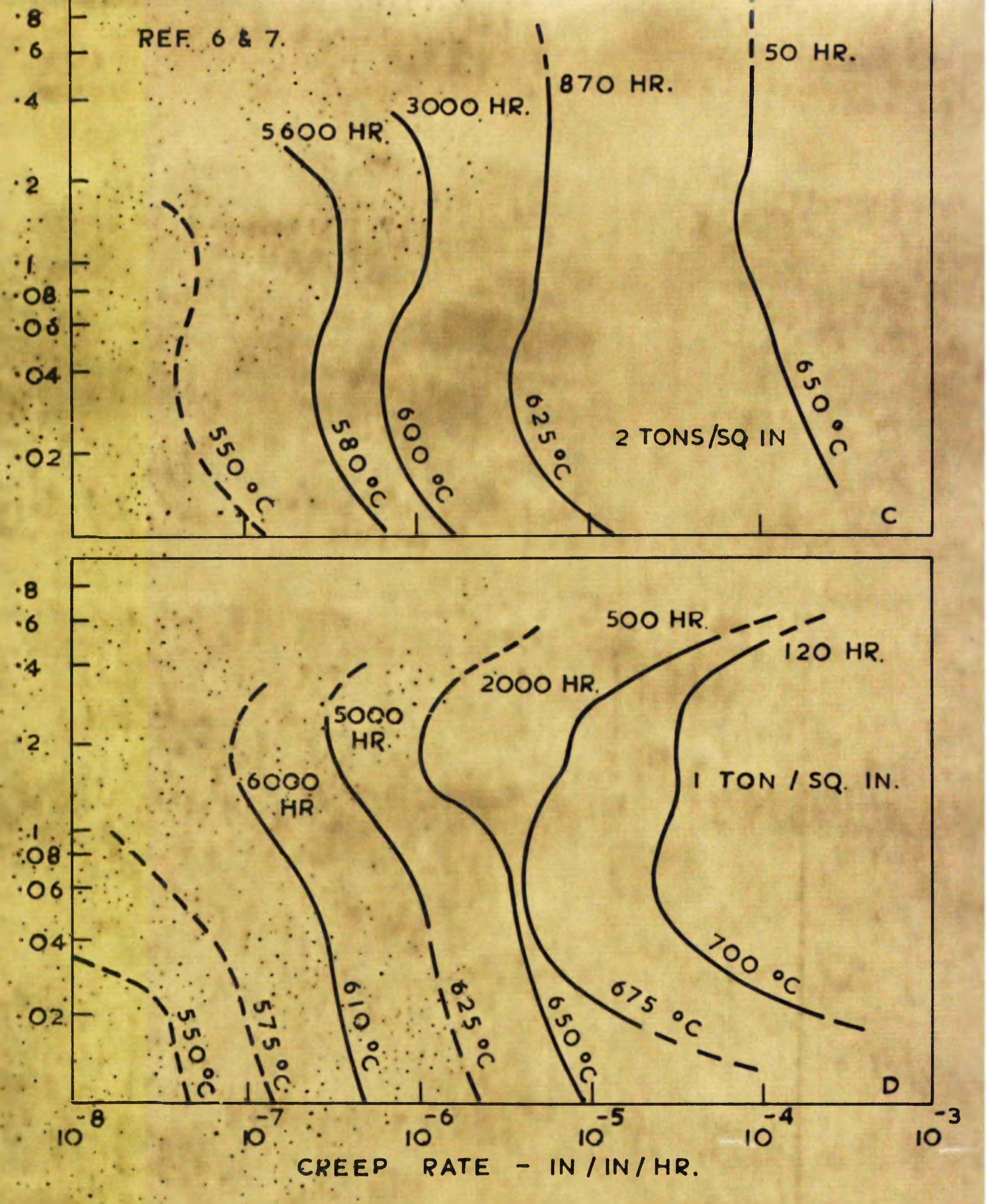
STRAIN – RATE CURVES ON 0·5 % MO. STEEL. 17.

FIG. 27.



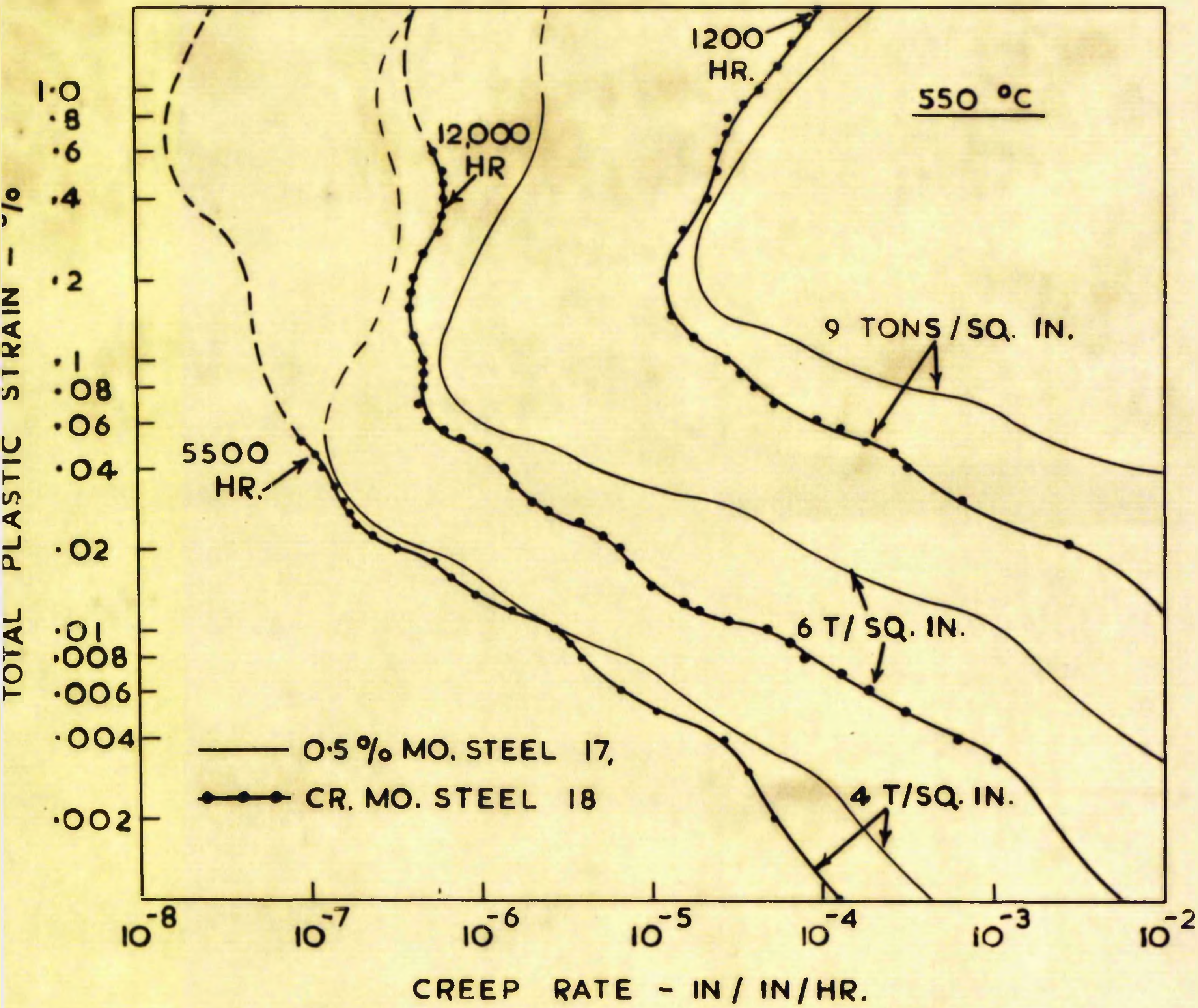
FAMILIES OF STRAIN-RATE CURVES ON 0·5% MO. STEEL.
(FIGS. C. & D. OVERLEAF)

FIG 28. A & B



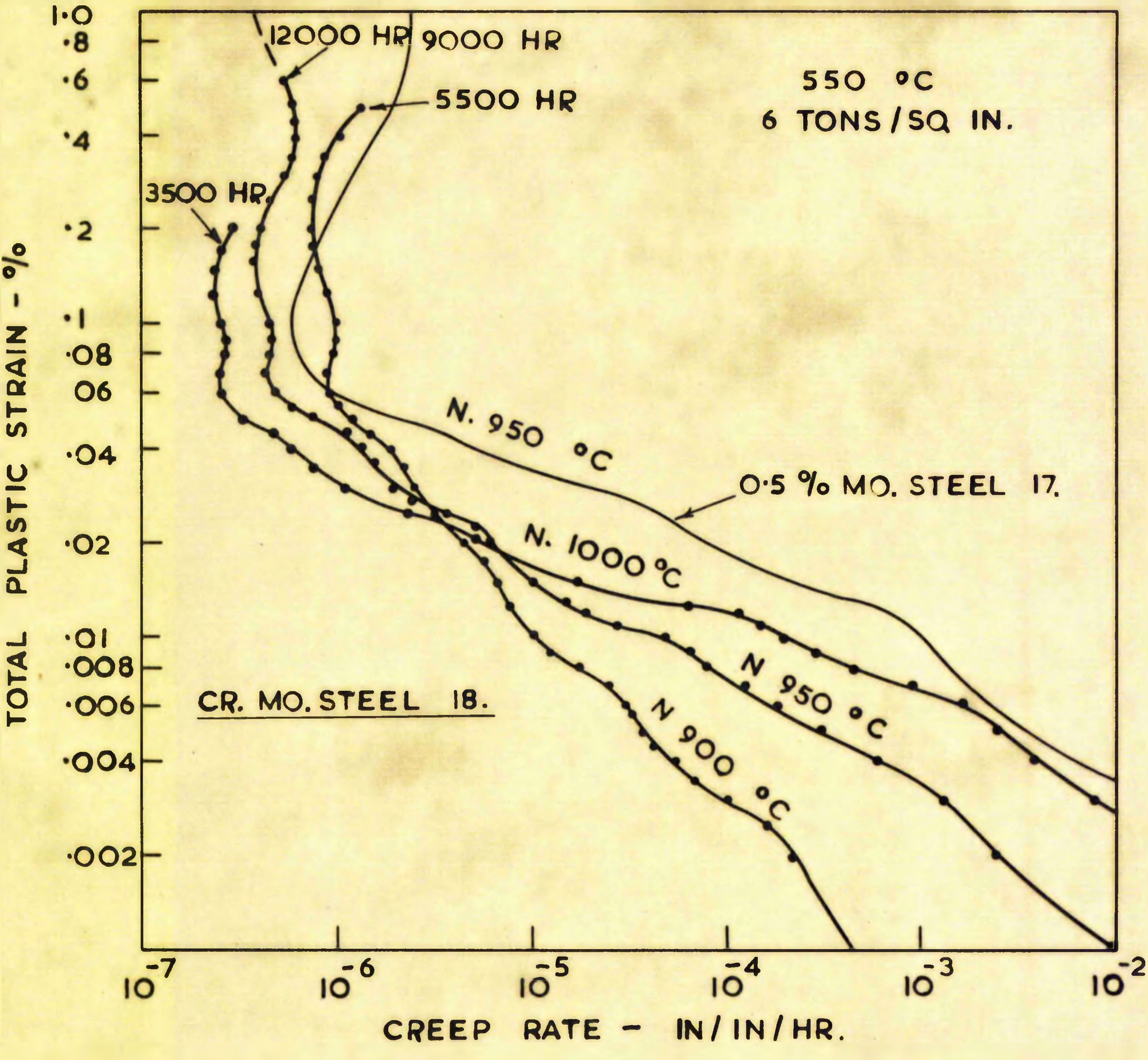
FAMILIES OF STRAIN - RATE CURVES ON 0·5% MO. STEEL.

FIG. 28 C. & D.



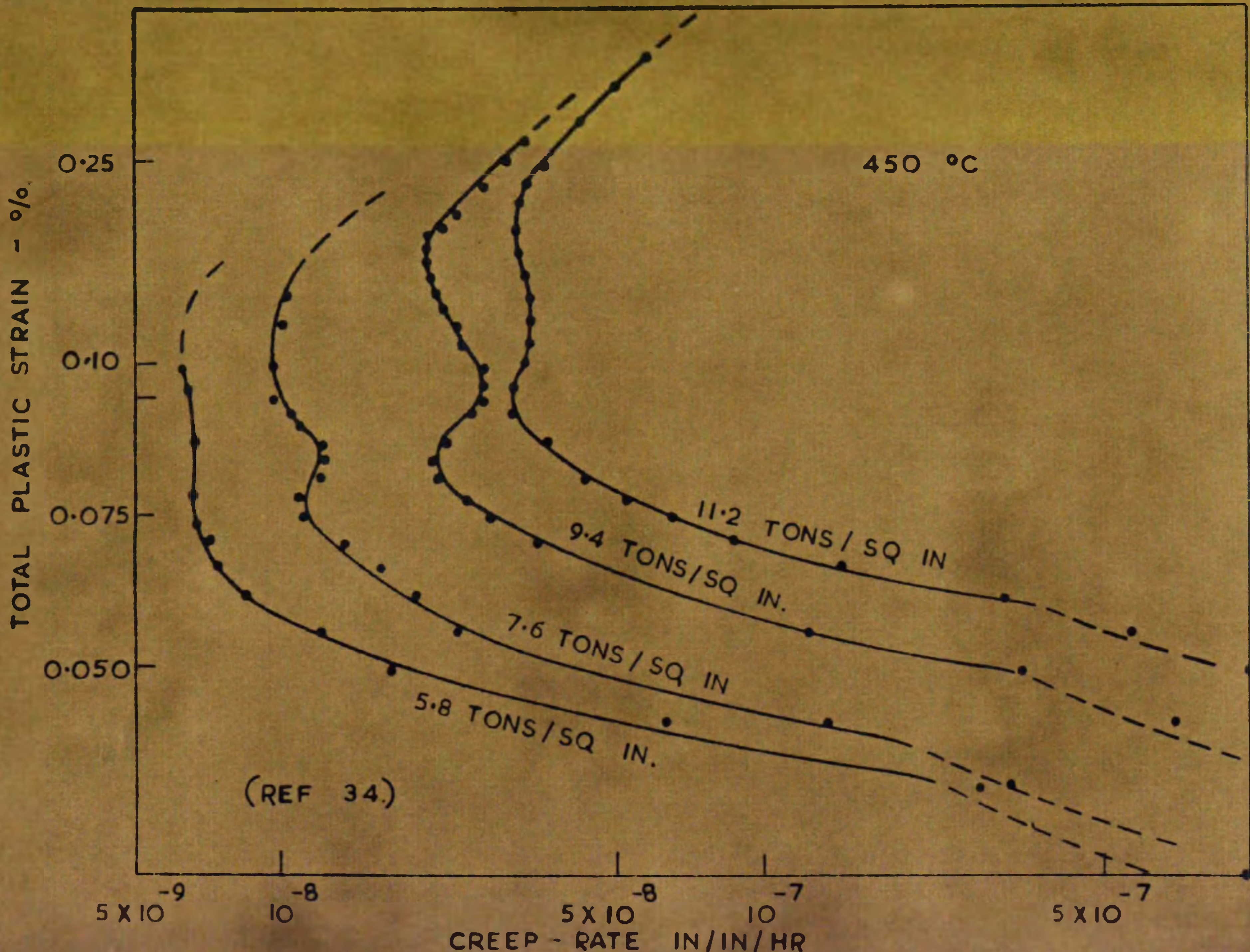
STRAIN - RATE CURVES ON CR. MO. STEEL. 18.

FIG. 29.



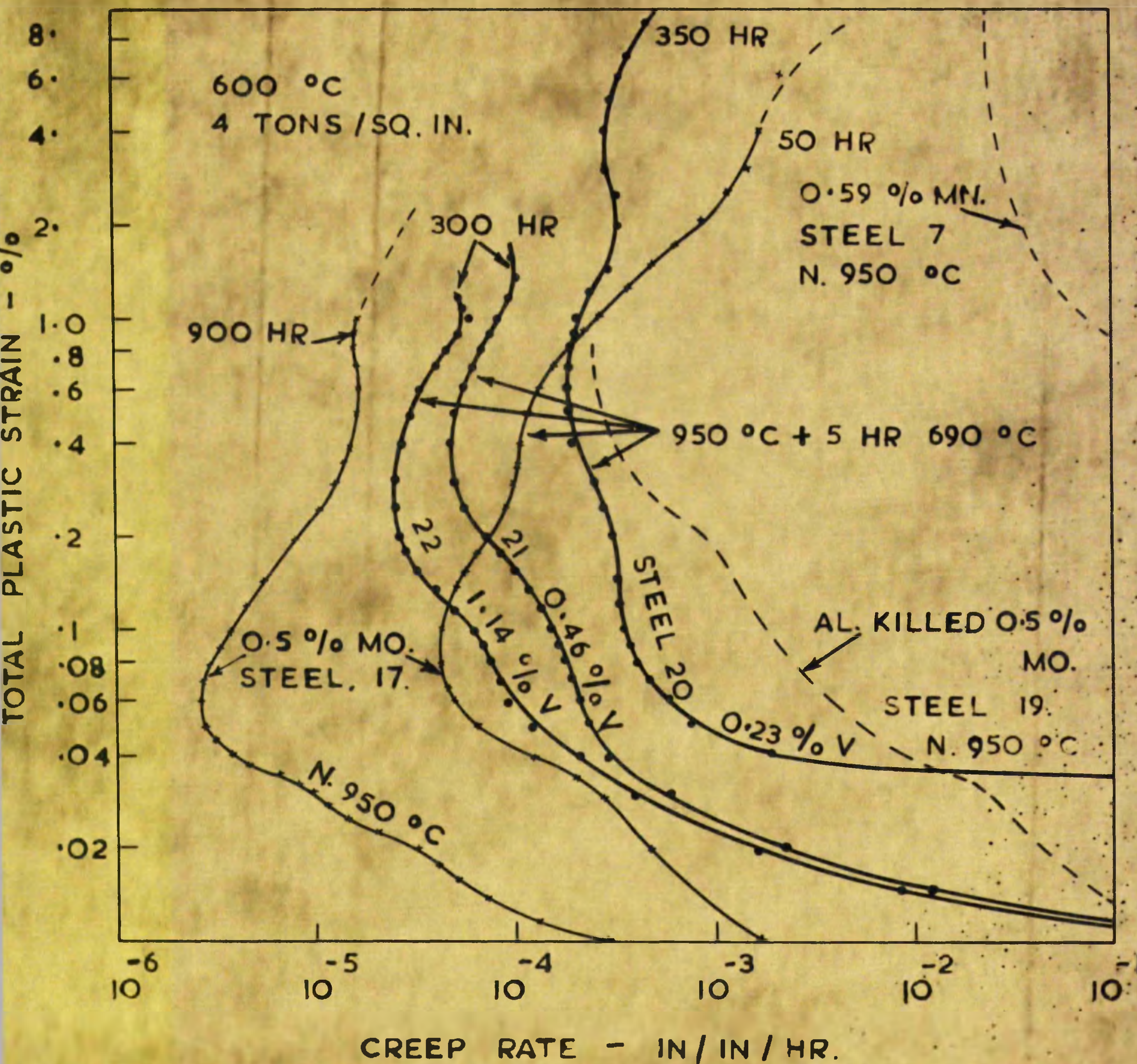
EFFECT OF VARIOUS NORMALISING TREATMENTS ON STRAIN - RATE CURVE FOR CR. MO. STEEL. 18.

FIG. 30.



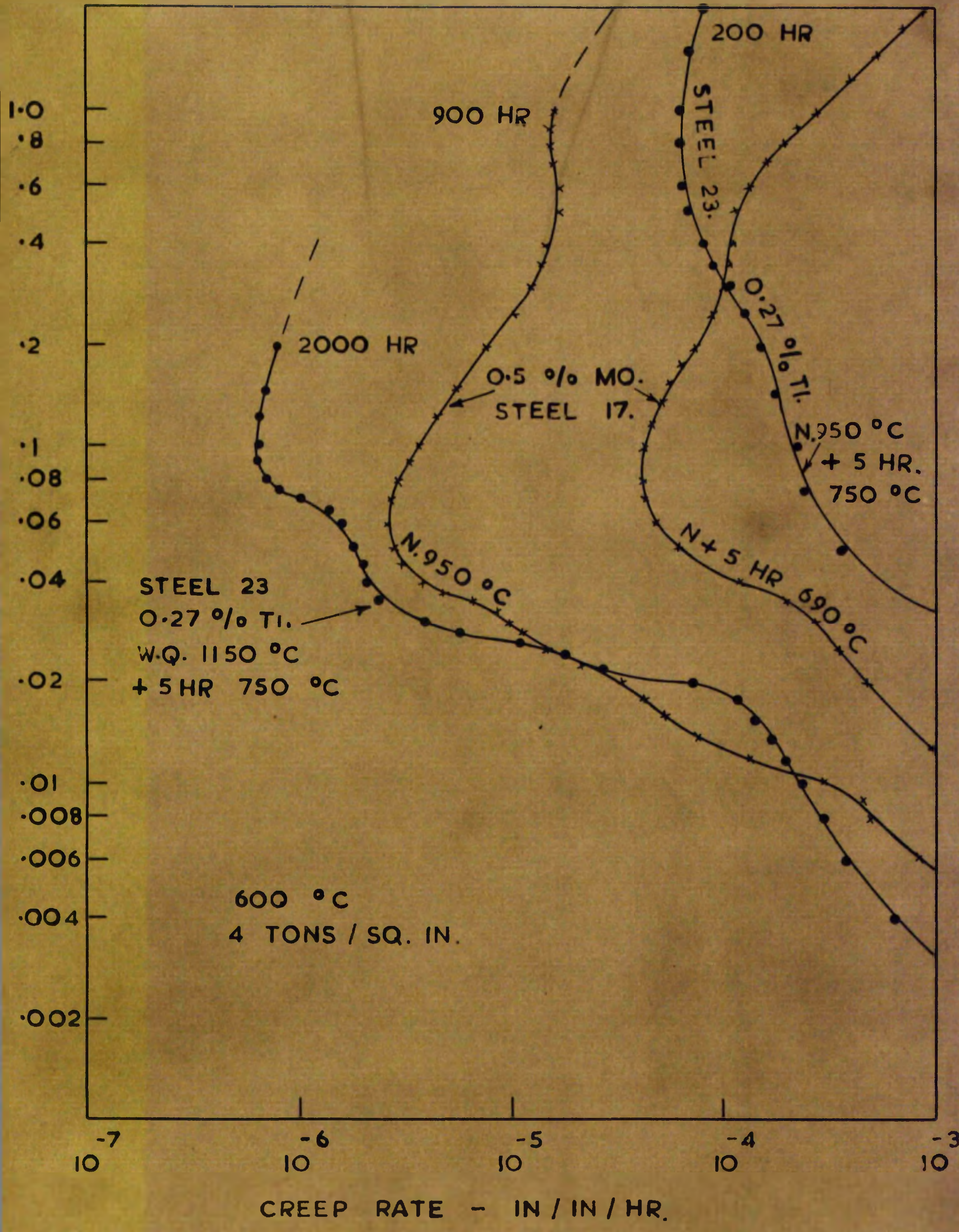
100,000 HR STRAIN RATE CURVES ON NI. CR. MO. STEEL.

FIG. 31.



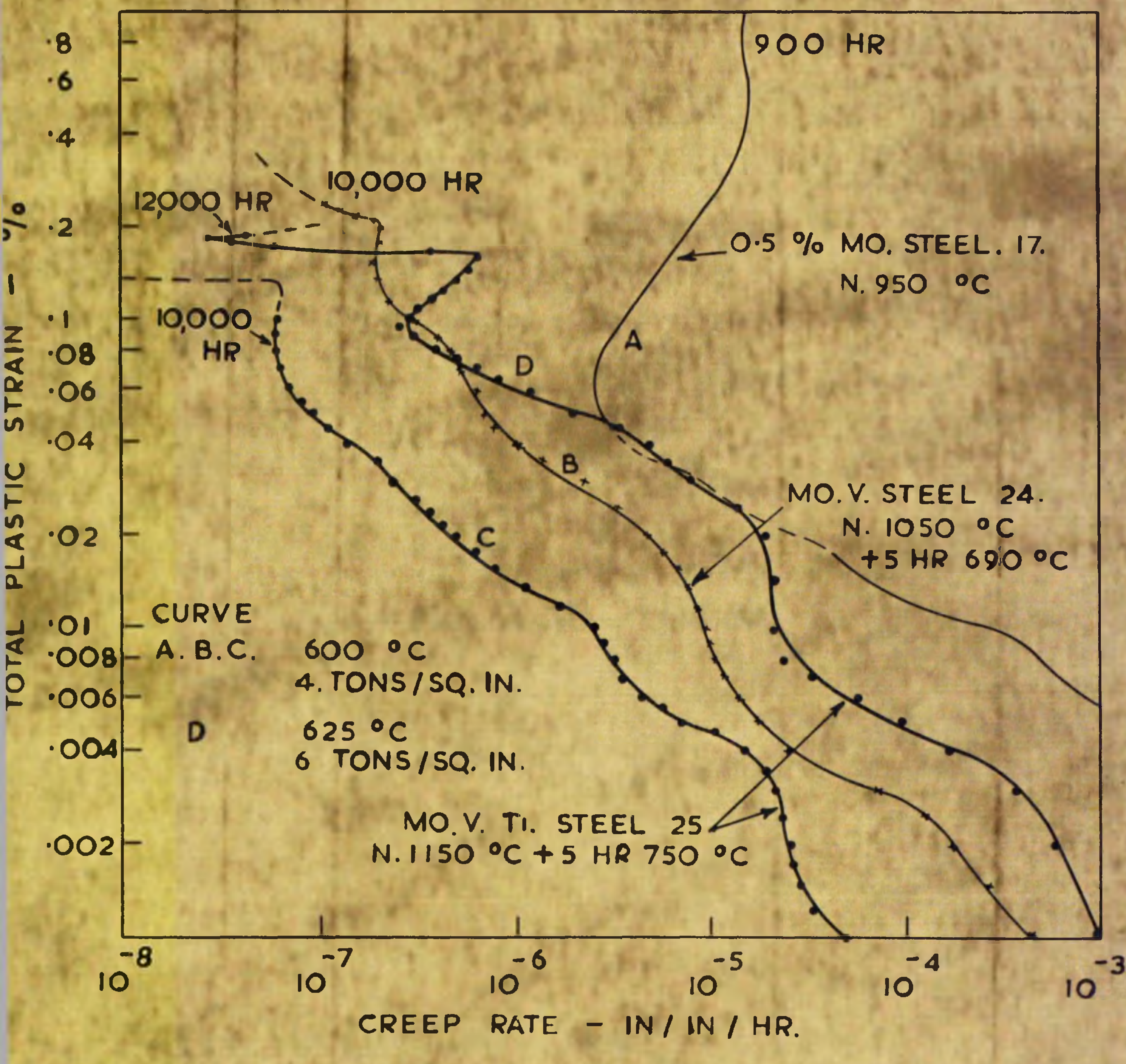
STRAIN RATE CURVES OF VANADIUM STEELS.

FIG. 32.



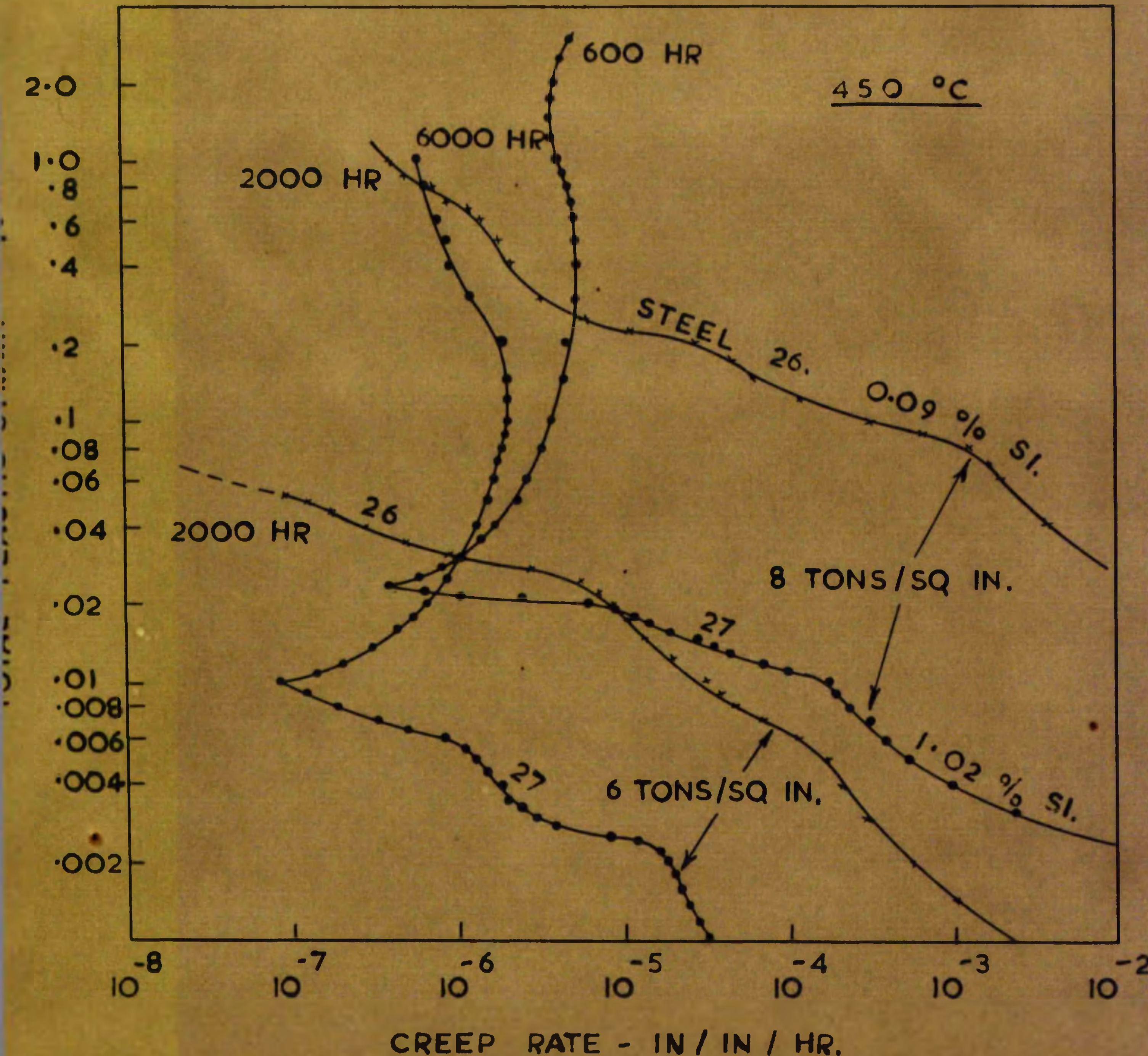
STRAIN-RATE CURVES ON TITANIUM STEEL.

FIG. 33.



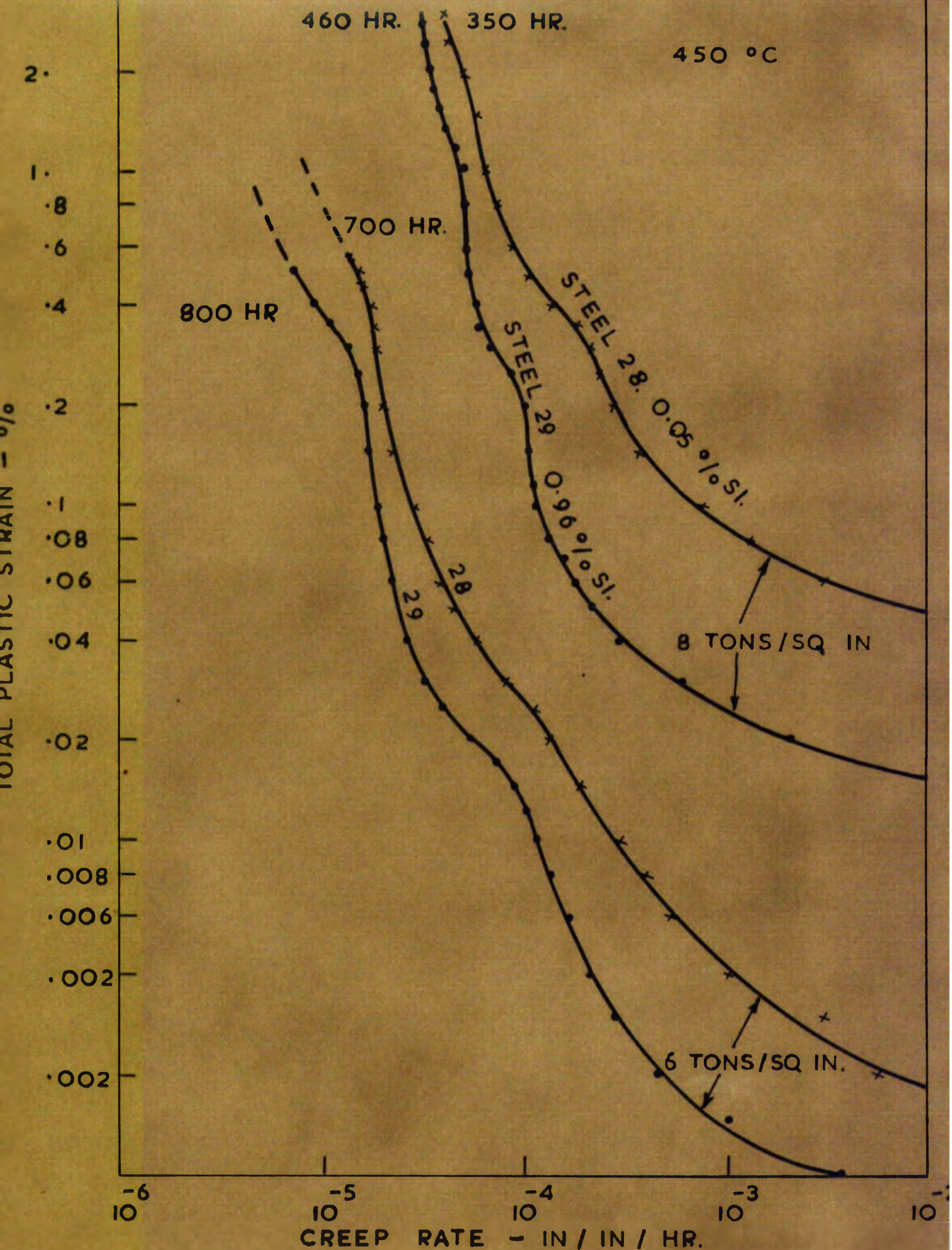
STRAIN - RATE CURVES ON MO. V. & MO. V. Ti. STEELS.

FIG. 34.



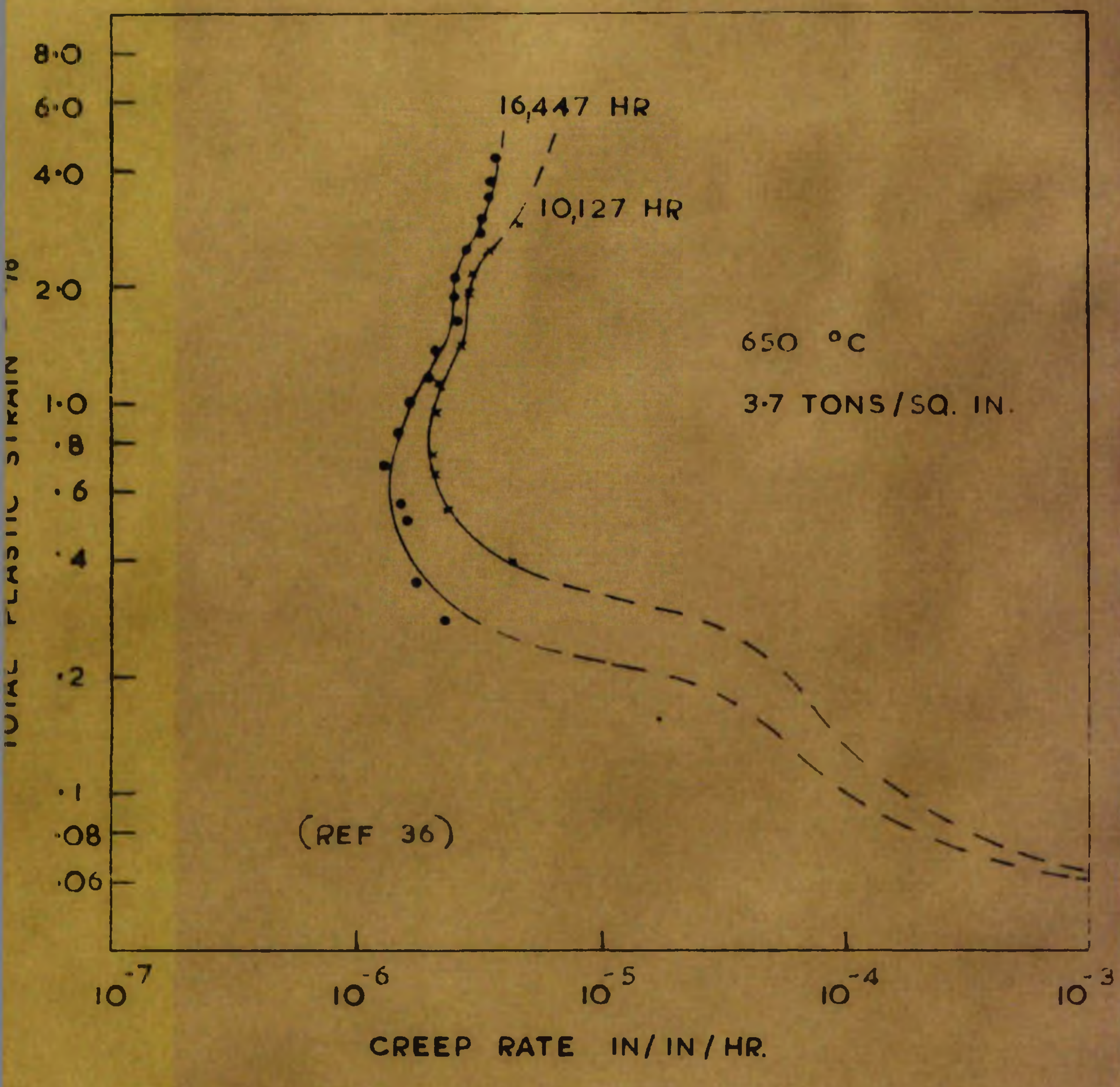
STRAIN - RATE CURVES ON 1 % SI. STEEL.

FIG. 35



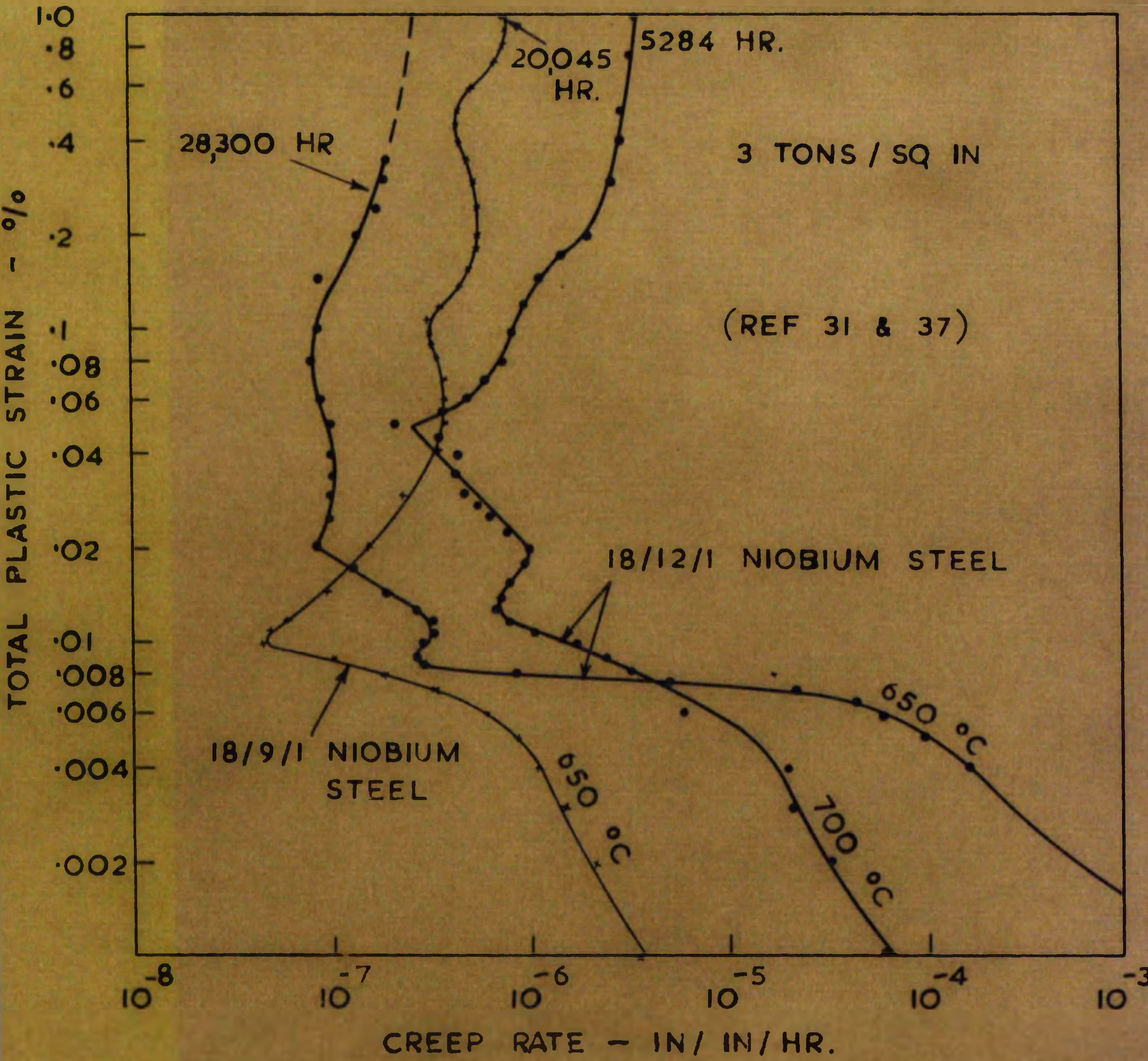
STRAIN - RATE CURVES ON 1 % SI. STEEL (AL KILLED)

FIG. 36



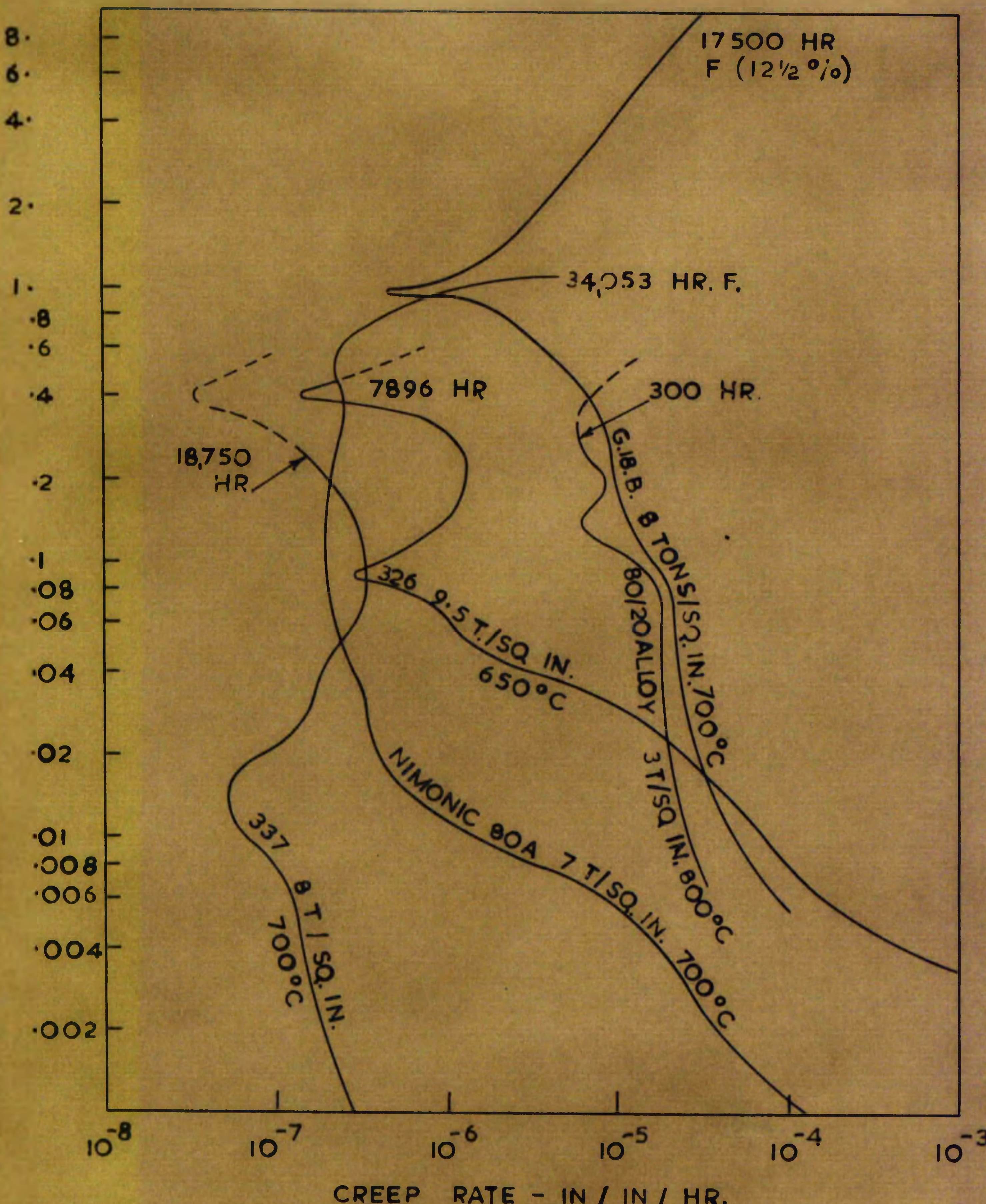
STRAIN-RATE CURVES ON 18/8 STEEL.

FIG. 37.



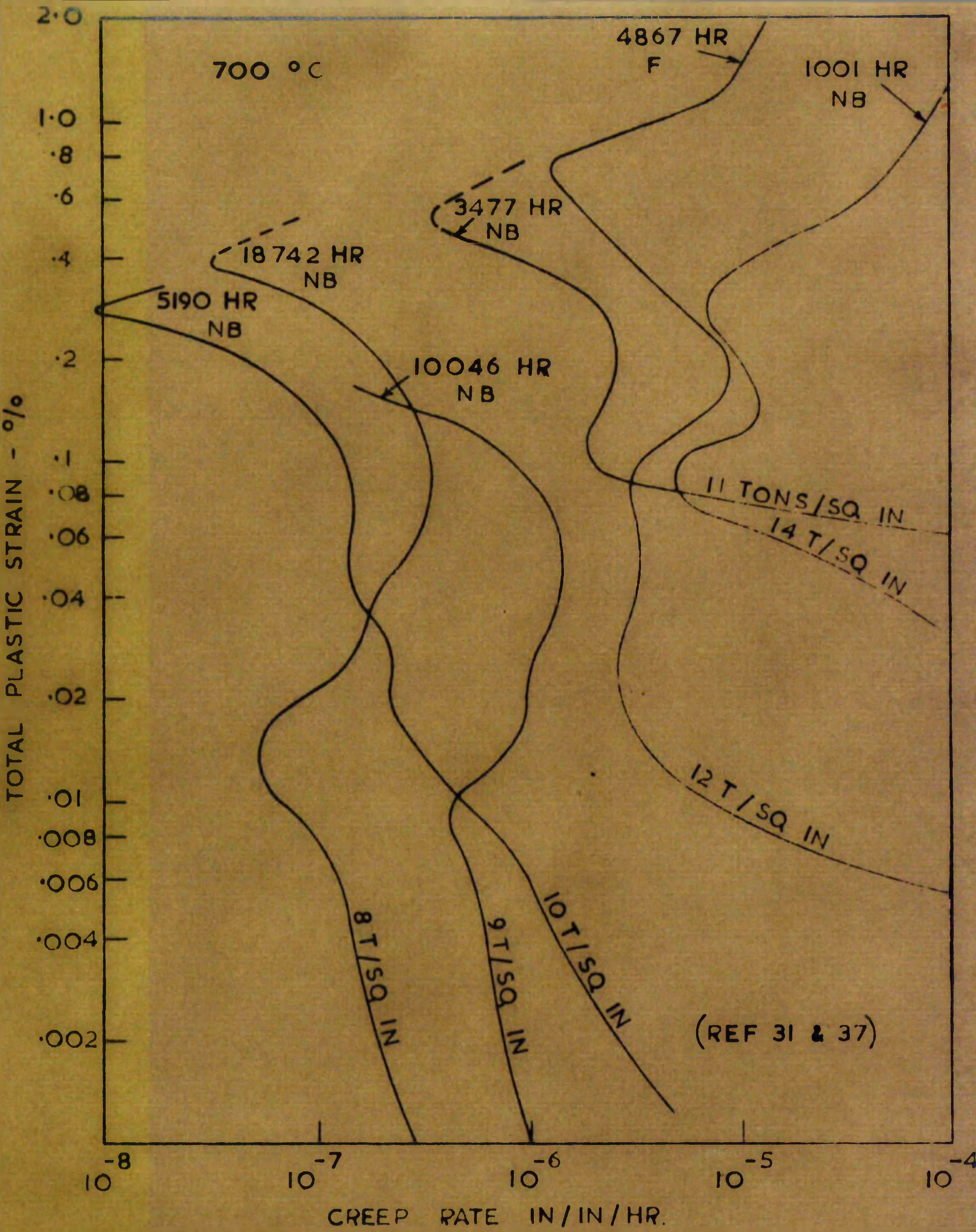
STRAIN - RATE CURVES ON 18/9/1 & 18/12/1 NIOBIUM STEEL.

FIG. 38.



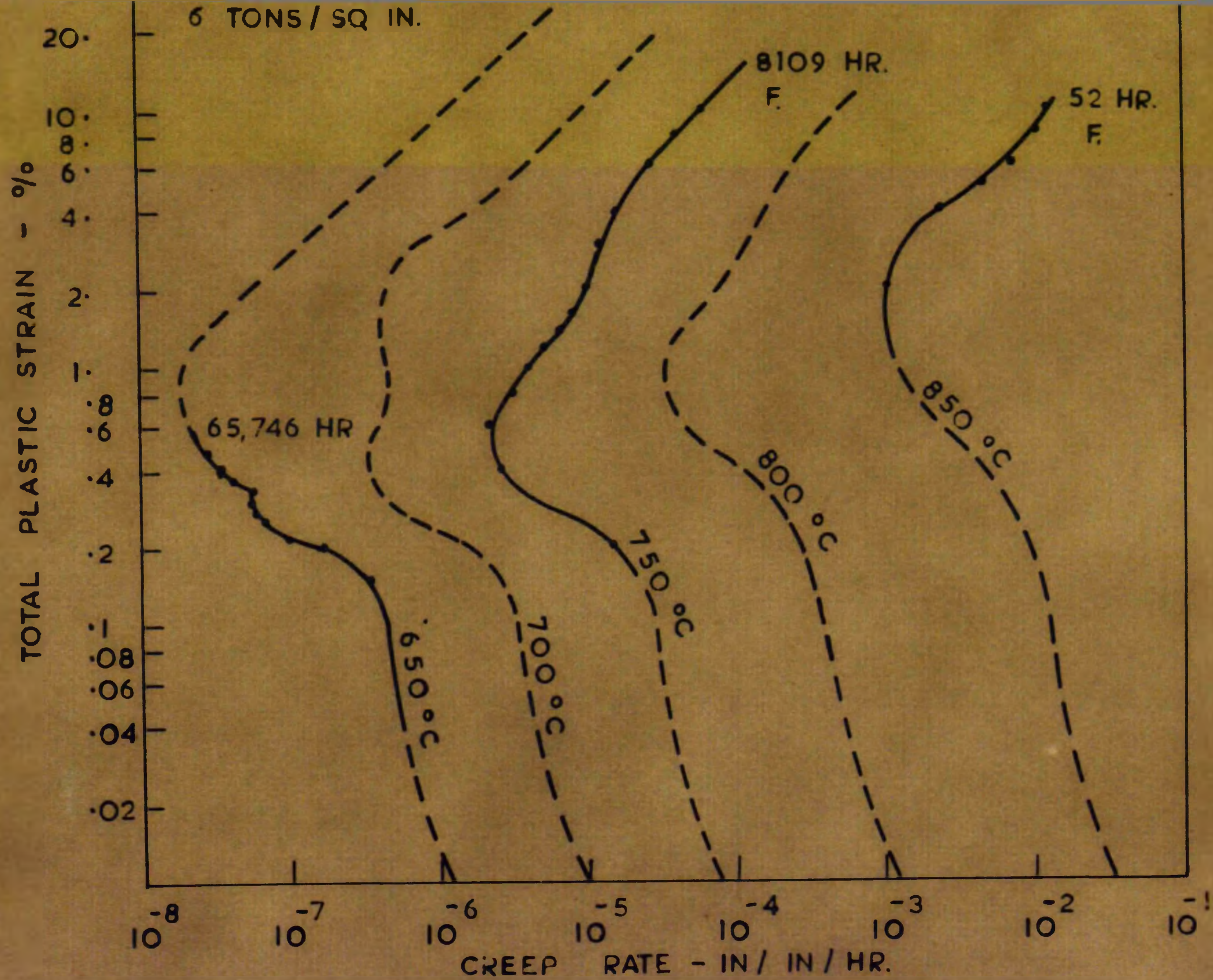
STRAIN RATE CURVES ON VARIOUS COMPLEX ALLOYS.

FIG. 39.



STRAIN RATE CURVES AT VARIOUS STRESS ON ALLOY 327.

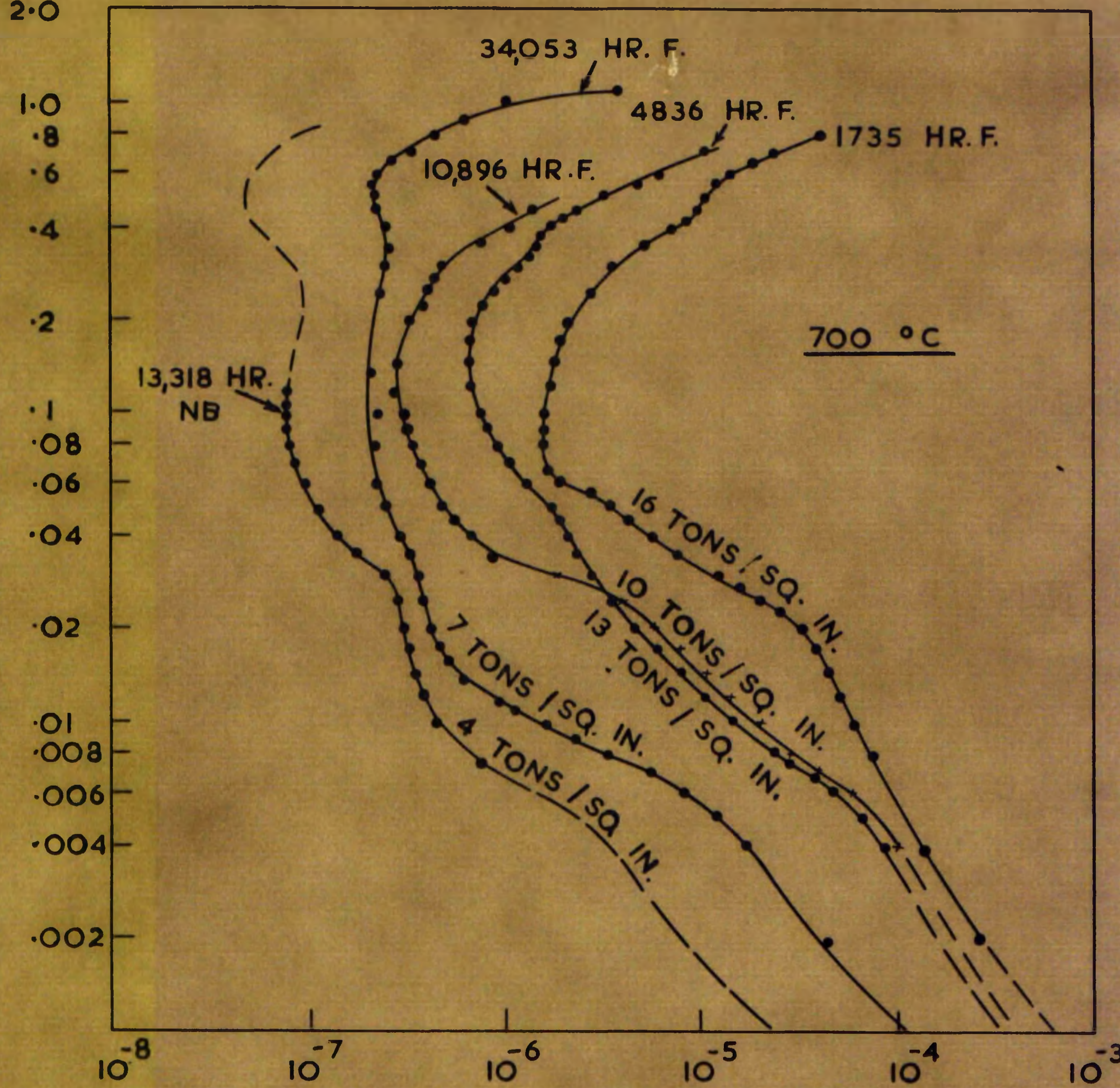
FIG 40



CURVES AT CONSTANT STRESS ON ALLOY G. 18. B.

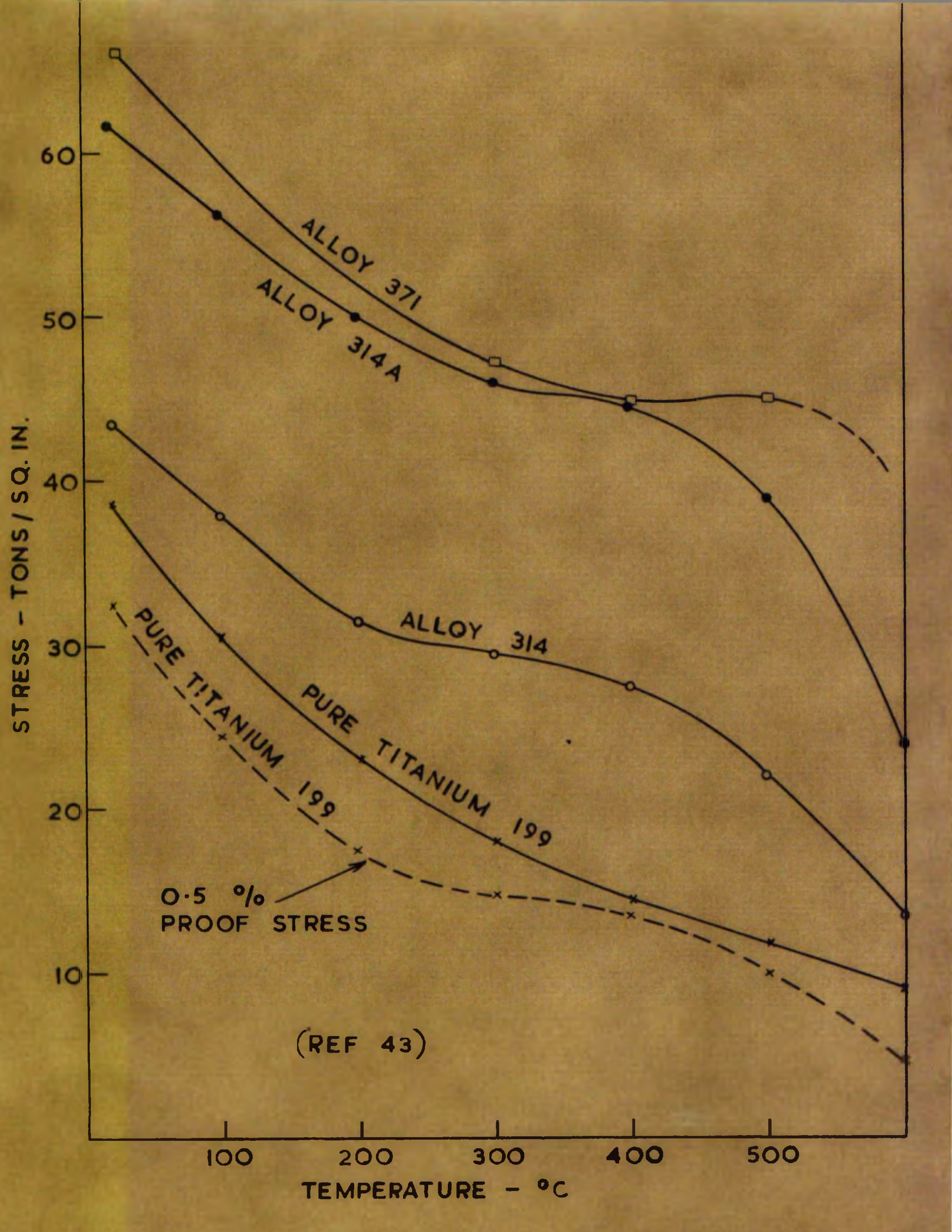
FIG. 41.

FIG. 4



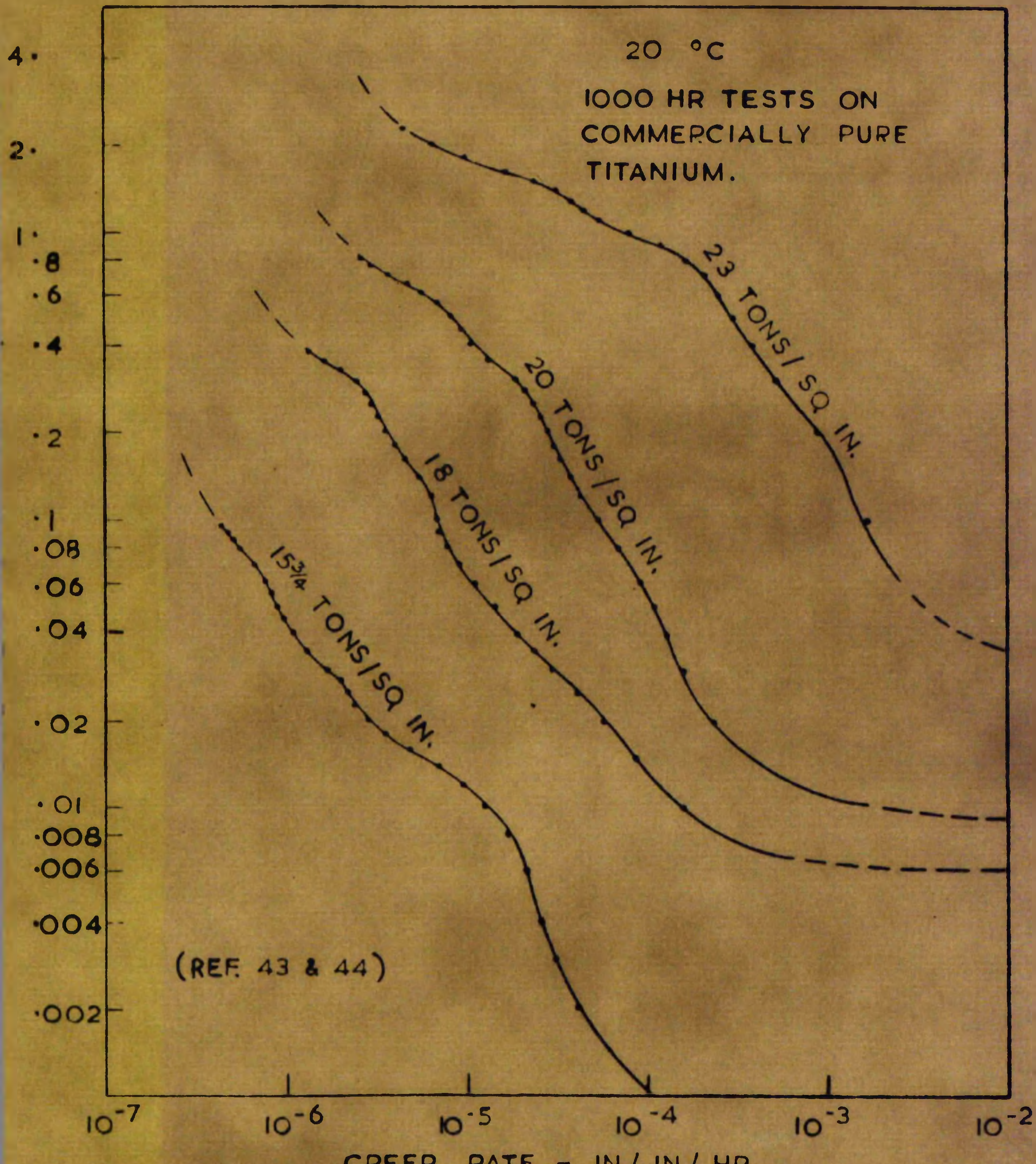
STRAIN - RATE CURVES AT VARIOUS STRESSES.
ON NIMONIC 80 A.

FIG. 42.



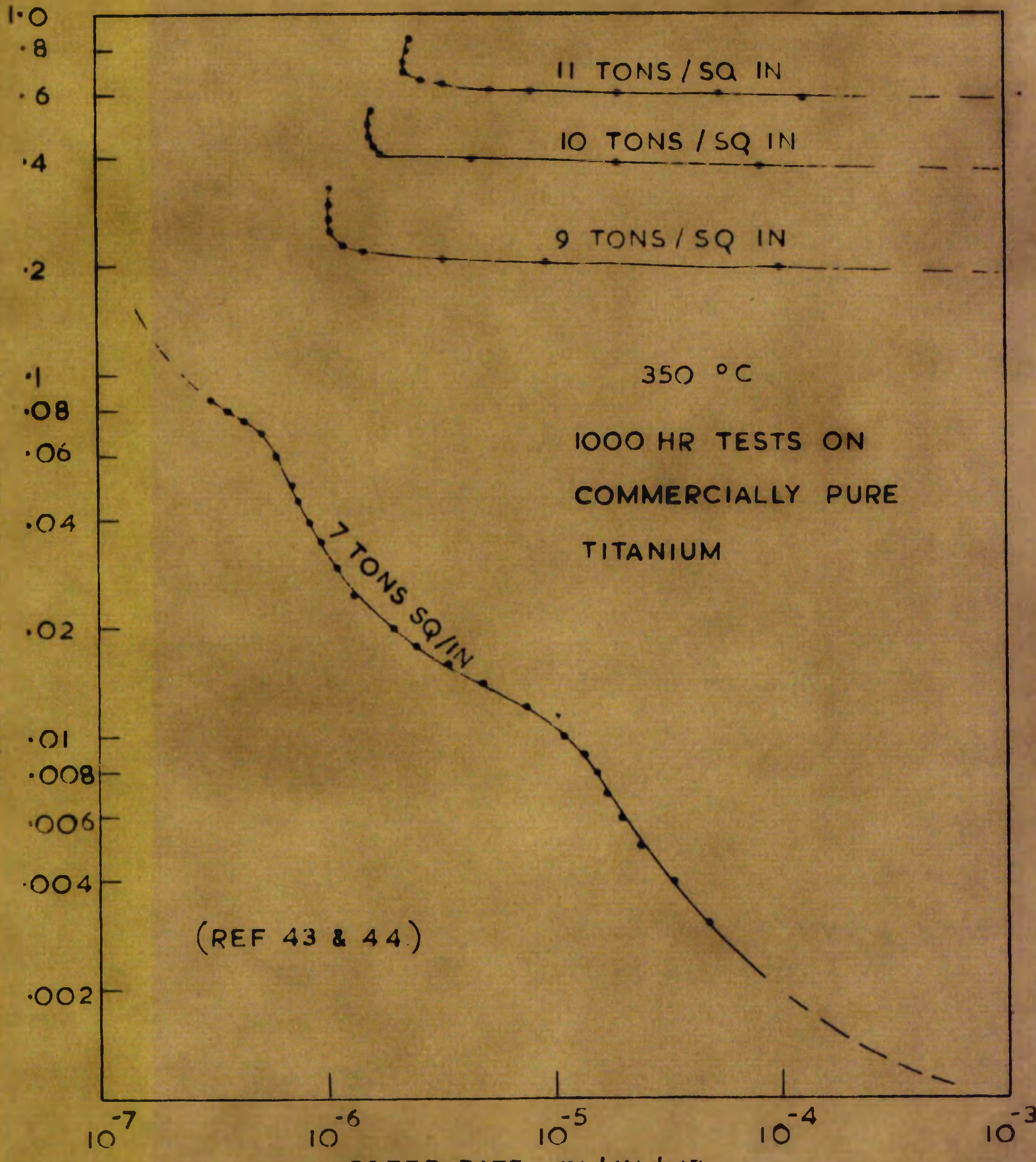
TENSILE STRENGTH OF VARIOUS TITANIUM ALLOYS.

FIG. 43.



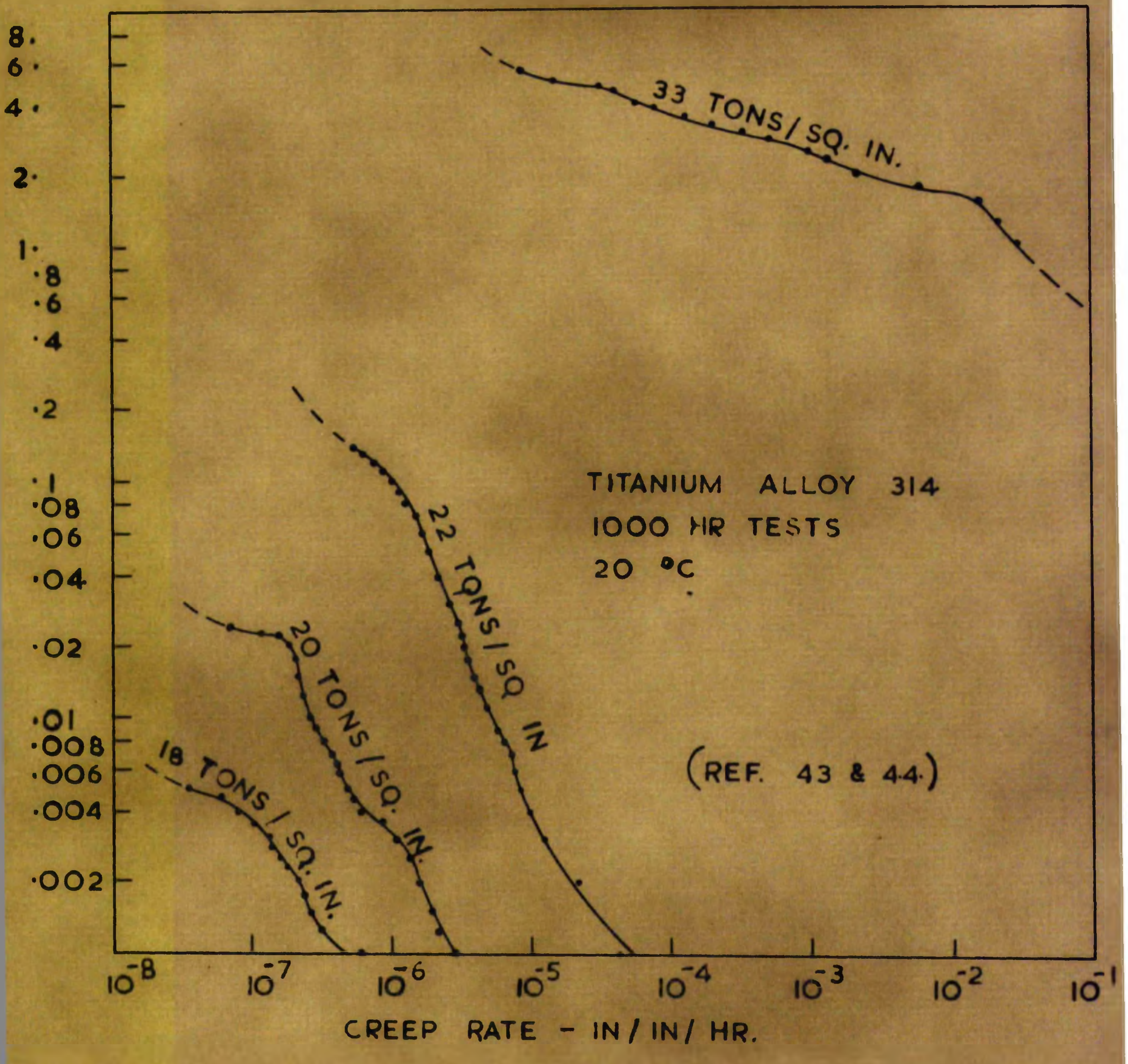
STRAIN RATE CURVES AT ROOM TEMPERATURE ON COMMERCIALLY PURE TITANIUM

FIG. 44



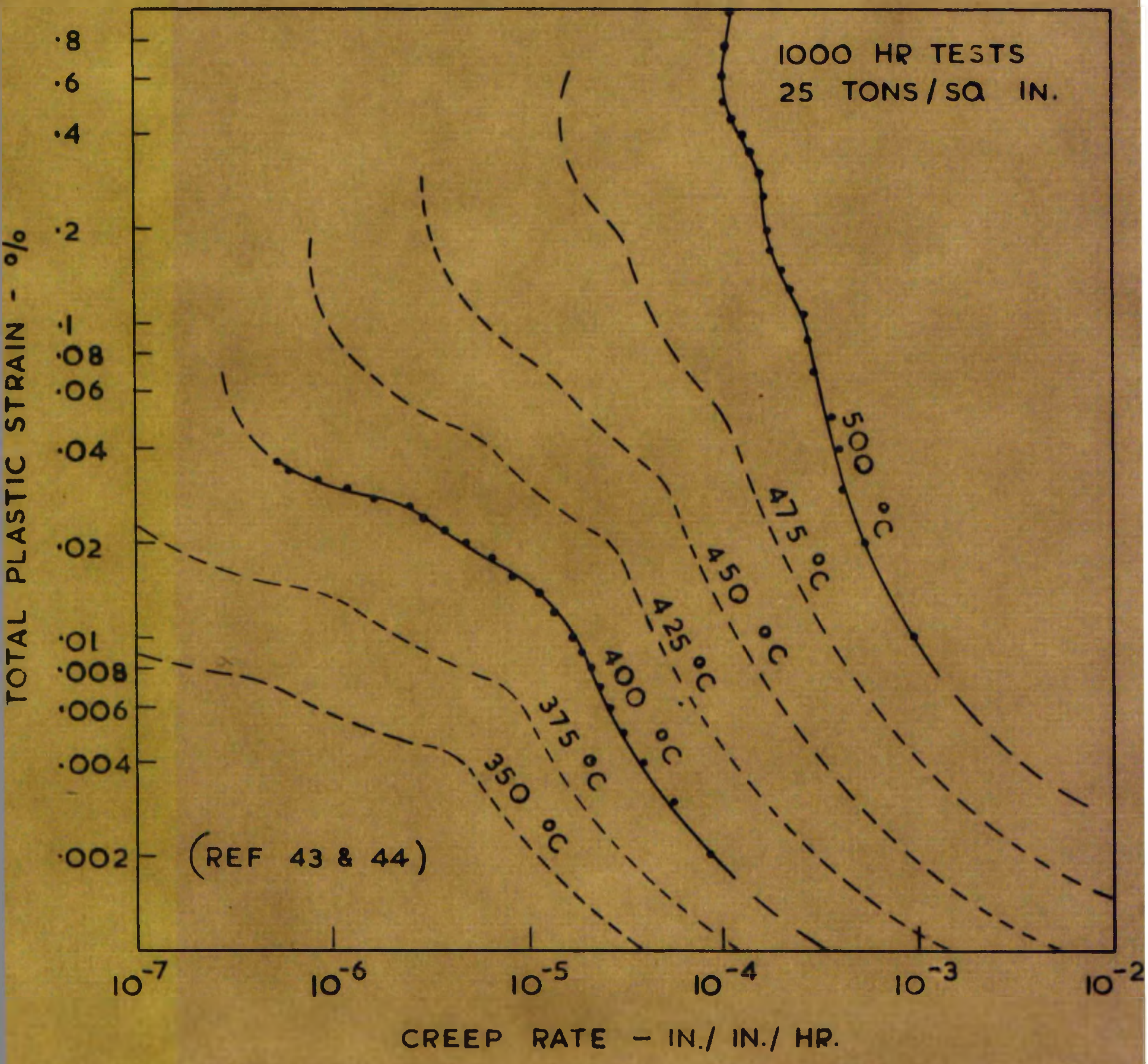
STRAIN-RATE CURVES AT 350 °C ON COMMERCIALLY PURE TITANIUM.

FIG. 45.



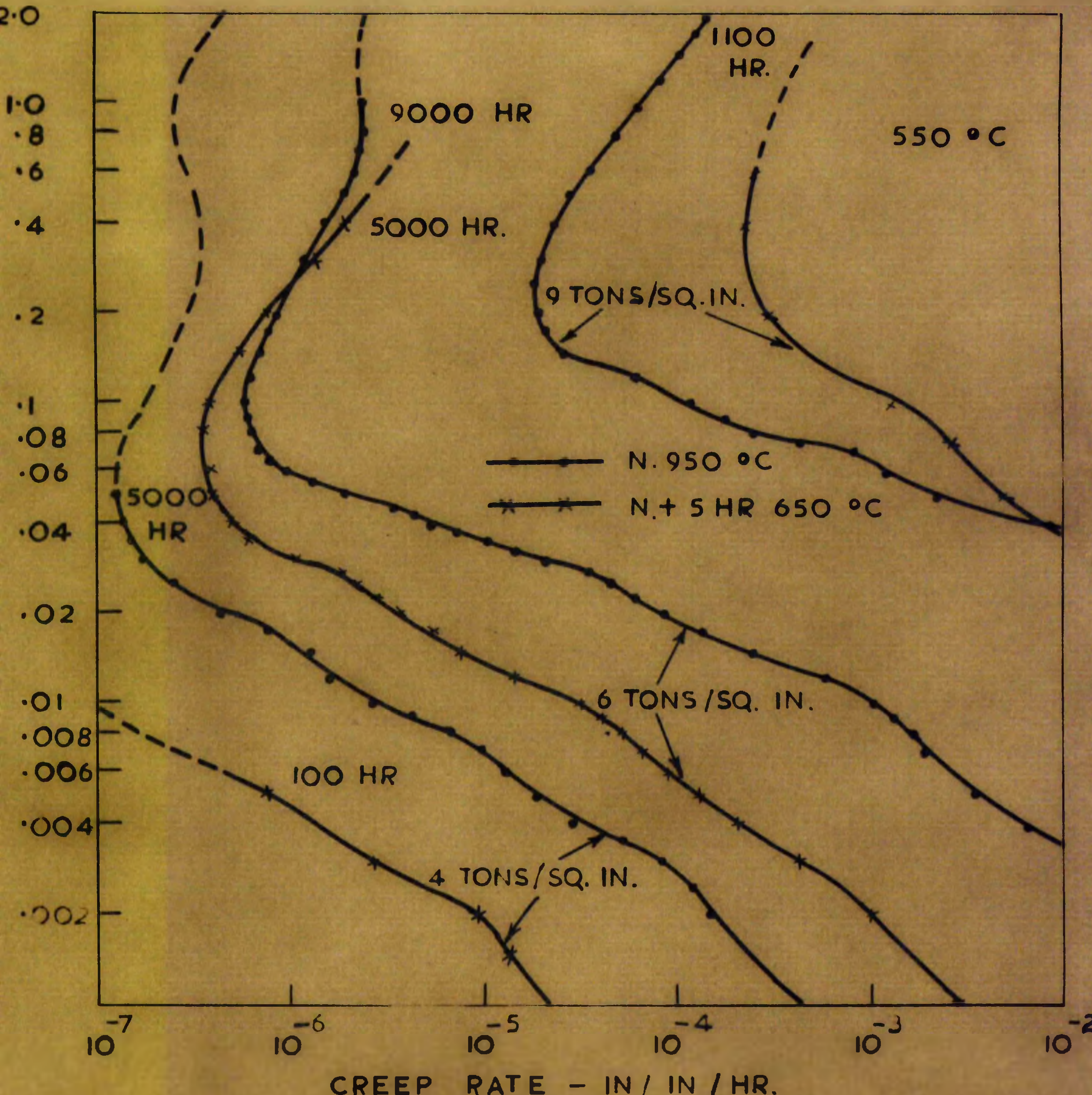
ROOM TEMPERATURE STRAIN-RATE CURVES ON TITANIUM ALLOY 314

FIG. 46.



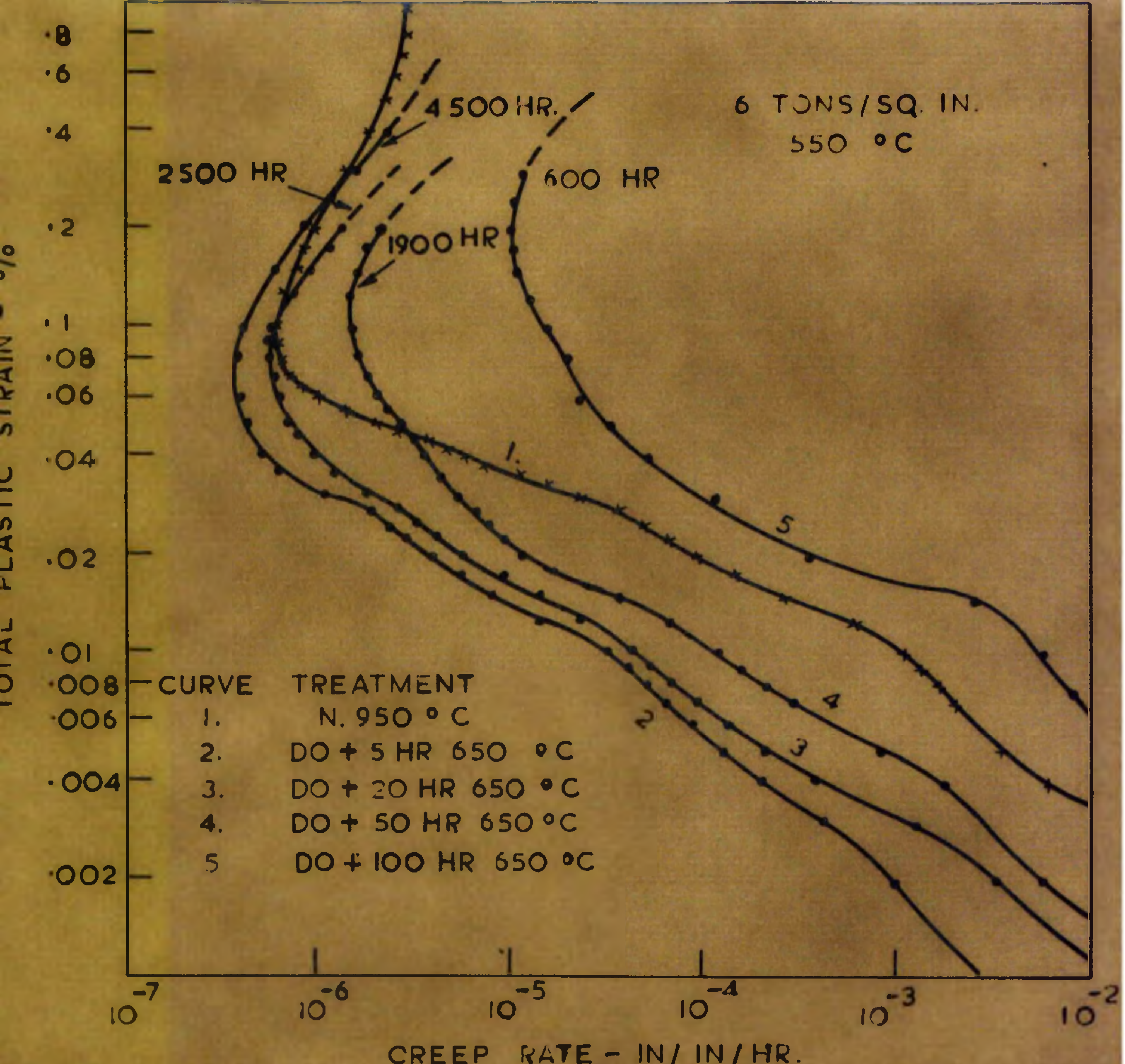
FAMILY OF STRAIN-RATE CURVES AT CONSTANT STRESS ON TITANIUM ALLOY 371.

FIG. 47.



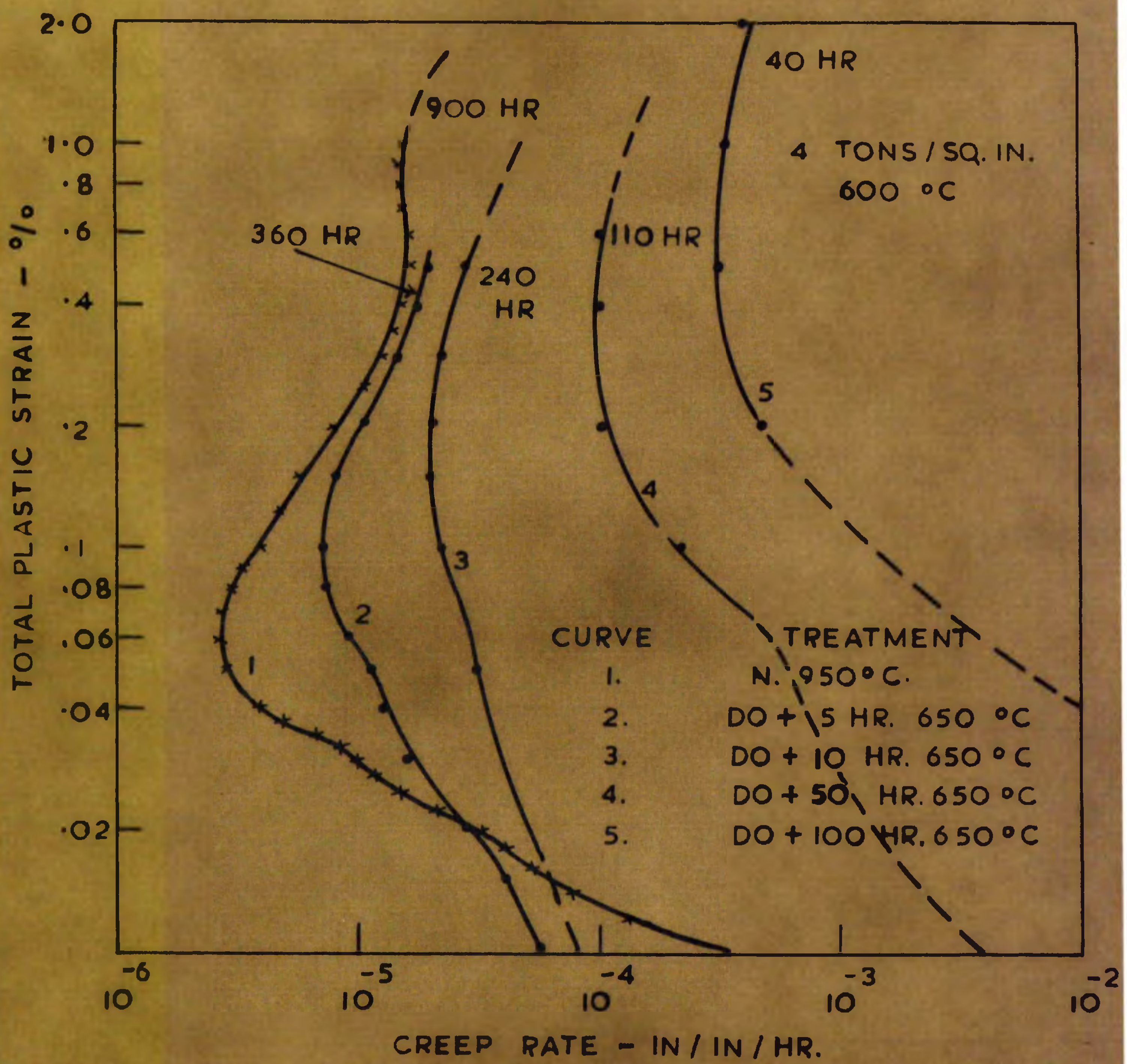
EFFECT OF TEMPERING ON STRAIN - RATE CURVES OF 0·5% MO. STEEL. 17.

FIG. 48.



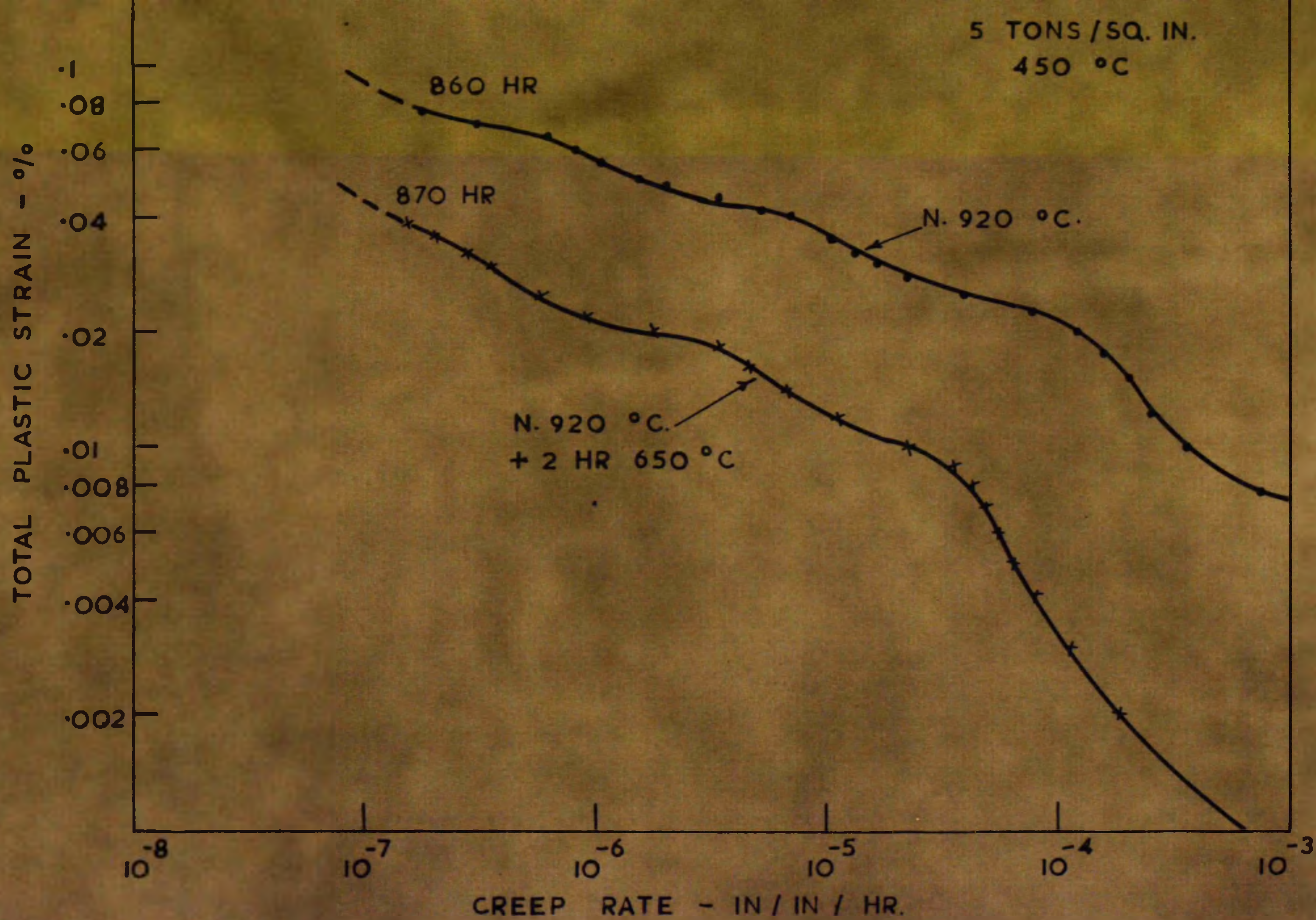
EFFECT OF TEMPERING AT 650 °C ON
0·5 % MO STEEL 17.

FIG. 49.



EFFECT OF TEMPERING AT 650 °C. ON 0·5 % MO. STEEL. 17.

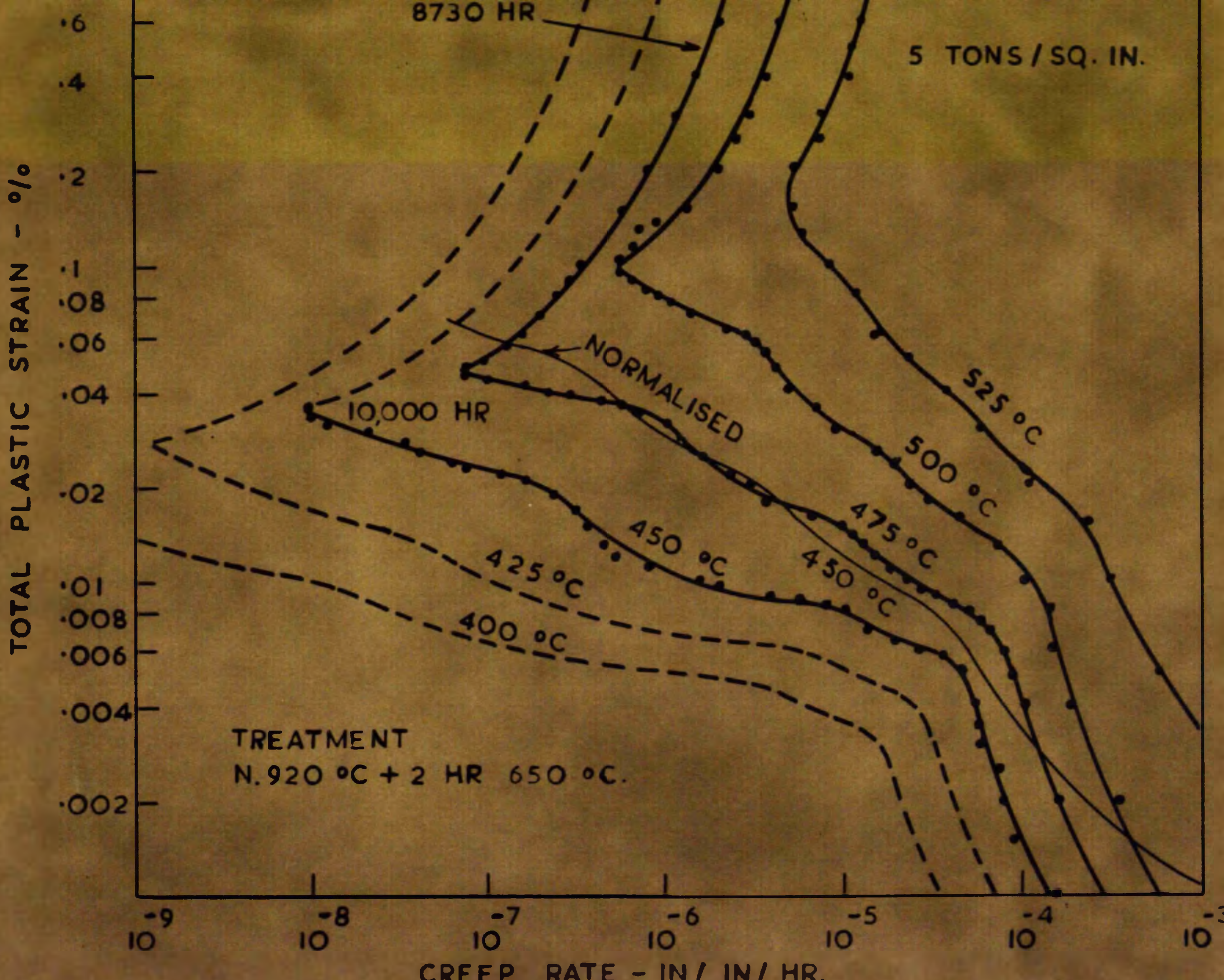
FIG. 50.



EFFECT OF STRESS RELIEVING TREATMENT ON STRAIN-RATE CURVE OF 0·6 % MN. STEEL. 30.

FIG. 51.

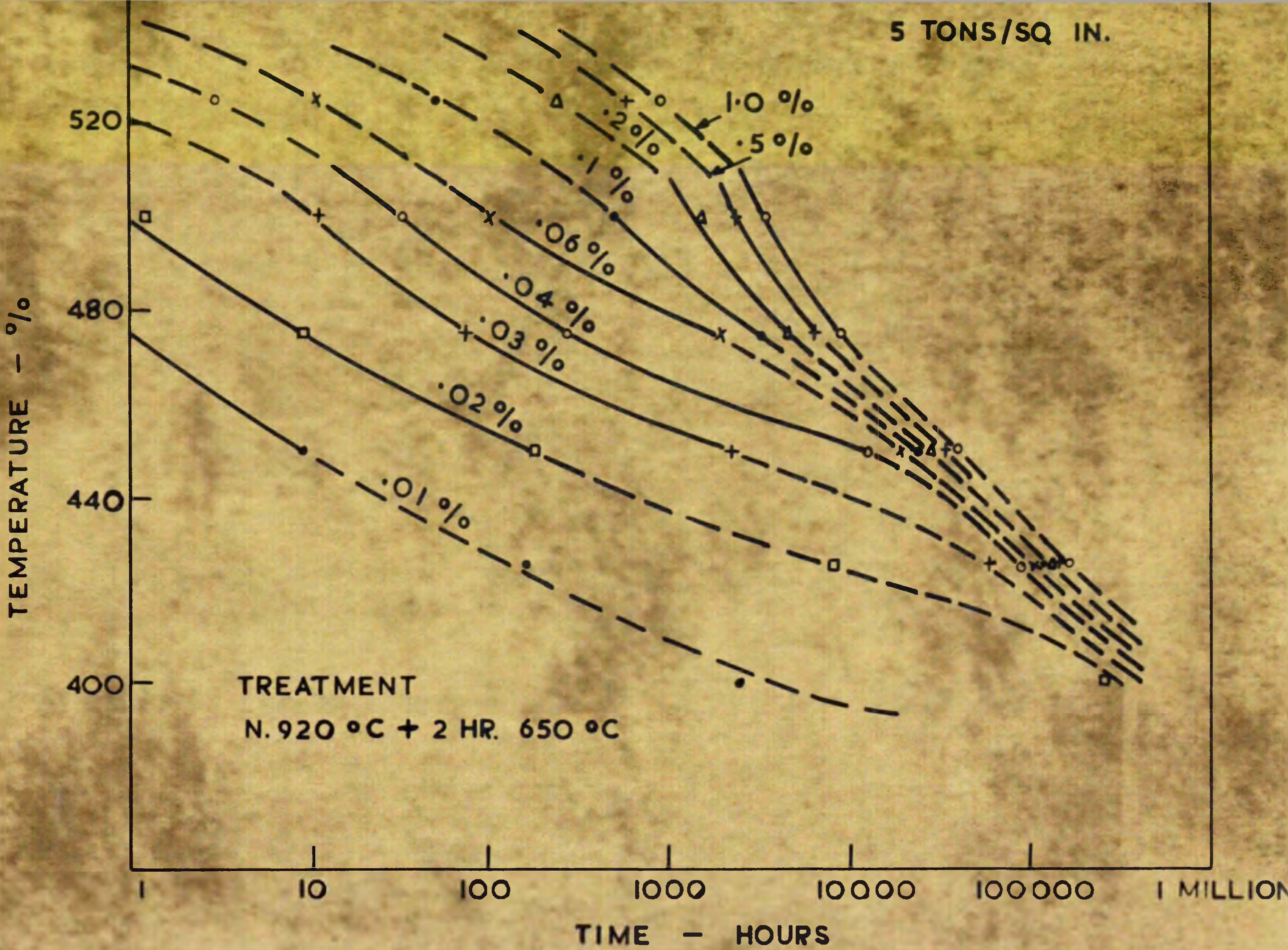
FIG. 5



FAMILY OF STRAIN RATE CURVES ON 1 % MN. STEEL. 31.
IN STRESS-RELIEVED CONDITION.

FIG. 52.

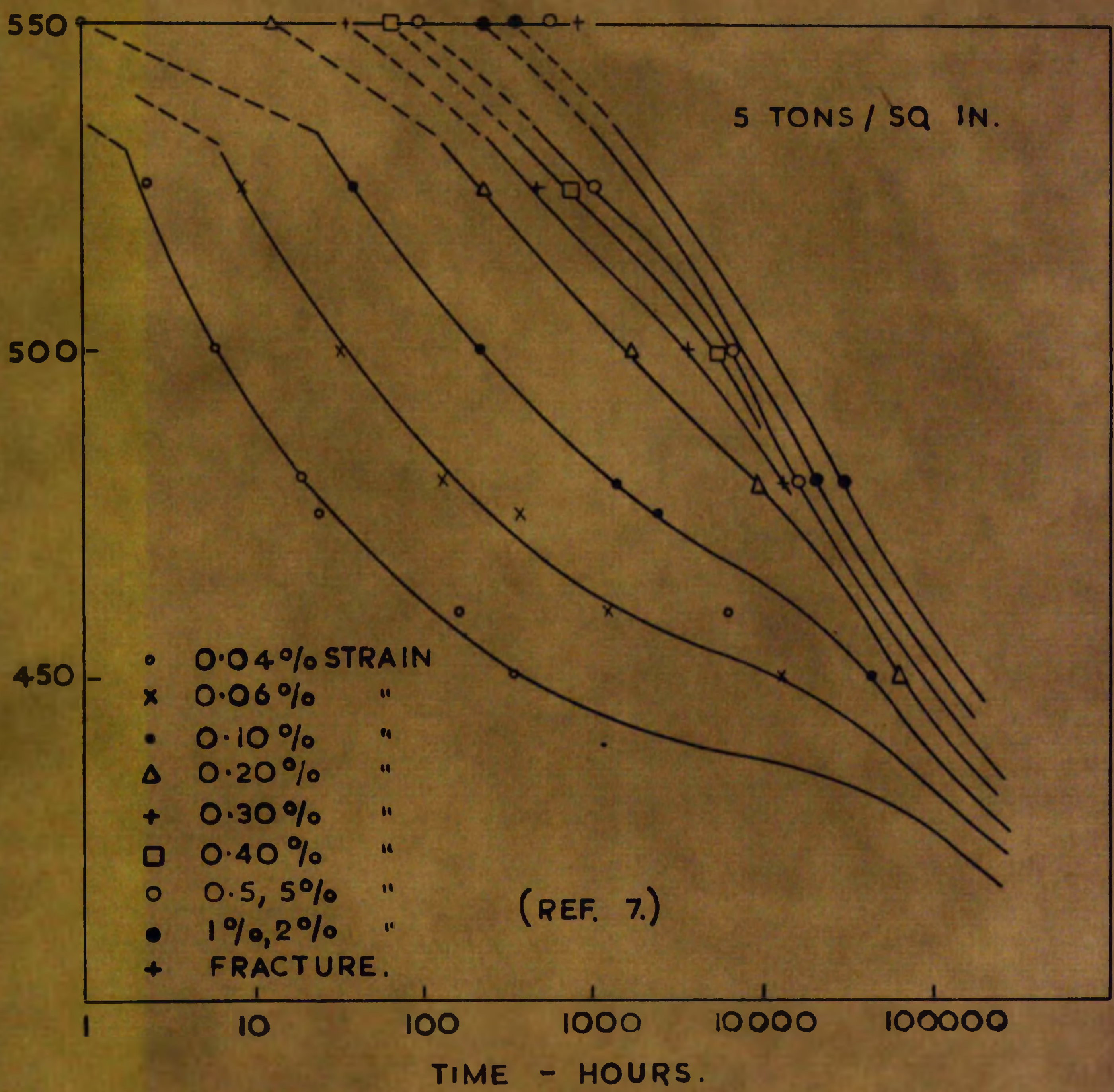
FIG. 52.



ISO-STRAIN CURVES ON 1·2 % MN. STEEL.

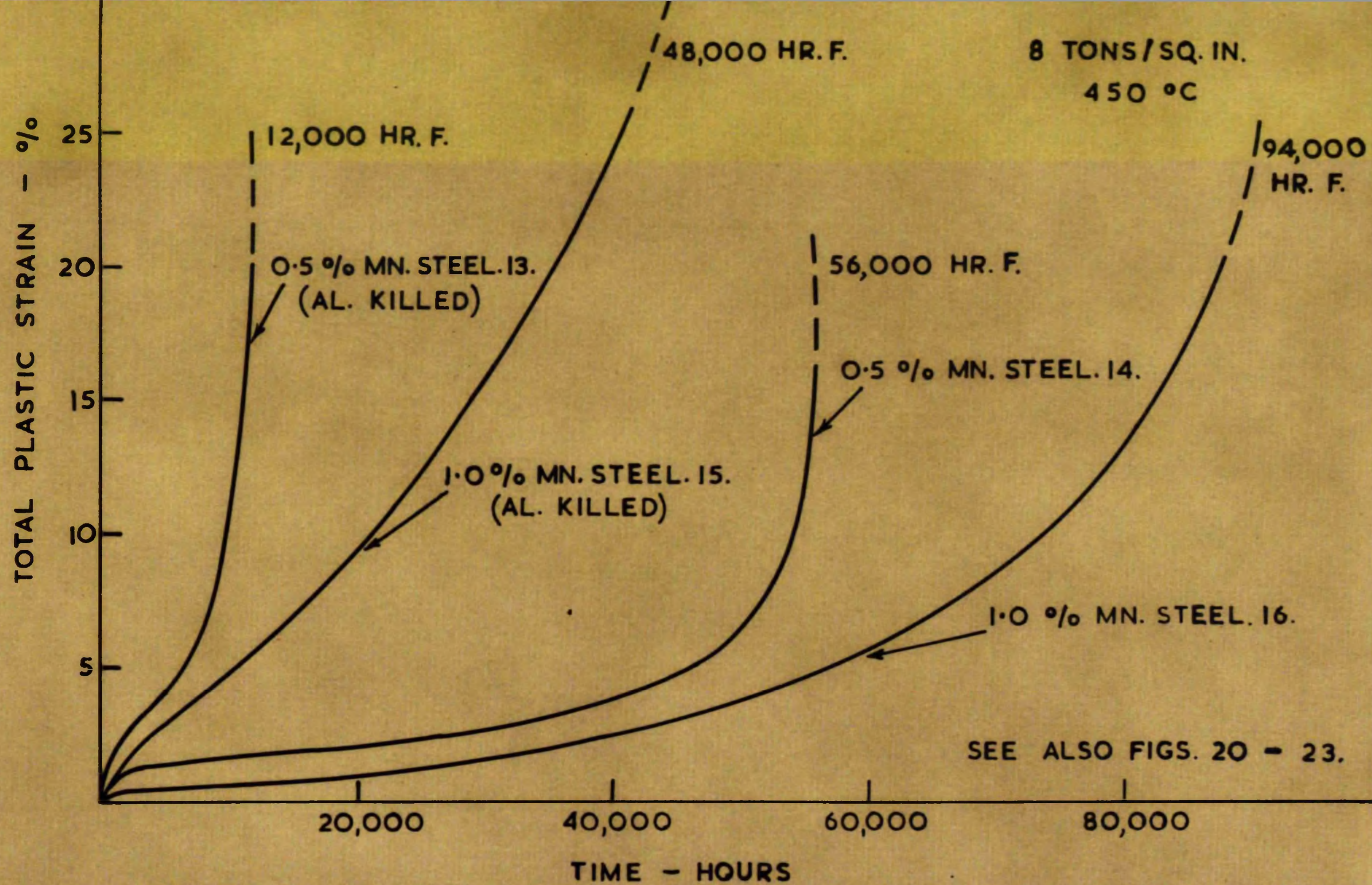
FIG. 53.

FIG.



SUGGESTED CORRECTION OF ISO-STRAIN CURVES ON CARBON STEEL.

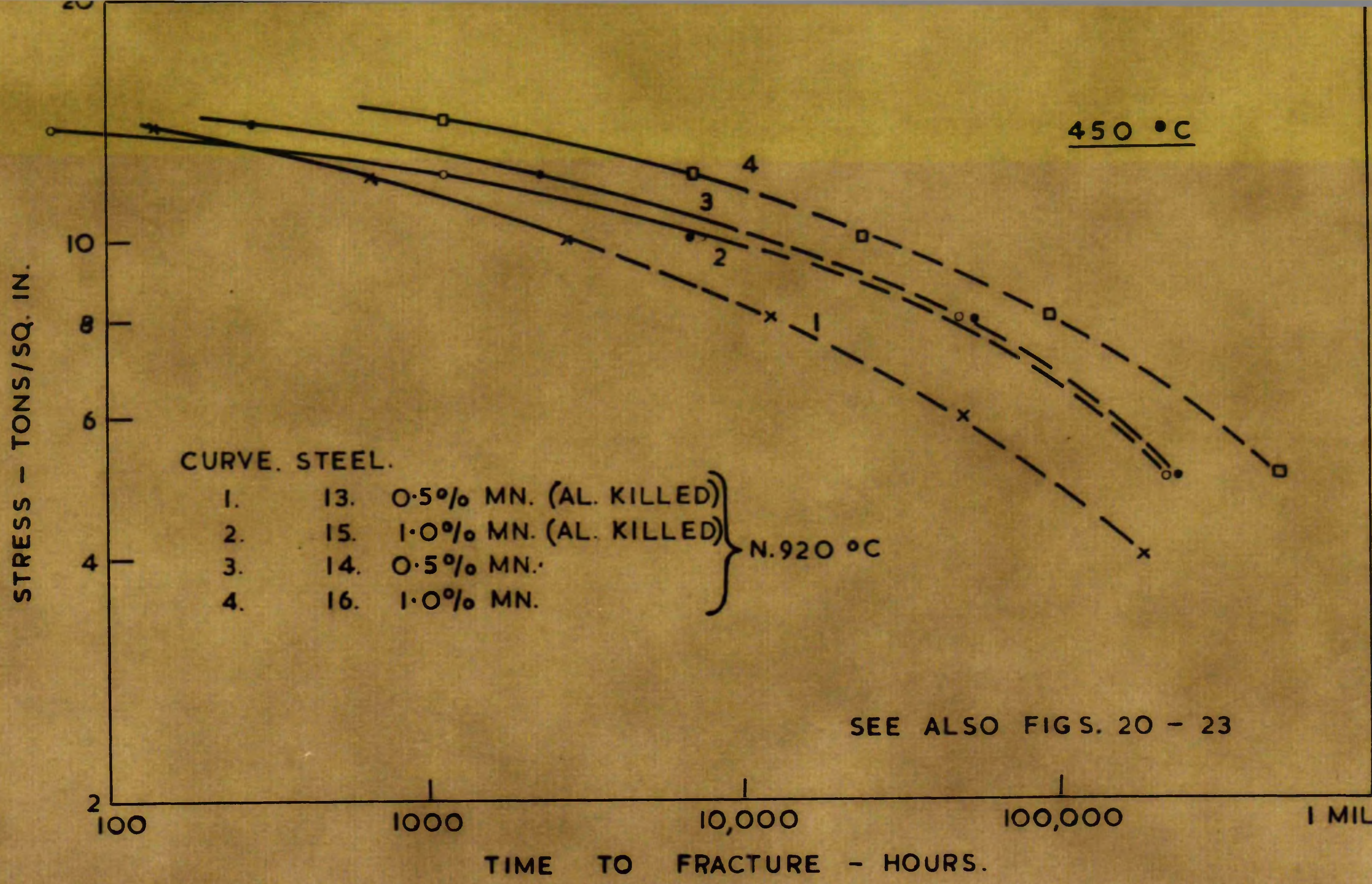
FIG. 54



CREEP CURVES DERIVED FROM STRAIN-RATE CURVES.

FIG. 55.

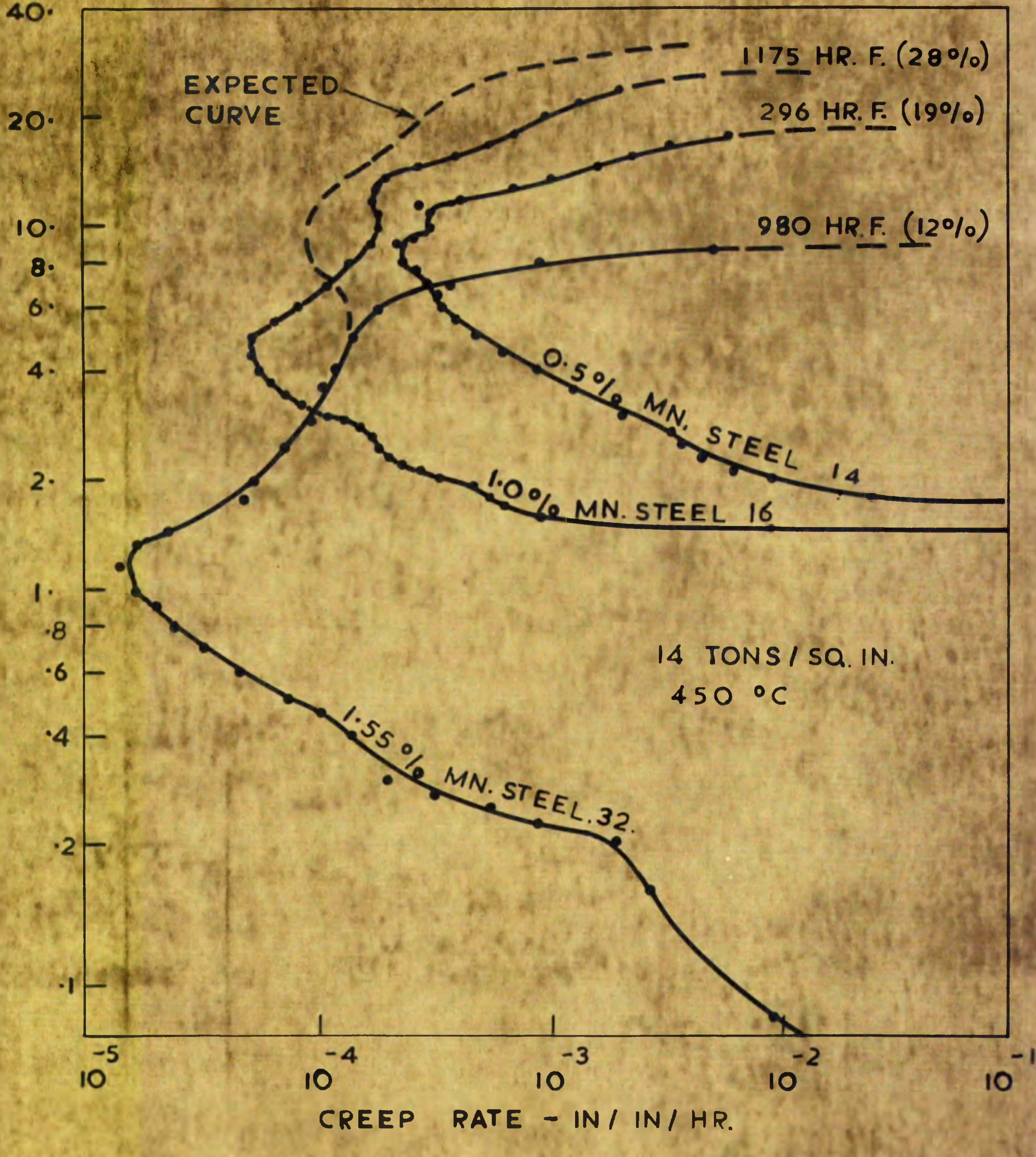
FIG.



RUPTURE TESTS CALCULATED FROM STRAIN-RATE CURVES.

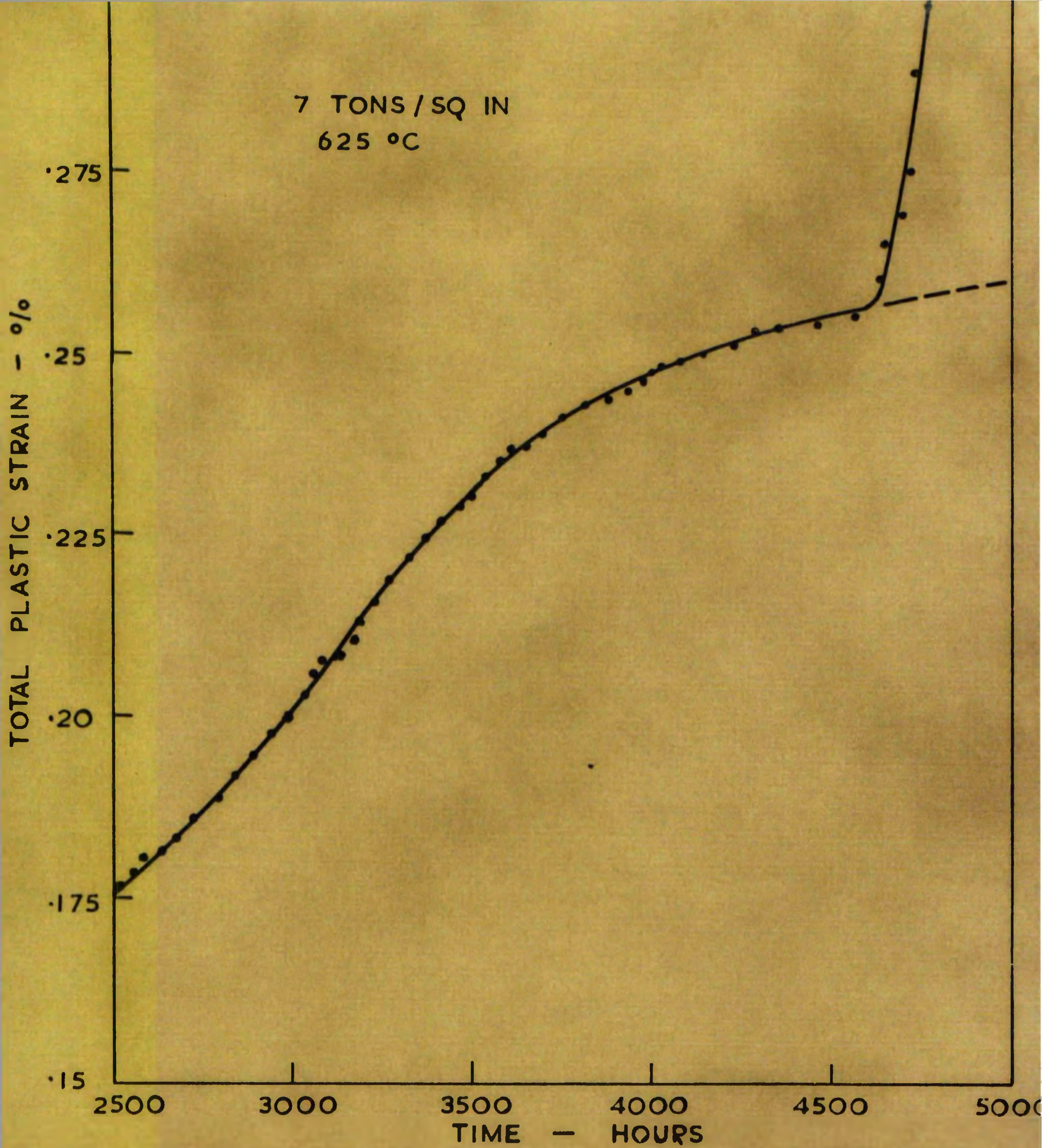
FIG. 56.

FIG. 5



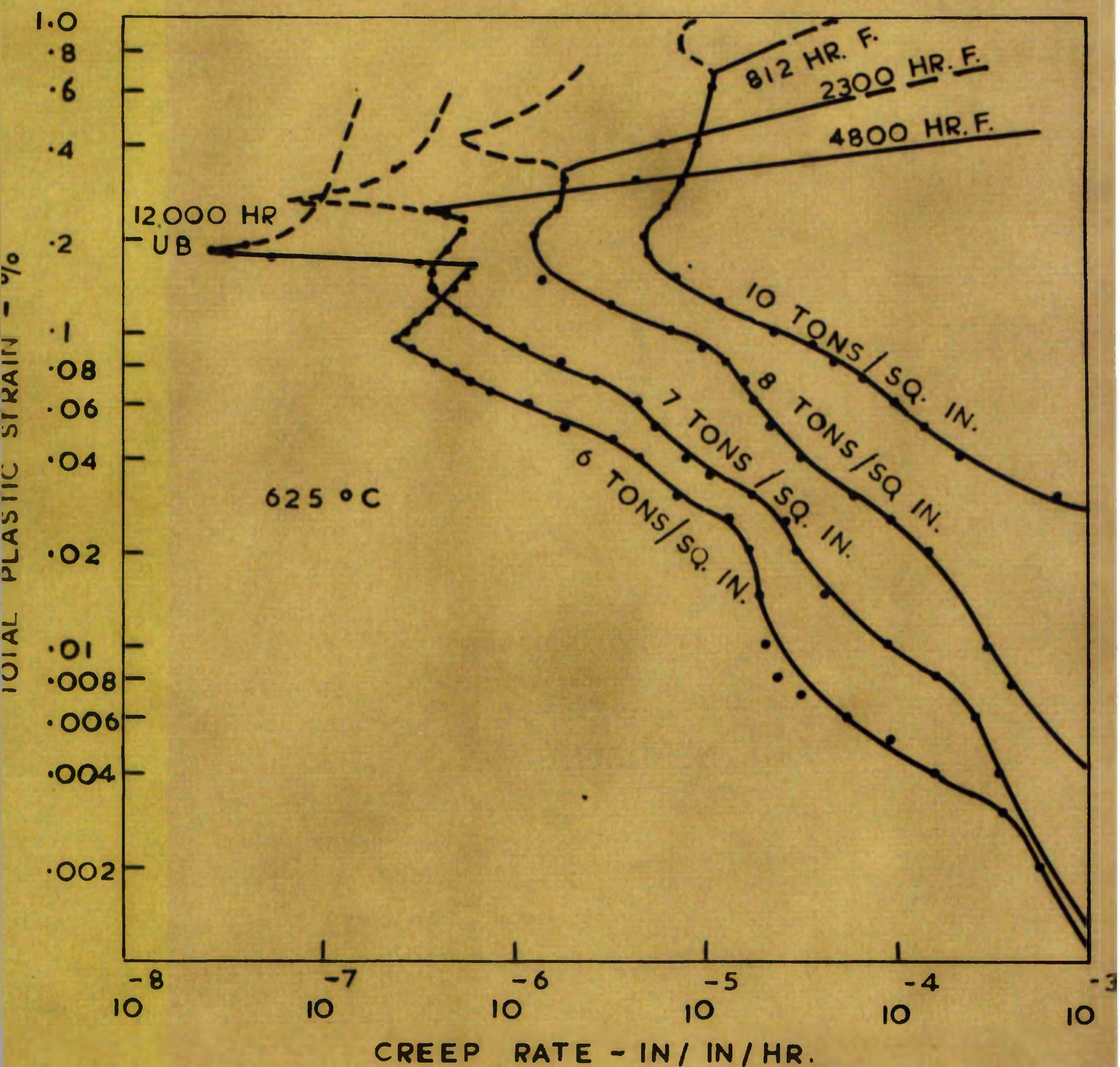
STRAIN - RATE CURVES TO FRACTURE FOR VARIOUS MN. STEELS.

FIG. 57.



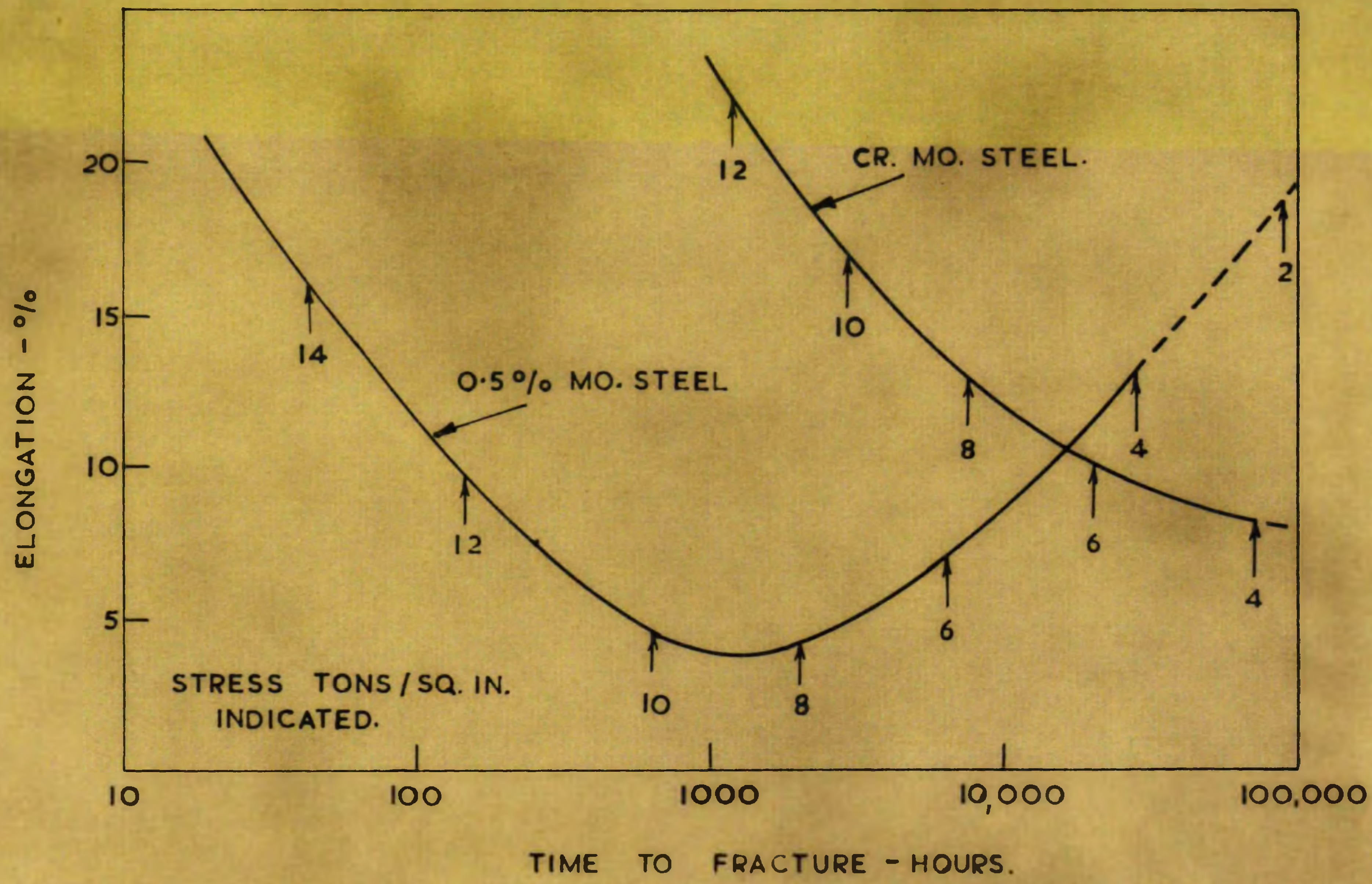
PART CREEP CURVE MO. V. Ti. STEEL. 25.

FIG. 58



STRAIN - RATE CURVES TO FRACTURE MO.V. Ti.STEEL. 25.

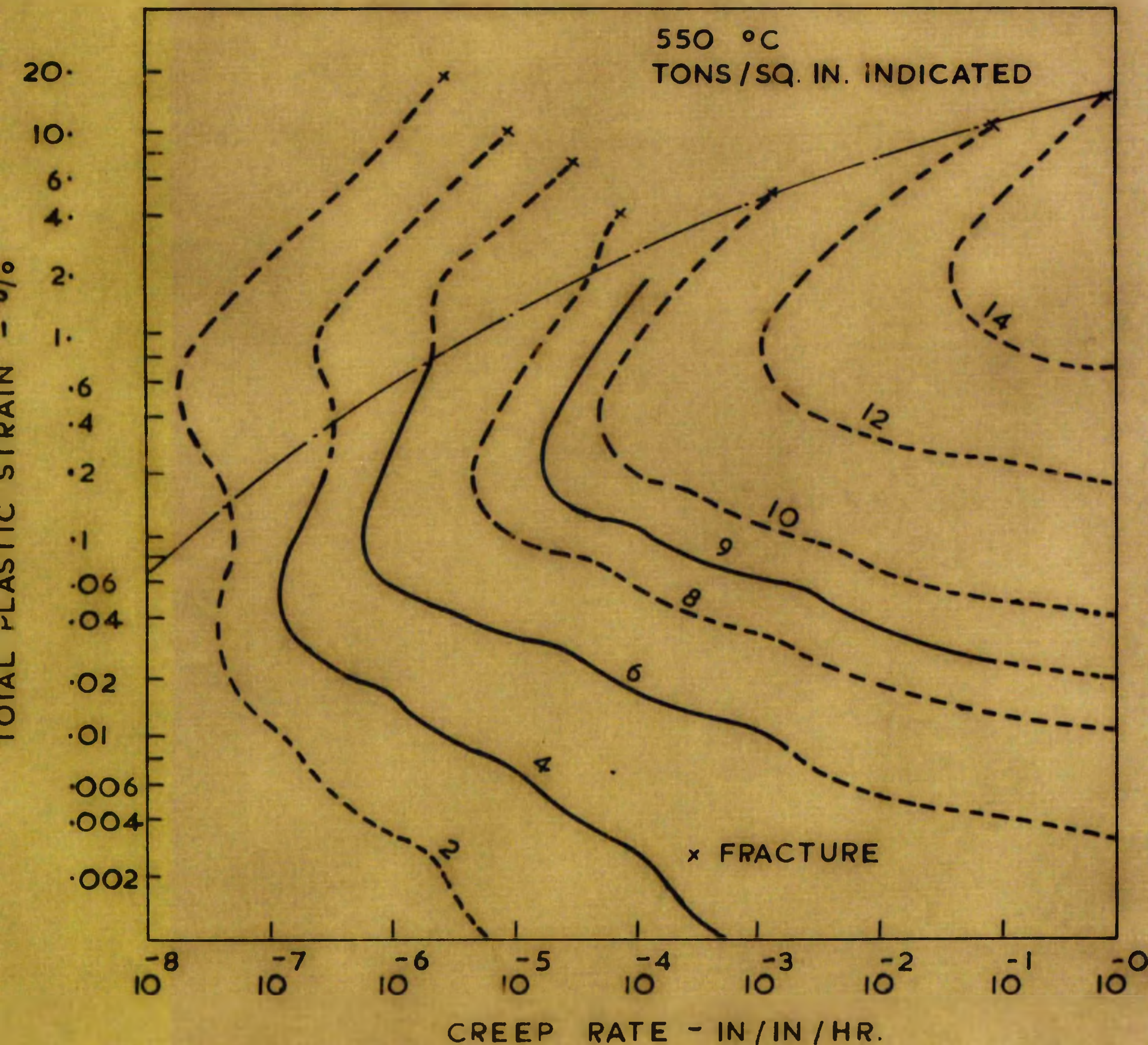
FIG. 59.



ELONGATION IN RUPTURE TESTS AT 550 °C.

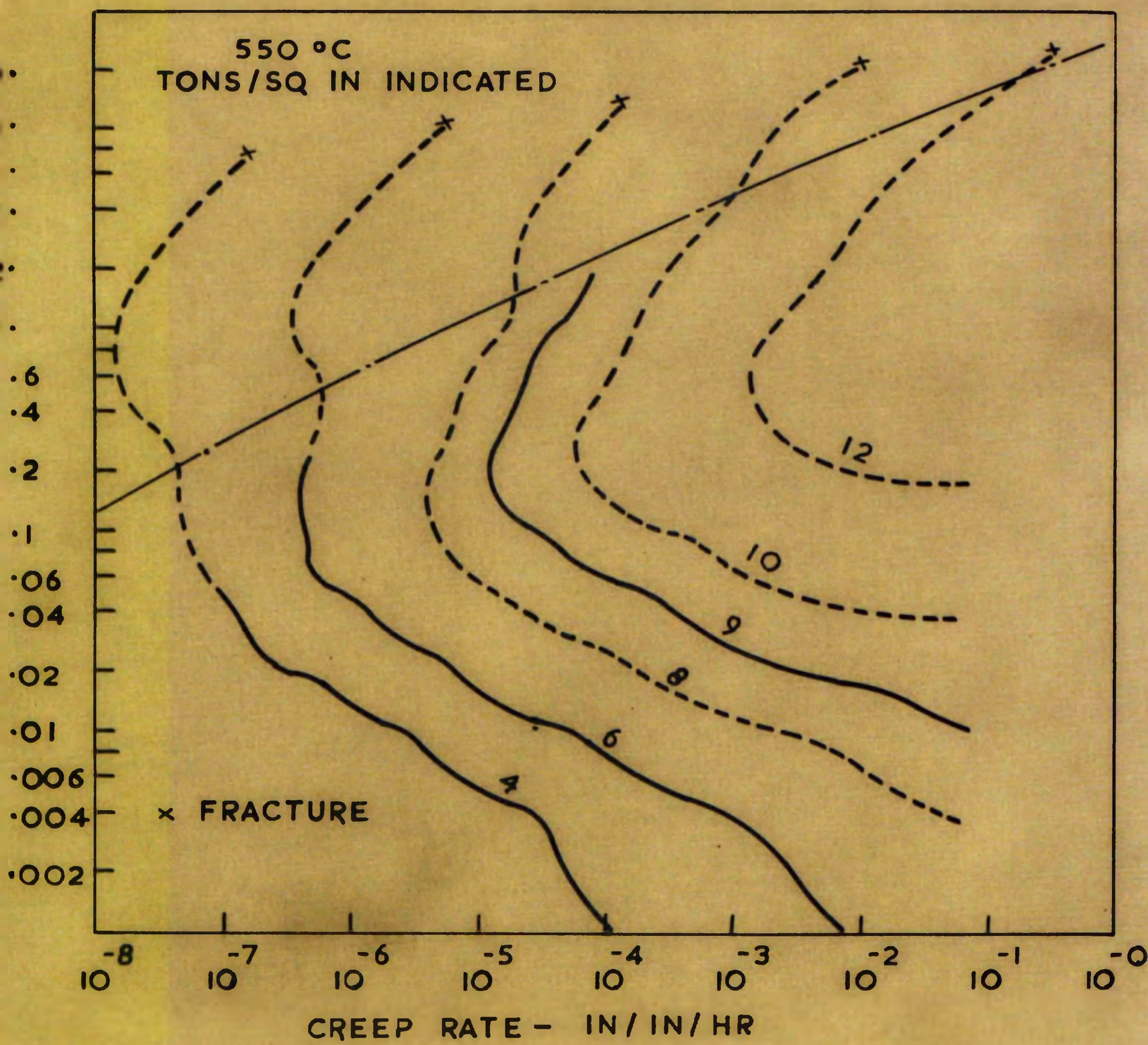
FIG. 60.

FIG. 60



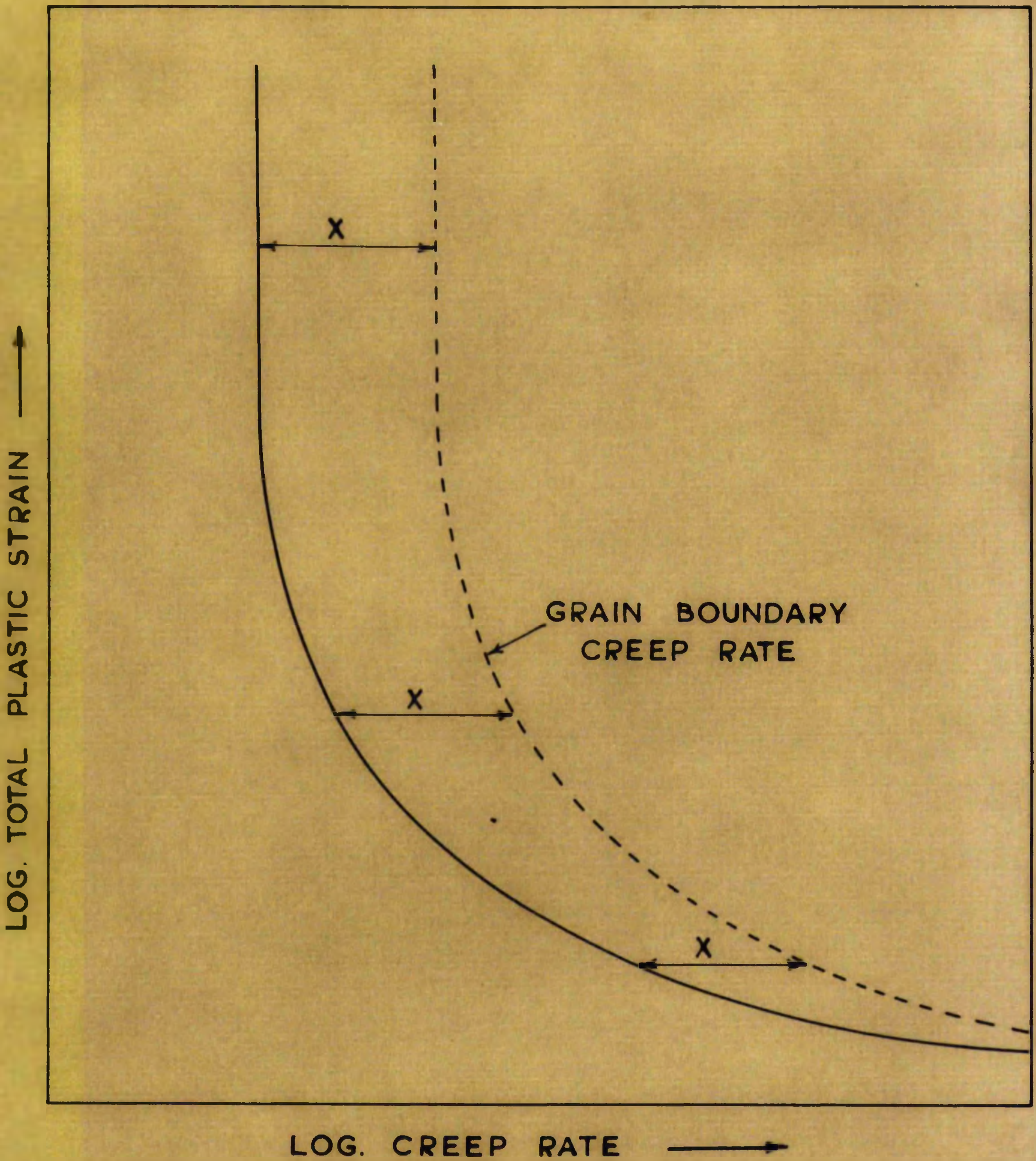
PARTLY TENTATIVE CURVES 0·5 % MO. STEEL.

FIG. 61.



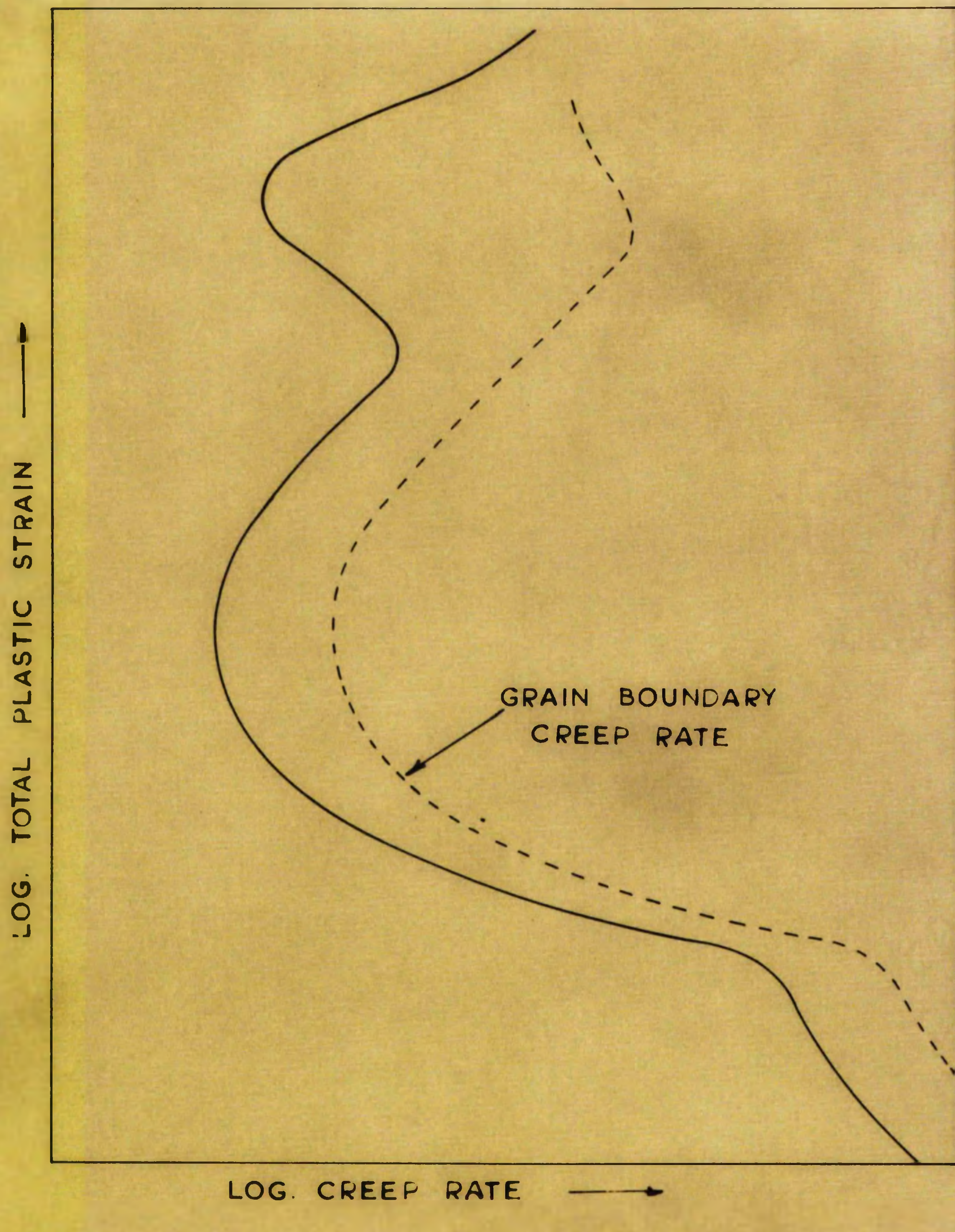
PARTLY TENTATIVE CURVES CR. MO. STEEL.

FIG. 62.



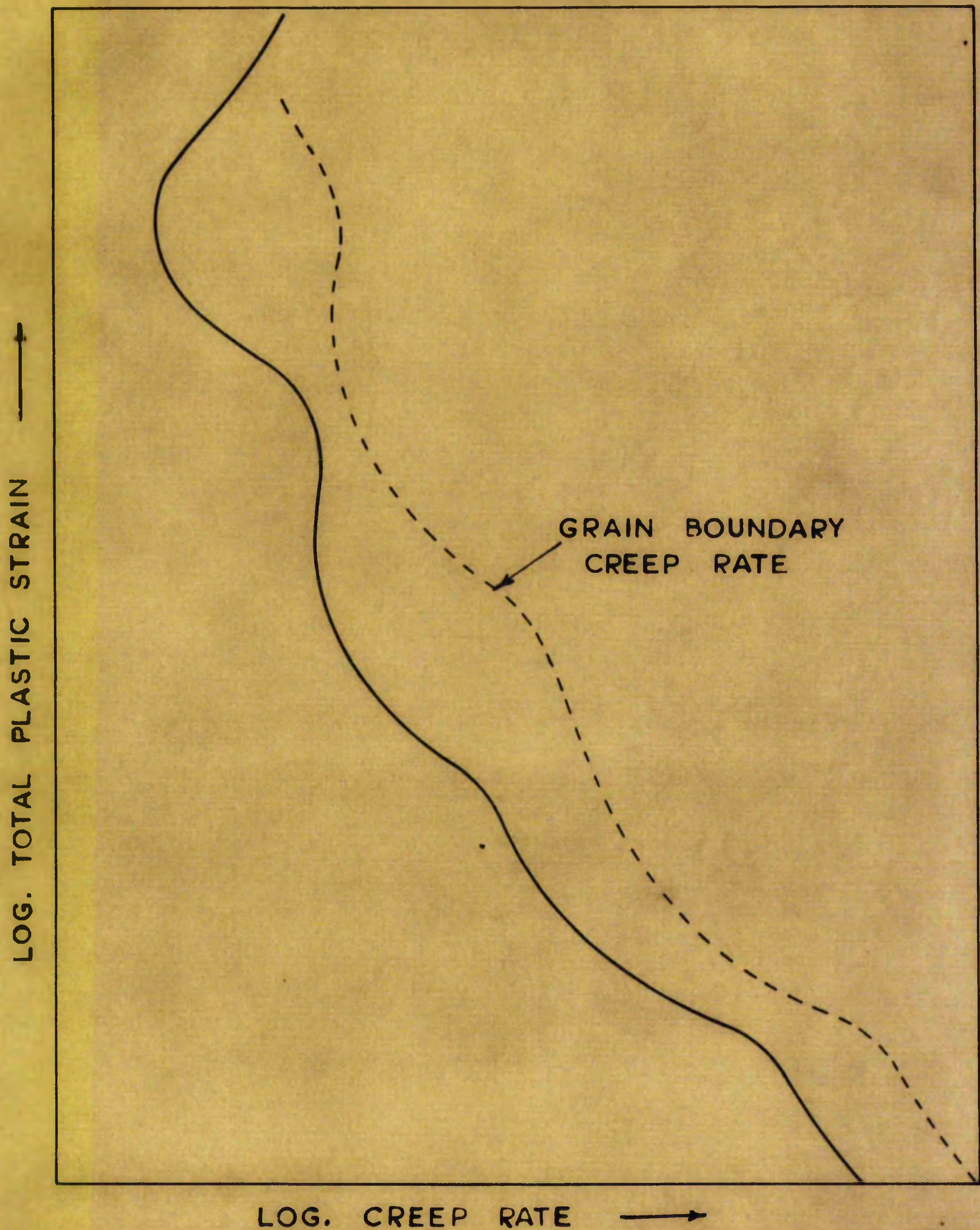
GRAIN BOUNDARY CREEP OF PURE METAL.
(DIAGRAMMATIC)

FIG. 63.



GRAIN BOUNDARY CREEP OF ALLOY WITH ONE SIMPLE AND ONE COMPLEX TRANSITION (DIAGRAMMATIC)

FIG. 64.



GRAIN BOUNDARY CREEP OF ALLOY WITH ONE SIMPLE AND TWO COMPLEX TRANSITIONS.
(DIAGRAMMATIC)

SUMMARY OF D.Sc. THESIS AND ADDITIONAL PAPERS

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A large number of creep tests on various steels were carried out to study the effect of composition, heat treatment, and testing variables on the creep resistance. Some of this work has been published [1,2,4]. It was found however that the order of creep resistance of various steels could be changed by altering the stress or temperature of testing. A similar effect can occur on comparing the same steel in the normalised and in the normalised and tempered condition. The former is always the better at high stresses and /or temperatures but the tempered steel may be the better at low stresses and temperatures. Very long time creep and rupture tests were carried out on 0.5% molybdenum steel, 1% chromium - 0.5% molybdenum steel and on molybdenum-vanadium steel. From these tests an estimate of the creep and rupture strength in 100,000 hours was obtained and the change in rupture ductility with increasing testing time was demonstrated.

The above tests indicated that the creep of commercial alloys was too complex a problem to be easily solved even by the accumulation of a large body of creep test data, and it was decided to approach the problem from a different angle. The effect of individual alloying elements was thus investigated by carrying out true stress-true strain tensile tests on various steels over a range of temperature [3,5]. It was found that, with/

with a simple iron-carbon alloy, a maximum in stress was obtained at about 200°C., and that this maximum was intensified if a little nitrogen (0.005%) was present. A minimum in reduction of area was obtained at about the same temperature. With the addition of either manganese chromium, molybdenum, tungsten or copper to the steel a second maximum in stress and a second minimum in reduction of area was obtained at a temperature higher than 200°C., the temperature depending on the element added. This second strain-age-hardening effect appears to be associated with some form of coherent precipitation of alloy carbide (or copper) in the dislocations formed during straining. It was thus concluded that the first strain-age-hardening effect was due to the precipitation of iron carbide and nitride. It was also found that a powerful carbide former tends to modify the effect of a less powerful one. This was attributed to the affinity of carbon and nitrogen atoms in solution for alloy atoms. Carbon and nitrogen atoms are fixed in interstices adjacent to alloy atoms and thus have less tendency to migrate into the dislocations during straining.

In the belief that similar effects must operate during creep testing a careful reappraisal was made of the shape of the many creep test curves which were available. It was found that unlike those of pure metals or simple solid solutions the creep curves of most alloys showed a rapid deceleration of creep rate over one or more intervals of time. These phenomena were called transitions in creep rate. In a simple iron-carbon alloy/

alloy only one transition in creep rate was obtained. If nitrogen was present two transitions were obtained. If manganese was added to the steel a third transition was obtained. The addition of chromium and molybdenum also resulted in transitions in creep rate. These results are convincing proof that strain-age-hardening phenomena do in fact operate during creep testing.

A simple theory was developed to explain the results obtained, log strain-log creep rate curves being used to present the data. The basic idea of the theory is as follows. Strain-hardening is the most important factor in the initial stage of a creep test, the movement and multiplication of dislocations being so rapid that strain-age-hardening has no time to occur. As the creep rate decreases, the rate of formation of dislocations decreases and the number stationary at any instant increases so that the strain-ageing constituent can precipitate in a few dislocations. When these are fixed other dislocations can pile up giving more time for further ageing. The process is thus self aggravating so that the creep rate slows down rapidly over an interval of time giving a transition in creep rate.

Apart from the steels above mentioned evidence is presented to show that alloys with a face-centred cubic structure (austenitic alloys and nimonic) and hexagonal close-packed alloys (titanium alloys) behave in quite a similar way.

From a family of strain-rate curves at constant temperature or constant/

constant stress it is possible to estimate the strain-rate curves for lower stresses or temperatures and by integration to obtain the ordinary creep curves. In this way extrapolation to long testing times can be made with some theoretical justification. Examples of such extrapolations are given.

Finally it is shown that strain-age-hardening can result in intercrystalline failure with a low extension. It was found that such failures occurred at or near to the beginning of a transition in creep rate. A tentative hypothesis is put forward to account for this phenomenon.

<u>THESIS</u>: "The Effect of Alloying Elements on Creep Behaviour with Particular Reference to Steel"

<u>PAPERS</u>: 1) "Abnormal Creep in Carbon Steels" (Journal I.S.I. April 1947, pp. 501-512).

2) "The Creep Properties of Molybdenum, Chrome-Molybdenum and Molybdenum-Vanadium Steels" (Journal I.S.I. January 1948, pp. 37-80).

3) "An Experimental Study of the Strength and Ductility of Steel at Elevated Temperatures" (A.S.T.M. Special Technical Publication 128, 1953, pp. 184-224).

4) "Some Additional Creep and Rupture Data" (Journal I.S.I. April 1955, pp. 320-336).

5) "The Effect of Alloying Elements on High Temperature Tensile Strength of Normalised Low-Carbon Steel" (Journal I.S.I. May 1957, pp. 21-48).