Lawrence Berkeley National Laboratory
Recent Work

Title
ANALYSIS OF HIGH-TEMPERATURE TRANSIENT CREEP OF PURE METALS AND ALLOYS

Permalink
https://escholarship.org/uc/item/7qx1w349

Author
Amin, Kamal E.

Publication Date
1971-08-01

ANALYSIS OF HIGH-TEMPERATURE TRANSIENT CREEP OF PURE METALS AND ALLOYS

Kamal E. Amin
(Ph. D. Thesis)

August 1971
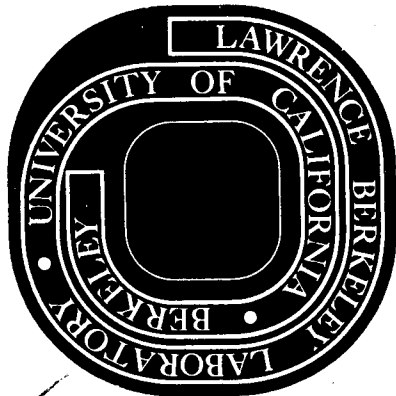
DISCLAIMER

ANALYSIS OF HIGH-TEMPERATURE TRANSIENT CREEP
OF PURE METALS AND ALLOYS

Contents

Kamal E. Amin

Inorganic Materials Research Division, Lawrence Berkeley Laboratory
and Department of Materials Science and Engineering, College of
Engineering, University of California, Berkeley, California

ABSTRACT

It is suggested that transient creep at high temperatures arises princi-
pally as a result of the dispersal of entanglements by the climb mech-
anism. The dispersal of the entanglements is assumed to follow a uni-
molecular reaction kinetics with a rate constant that depends on stress
and temperature in the same way as does the secondary creep rate. The
analysis shows that the strain (ϵ) versus time (τ) relation can be
represented by

$$\epsilon = \epsilon_o + \dot{\epsilon}_s \cdot t + \frac{\beta-1}{K}\left[1 - \exp^{-(K\dot{\epsilon}_s \cdot t)}\right]$$

where ϵ_o is the instantaneous strain on loading, $\dot{\epsilon}_s$ the secondary creep
rate, $K\dot{\epsilon}_s$ the rate constant, and B the ratio of initial to secondary
creep rate. The experimental creep data on several b.c.c and f.c.c
metals and alloys correlate quite well with the proposed mechanism. The
constants β and K were found to be independent of temperature and
stress. The proposed formulation becomes inapplicable for correlating
creep data in polycrystals at low stresses because of the significant
contribution of grain-boundary sliding to the total creep at these
stress levels.

I. INTRODUCTION

1. Typical strain-time curves:

Several common types of high-temperature creep curves are illustrated schematically in Fig. 1, and documented in table I. The creep curve is usually separated into three distinct regions; primary (transient), secondary (steady state) and a later tertiary stage. Most metals and alloys exhibit a reasonably extensive secondary stage, followed by an increasing creep rate over stage III. Stage III is characterized by microfissuring, local plastic deformation and creep rupture. Major differences between the creep curves of various materials, however are noted over the transient stage I. Materials[1-6] in which the substructure pertinent to creep remains substantially constant, enter the steady-state almost immediately after loading (type B). All annealed[1] metals and some alloys exhibit a deccelerating transient creep rate (type A), reflecting the continued formation of a more creep resistant substructure during the transient stage. These same metals and alloys give creep curves of type C when they have been previously crept at a high applied stress[7], or have been cold worked. Their increasing creep rate over stage I is due to recovery of the pertinent substructure to a steady-state one, whereby their steady-state creep rates are independent of any previous mechanical treatment.[8] The secondary creep rate is generally obtained when a balance is reached between the rate of hardening of the developed substructure, and the rate of its thermal recovery under the applied stress.

Type D curve has been observed in certain special dispersed phase alloys[9]. Other types of creep curves were obtained as a result of periodic re-crystallization, or as a result of microstructural changes attending

precipitation hardening, overaging, etc[11].

A review of all pertinent data by Bird, Mukherjee and Dorn[12] reveals that only those metals and alloys that undergo initial straining upon stressing exhibit the usual normal transient, NT, (type A, Fig. 1) stage of creep. This group includes all metals and almost all alloys that obey the semi-empirical relation:

$$\frac{\dot{\varepsilon}_s \cdot K.T}{D.G.b.} = A \; (\sigma/G)^n \tag{1}$$

for which $n \simeq 5.0$, and to a lesser degree some alloys for which $n \simeq 3.3$. In the above equation (1), $\dot{\varepsilon}_s$ is the secondary creep-rate, D: the self diffusivity, G: shear modulus, b, the Burger's vector, and σ the applied stress. Those alloys which do not undergo significant initial straining at $t = 0$ upon application of the stress do not exhibit the usual normal transient stage (NT) of creep. They instead display one of three different types of transient behavior. The few $n \simeq 5$ alloys of this type show an almost negligible initial strain and an extremely brief normal transient (BNT). Alloys of the $n \simeq 3.3$ class which undergo no initial strain show either a brief inverted transient (BIT) over which the creep rate increases towards the steady state or no transient behavior at all (NOT). A few $n \simeq 3.3$ alloys show initial straining and others do not. Some examples of this variant transient creep behavior are given in table 1.

2. Strain - time equations:

Early investigations[13] have shown that:

$$\varepsilon = \varepsilon_0 + f \; (t \; e^{-Qc/RT}) \quad \text{for } \sigma = \text{constant,} \tag{2}$$

over both the transient and the steady-state stages of creep at high temperatures. In this equation, ε is the total creep strain, which is

the sum of the initial strain ε_0 upon stressing at $t = 0$ and the creep strain up to time t, σ, is the applied stress, Q_c is the apparant activation energy for creep, and RT is the usual gas constant times the absolute temperature in K. Q_c as defined by

$$\left\{\frac{\partial \ln \dot{\varepsilon}}{\partial (-1/RT)}\right\} = \left\{\frac{\ln (\dot{\varepsilon}_2/\dot{\varepsilon}_1)}{\frac{1}{R}\left(\frac{1}{T_1} - \frac{1}{T_2}\right)}\right\}_\sigma \qquad (3)$$

was shown to be insensitive to the applied stress, strain, grain-size, temperature etc. It is always slightly different from the activation enthalpy of diffusion:

$$Hd = \frac{\partial \ln D}{\partial (-1/RT)} \qquad (4)$$

For many materials, the creep strain is found to follow Andrad's[17] empirical equation, viz

$$\varepsilon - \varepsilon_0 = \dot{\varepsilon}_s \cdot t + Bt^{1/3}, \qquad (5)$$

where $\dot{\varepsilon}_s$, is the steady-state creep rate and B is a constant. This formulation often appears to be in fair agreement with high temperature creep data, for its parameters can be adjusted to fit most of the high timperature data spanning both the transient and steady states. One objection to the universal application of Andrade's equation concerns its prediction of infinite initial creep rates. Although it is admittedly difficult to measure the initial creep-rates accurately, many carefully conducted experiments suggest that physically they must be finite.

An alternative formulation that has been employed is:

$$\varepsilon - \varepsilon_0 = \dot{\varepsilon}_s \cdot t + \varepsilon_T \left(1 - e^{-rt}\right) \qquad (6)$$

where ε_T, is the total transient strain for $t = \alpha$, and r is the ratio of the transient creep rate to the transient creep strain. This relation

was first suggested on a purely empirical basis by McVetty[18]. Later it has been applied by Garofalo[19], Conway and Mullikin[20], and Evans and Wilshire[21]. Also several authors[22,23,24] have arrived at expressions analogous to equation 6, from detailed consideration of specialized dislocation models for the rates of multiplication and annihilation of dislocations during high temperature creep.

It is the purpose of this investigation to provide a unified analysis for high-temperature creep that will incorporate the effect of both temperature and stress on the shape of the transient creep curve, and provide a physical picture of the primary stage of deformation, Stage I. The investigation will first summarize the pertinent substructural changes that accompany transient creep and will emphasize the role of dislocation climb in the dispersal of entanglements. This will be followed by the development of a unimolecular reaction rate kinetic approach to describe the transient creep behavior. The validity of the presented approach will be confirmed by correlation with experimental data.

II. SUBSTRUCTURAL CHANGES DURING TRANSIENT CREEP

The decreasing creep rates over the normal transient stage of creep, under constant values of the independent variables of stress and temperature, must be ascribed to the development of more creep resistant substructures over this stage. Since normal transient creep is contingent upon the occurrence of initial straining, normal transient creep rates are the result of changes in substructure from that which is produced immediately upon stressing to that which pertains to the steady-state. The substructures which are formed upon initial straining at creep temperatures closely resemble those which are developed during strain

hardening over stage III deformation at lower temperatures[25,26,27].
Most of the dislocations are arranged in rough cellular patterns, the
walls of which are composed of dislocation entanglements. Some cells
contain a few randomly meandering dislocations in their interior. Etch
pitting[25,28] and electron microscopy studies[27,28,29] have shown that
dislocations rearrange themselves during the early part of transient
creep as a result of the extra degree of freedom resulting from dislo-
cation climb which becomes facile at high temperatures. During the early
portions of transient creep the cellular structure remains observable,
but in many localized areas dislocations rearrange to become either
randomly distributed or to develop partially formed subboundaries.
Electron transmission[27,28,29] microscopy shows that at this time loosely-
knit subgrain boundaries are being developed through dislocation untang-
ling and redistribution of dislocations. No abrupt changes in the dis-
location substructure occur during the transient stage. Rather, as strain
increases, larger numbers of roughly-formed subgrain boundaries develop,
cells become diffuse as dislocations climb and glide away, and those
subboundaries which are already formed become better defined and more
tightly knit. A subgrain structure emerges gradually and develops clearly
over the latter stages of the transient creep.

The entanglements disperse and adjacent cells coalesce so as to produce
more sharply delineated subgrains which have a volume about eight
times that of the original cells over the secondary stage of creep, and
an invariant substructure persists which depends only on the magnitude
of the applied stress. It consists of subgrains separated from one
another by low-angle boundaries within which there exists a few isolated
randomly meandering dislocations.

The substructural changes that take place during normal transient creep

are extremely complex. This complexity is clearly revealed by the de-
tailed investigations of Clauer, Wilcox and Hirth[28]. Undoubtedly these
substructural changes are significant to the transient creep rate. Such
changes in the different substructural details over the transient stage
are undoubtedly interrelated. For example, the dispersal of the
entanglements, the build-up of low angle boundaries, the changes in
density of dislocations and the alterations in misorientations across
low-angle boundaries, etc. are not mutually independent. Among these
various details only the dislocation density has been placed on a
quantitative basis.

Several investigators have attempted to follow the change in dislocation
density with strain over the transient stage of creep[25,28,30,31]. Ad-
mittedly to accomplish this with satisfactory accuracy is extremely
difficult, particularly in view of the major changes that take place
in the geometrical distribution of dislocations. Furthermore, it is
not immediately apparent how the density of dislocations should be
partitioned among those entangled in the disappearing cell walls, those
adding to the growing low angle boundaries, and those that are meandering
about more or less randomly. Early in the transient stage it is customary
to count all dislocations including those in the entangled cell walls
and those in the body of the cells despite the fact that each group
might contribute differently to the transient creep rate. As the
transient creep rate approaches that for the steady state, it is custom-
ary to count only those dislocations that are within the subgrains.
However Clauer, Wilcox and Hirth[28] have shown that moving tilt-boundaries
form a significant part of the substructure during transient creep of
single crystals of molybdenum, and consequently the dislocations in the
sub-boundaries must be considered.

Results on the variation in the density of dislocation during creep

within the subgrains are presented in Fig. 2 as taken from reference 12.

The plot shows the change in $(\dot{\varepsilon}/\dot{\varepsilon}_s.\rho.b^2)$ as a function of ε for Cu single

crystals and polycrystalline Fe-3.1% Si, and as a function of $(\varepsilon-\varepsilon_o-\dot{\varepsilon}_s t)/\varepsilon_T$

for molybdenum single crystals, where ε_o is the initial strain, ε_T is the

total transient strain, and ρ is the density of dislocations. This plot

shows the $(\dot{\varepsilon}/\dot{\varepsilon}_s.\rho.b^2)$ seems to decrease over the primary stage. If creep

were controlled by the motion of jogged screw dislocations, it might be

expected that $\dot{\varepsilon}$ would increase linearly with ρ. The effect of dislocation

density of climb-controlled creep, however, need not be linear because

with high densities of dislocations the climb time would be shorter

but the area swept out by the dislocations would also be smaller. The

effect of these counteracting influences on $\dot{\varepsilon}$ has not yet been accurately

formulated.

Although the state of knowledge concerning substructure is admittedly

unsatisfactory, the available experimental data suggest that dislocation

densities decrease only slightly during the dispersal of the entangle-

ments in transient creep. The decrease in creep rate during normal

transient stage might be assumed to indicate that strain-hardening is

taking place. Strain-hardening however has the implication of increased

density of dislocations. It therefore appears that the term recovery

strengthening due to dispersal of entanglements by dislocation climb is

preferable.

III. ANALYSIS

The substructural changes taking place during transient creep are so

complex as to prelude any detailed mechanistic analysis of the process

at this time. Although the density of dislocations decreases over the

transient stage of creep, it alone cannot account for the much greater

percentage decrease in the creep rate. The major substructural change,

as pointed out by Gupta and Strutt[25] centers about the dispersal of en-

tanglements, but the relation between such dispersal and the transient

creep rate is not yet understood. Undoubtedly all substructural changes

during high-temperature transient creep are interrelated and dependent

upon the dislocation climb mechanism. It is therefore suggested that

transient as well as steady state creep is controlled by the rate of

climb of dislocations. This is consistent with the validity of equation

2. Furthermore, it is also assumed that the reactions taking place

obey the laws of unimolecular kinetics. This assumption will be verified

a posteriori.

Assuming that dislocation climb kinetics control transient creep, the

rate constant can be given in terms of the temperature and stress as $K\dot{\varepsilon}_s$,

where $\dot{\varepsilon}_s$ is the secondary state creep rate as controlled by climb and K

is a constant. As shown by Webster, Cox and Dorn[32], the unimolecular

can be formulated as:

$$\frac{d(\dot{\varepsilon}-\dot{\varepsilon}_s)}{dt} = - K \dot{\varepsilon}_s.(\dot{\varepsilon} - \dot{\varepsilon}_s) \tag{7}$$

By integrating, one obtains

$$\ln (\dot{\varepsilon}-\dot{\varepsilon}_s) = K \dot{\varepsilon}_s.t + \text{constant} \tag{8}$$

At t = o $\quad \dot{\varepsilon} = \dot{\varepsilon}_i \quad$, substituting in (8) for the constant gives,

$$\dot{\varepsilon} - \dot{\varepsilon}_s = (\dot{\varepsilon}_i - \dot{\varepsilon}_s) e^{-K \dot{\varepsilon}_s.t.} \tag{9}$$

Integrating once again gives:

$$\epsilon = \dot{\epsilon}_s \cdot t - \frac{\dot{\epsilon}_i - \dot{\epsilon}_s}{K \dot{\epsilon}_s} \cdot e^{-k \cdot \dot{\epsilon}_s \cdot t} + \text{constant} \qquad (10)$$

At $t = 0$ $\qquad \epsilon = \epsilon_0$, substituting in (10) for the constant gives

$$\epsilon = \epsilon_0 + \dot{\epsilon}_s \cdot t + \left(\frac{\dot{\epsilon}_i - \dot{\epsilon}_s}{K \dot{\epsilon}_s}\right)\left[1 - e^{-K \dot{\epsilon}_s \cdot t}\right] \qquad (11)$$

where $\dot{\epsilon}_i$ is the initial creep-rate at time = 0, ϵ_0 is the initial strain upon stressing, and

$$\frac{(\dot{\epsilon}_i - \dot{\epsilon}_s)}{K \dot{\epsilon}_s} = \epsilon_T \qquad (12)$$

is the limiting transient strain. Since the transient and steady state stages have the same energy of activation for creep, and also the same kinetics of reaction as implied by equation 2, it follows necessarily that:

$$\dot{\epsilon}_i = \beta \dot{\epsilon}_s \qquad (13)$$

where β = constant > 1. Therefore, the total transient strain:

$$\epsilon_T = \frac{(\dot{\epsilon}_i - \dot{\epsilon}_s)}{K \dot{\epsilon}_s} = \frac{\beta - 1}{K} \qquad (14)$$

where β and K are constants independent of stress and temperature. According to the above analysis, there must apply:

i) The initial strain upon stressing ϵ_0, is athermal and depends on the original state of the metal or alloy and the value of σ/G where σ is the stress and G is the shear modulus;

ii) The ratio of the initial to secondary creep rates, β, is a constant independent of stress, temperature and initial strain;

iii) The rate of dispersal of the entanglements depends on the same function of stress and temperature as does the secondary creep rate;

iv) For a given metal or alloy the transient strain, ϵ_T, is constant independent of stress or temperature; and

v) for a given metal or alloy there exists a universal high temperature transient and steady-state creep curve of the form

$$\epsilon - \epsilon_0 = f(\dot{\epsilon}_s \cdot t) = \phi \ (\sigma/G)^n \frac{DGb}{kT} \cdot t \qquad (15)$$

where the function f is derived from equation 2 mentioned earlier in the text. The relation depicted in 15 is an extension of 3 which now incorporates the effects of stress as well as temperature on the shape of transient creep curve.

IV. CORRELATION OF EXPERIMENTAL DATA

A careful survey and re-analysis was made of all the experimental results on high temperature transient creep where the pertinent data on diffusivity and shear modulus were also available. A summary of all the examples that have been analyzed is given in Table II. The validity of equation 11 is shown by the creep curves depited in Fig. 3 for experimental data on, Al[9] Ag[32], purified Fe[31], Fe containing C[33], Cu[34] plain carbon steel,[35] Mo single crystals,[14] low carbon Ni,[36] Pb[37], and stainless steel[29]. In general, the data spanned a range of stress levels and, in the case of data on Pt, low carbon Ni, Al, Ag, and Fe containing carbon, a range of temperatures as well. The creep curves clearly reveal that regardless of temperature and stress, the data for each material fall well on to a single curve which agrees well with the solid theoritical curve of equation 11. Figure 4 shows the initial strain, ϵ_0, as a function of σ/G for the cases that were analyzed. These data refer to the modulus adjusted athermal stress-strain curves at high temperatures. The secondary creep rates $\dot{\epsilon}_s$, for the metals and alloys in question are

shown in Fig. 5, where ln ($\dot{\epsilon}_s$kT/DGb) is plotted as a function of ln(σ/G).
The curves agree fairly well with similar curves that were presented in
a recent review by Bird, Mukherjee and Dorn[14]. In general, the plots
clearly fall into the category of creep as controlled by the dislocation
climb mechanism.

A typical example of the Universal creep curve for the case of poly-
crystalline Ni is shown by the datum points of Fig. 6a. Regardless of
stress all data fall well on to a single curve. The solid line refers
to the theoritical expression given by Eq. 11. An even more convincing
proof of the validity of the universal creep curve is shown in Fig. 6-b
for Nb, where the experimental data cover ranges of temperature as well
as stress and the secondary creep rates vary by two orders of magnitude.
The remaining cases that were examined were about equally consistent
and illustrated the good agreement of the theory with the experimental
results. The stress dependence of $\dot{\epsilon}_i$ and $\dot{\epsilon}_s$ is shown in Fig. 7 where
($\dot{\epsilon}$kT/DGb) is plotted against σ/G in dimensionless units. The datum
points refer to creep of Ni by Parker[38] at one temperature and five
stress levels and to creep of Nb by Brinson and Argent[39] for a series
of temperature or stress levels. Fig. 7 illustrates that the initial
creep rates, $\dot{\epsilon}_i$, for a series of stresses and temperatures are greater
than the secondary creep rates, $\dot{\epsilon}_s$, by a constant factor β. Both $\dot{\epsilon}_i$
and $\dot{\epsilon}_s$ are dependent on the substructural details that develop and hence
are functions of stress. As shown in Fig. 7, their difference is a
constant regardless of stress and temperature and is not a function of
state.

Values for the transient strain ϵ_T were deduced from curves in Fig. 3
and Fig. 6. The rate constant K for different cases was determined at
one-half of the transient strain where equation 11 gives K = ln (2/$\dot{\epsilon}_s$.t).

The calculated values for the constants K, β and ϵ_T are shown in Table 2.
Whereas the transient strain ϵ_T exhibits rather pronounced variations
form one case to another, the rate constant K seems to be somewhat less
variable. Undoubtedly both depend on significant substructural details.
Up to the present, however, no consistent trends in variations of K or β
have been uncovered relative to effect of grain size, stacking-fault
energies, or any other pertinent structural details.

Evans and Wilshire[21] working on polycrystalline Fe found that Eq. 11
becomes inaccurate for creep at low stresses. Similar discrepancies,
which are observed in the creep of austentic stainless steel, have been
discussed by Webster, Cox and Dorn[40] to arise from the increased con-
tribution of grain-boundary sliding to the total creep rate at the low
applied stresses. The data for stainless steel by Garofalo analyzed
here apply only to the higher stress levels where the contribution of
grain boundary sliding is less than 0.04 of the total creep strain. When
grain-boundary sliding pervails, K$\dot{\epsilon}_s$ is no longer the pertinent rate
constant, since it refers to a creep mechanism resulting from climb
along, and, under the circumstances, ϵ_T is not expected to be a constant
independent of stress. Once grain boundary sliding is better understood
quantitatively, this factor might also be taken into consideration so
as to account completely for high-temperature transient creep at low
stresses in polycrystals as well.

V. CONCLUSIONS

The following conclusions are drawn from the present work.

1. It has been shown that the creep equation

$$\varepsilon = \varepsilon_o + \dot{\varepsilon}_s \cdot t + \varepsilon_T [1 - \exp^{(-k\dot{\varepsilon}_s t)}]$$

can be derived from the assumption that transient creep follows a first order kinetic reaction rate theory, with a rate constant that depends on stress and temperature in the same way as does the secondary creep rate.

2. It is suggested that high-temperature transient creep results from substructural modifications due to stress and diffusion controlled climb of dislocations.

3. For a given metal or alloy there exists a universal high-temperature transient and steady-state creep curve that incorporates the effect of temperature as well as stress in the general form:

$$\varepsilon = \varepsilon_o + f(\dot{\varepsilon}_s t) = \phi \left[(\sigma/G)^n \cdot \frac{DGb}{kT} \cdot t \right] + \varepsilon_o.$$

4. For a given metal or alloy the transient strain ε_T is constant, independent of stress or temperature, when no significant grain-boundary sliding takes place.

5. The initial creep-rate, $\dot{\varepsilon}_i$ for a given metal or alloy is a constant multiple, β, of the steady-state creep rate $\dot{\varepsilon}_s$, regardless of stress and temperature.

6. The initial strain ε_o depends on the original state of the metal or alloy and the value of σ/G.

7. In polycrystals at low stress levels, $K\dot{\varepsilon}_s$ is no longer the pertinent rate constant and the effect of grain-boundary sliding needs yet be taken into consideration.

REFERENCES

1. T. Hazlett and R. D. Hansen, Trans. ASM $\underline{47}$, 508 (1954).

2. B. M. Alexander, M. H. Dawson, and H. P. Kling, J. Appl. Phys. $\underline{22}$, 439 (1951).

3. A. K. Mukherjee and J. E. Dorn, J. Inst. Metals $\underline{93}$, 397 (1964-65).

4. J. G. Harper and J. E. Dorn, Acta Met. $\underline{5}$, 654 (1967).

5. H. Laks, C. D. Wiseman, O. D. Sherby, and J. E. Dorn, J. Appl. Mech. $\underline{24}$, 207 (1957).

6. E. M. Howard, W. L. Barmore, J. D. Mote, and J. E. Dorn, Trans. AIME $\underline{227}$, 1061 (1963).

7. L. Raymond and J. E. Dorn, Trans. AIME $\underline{230}$, 560 (1964).

8. O. D. Sherby, T. A. Trozera and J. E. Dorn, Proc. ASTM $\underline{56}$, 784 (1956).

9. G. A. Webster and B. J. Pearcy, Met. Sci. J. $\underline{1}$, 97 (1967).

10. E. R. Gilbert and D. E. Munson, Trans. AIME $\underline{233}$, 429 (1965).

11. J. Glen, J. Iron and Steel Inst. $\underline{189}$, 333 (1958).

12. J. E. Bird, A. K. Mukherjee, and J. E. Dorn, Quantitative Relation Between Properties and Microstructure, edited by D. G. Brandon and A. Rosen (Israel University Press, Jerusalem, 1969) p. 255.

13. J. E. Dorn, J. Mech. Phys. Solids $\underline{3}$, 85 (1954).

14. Private communication from G. A. Webster and J. E. Dorn.

15. S. Karashima, H. Oikawa, and T. Watanabe, Trans. AIME $\underline{242}$, 1703 (1968).

16. N. Brown and D. R. Lenton, Acta Met. $\underline{17}$, 669 (1969).

17. E. N. Andrade, Proc. Roy. Soc. (London) $\underline{A84}$, 1 (1910).

18. P. G. McVetty, Mech. Eng. $\underline{56}$, 149 (1934).

19. F. Garofalo, Fundamentals of Creep and Creep-rupture in Metals, (MacMillan, New York, 1965).

20. J. B. Conway and M. J. Mullikin, Trans. AIME $\underline{236}$, 1496 (1966).

21. W. J. Evans and B. Wilshire, Trans. AIME $\underline{242}$, 1303 (1968).

22. N. S. Akulov, Acta Met. $\underline{12}$, 1195 (1964).

23. J. M. Li, Acta Met. $\underline{11}$, 1269 (1964).

24. G. A. Webster, Phil. Mag. $\underline{14}$, 775 (1966).

25. V. P. Gupta and P. R. Strutt, Canad. J. Phys. $\underline{45}$, 1213 (1967).

26. J. D. Livingston, Acta Met. 10, 229 (1962).

27. F. H. Hammad and W. D. Nix, Amer. Soc. $\underline{59}$, 94 (1966).

28. A. H. Clauer, B. A. Wilcox and J. P. Hirth, Acta Met. $\underline{18}$, 381 (1970).

29. F. Garofalo, C. Richmond, W. F. Domis and F. V. Gemingen, Joint International Conference on Creep, Inst. Mech. Eng. (London) 1963. p. 1-31.

30. G. R. Barrett, W. D. Nix and O. D. Sherby, Trans. ASM $\underline{59}$, 3 (1966).

31. F. Garofalo, L. Zwell, A. S. Keh, and S. Weissmann, Acta Met. $\underline{9}$, 721 (1961).

32. R. P. Carrecker and R. W. Guard, Report No. 55-RL-1414, (General Electric Co., Schenectady, New York, 1955).

33. D. McLean and K. F. Hale, Symposium on Structural Processes in Creep, Iron and Steel Inst. (London) 1961, p. 19.

34. P. Feltham and J. D. Meakin, Acta Met. $\underline{7}$, 614 (1959).

35. P. Feltham, Proceedings of the 2nd International Conference on Rheology, edited by V. Harrison (Butterworth, London, 1959) p. 266.

36. P. Coldren and J. Freeman, ASD Technical Report 61-440, Department of Chemical and Metallurgical Engineering, University of Michigan, Ann Arbor (1961).

37. P. R. Carreker, Jr., J. Appl. Phys. $\underline{21}$, 1289 (1950).

38. E. R. Parker, Trans. ASM $\underline{50}$, 85 (1958).

39. G. Brinson and B. Argent, J. Inst. Metals $\underline{91}$, 293 (1963).

40. G. A. Webster, A. D. Cox, and J. E. Dorn, Met. Sci. J. 3, 221 (1969).

Table I. Metals and alloys that display different transient.

Alloy or Metal	Thermomechanical History	Initial Strain ε_o	n	Type of Transient	Creep Curve Type	Ref.
Ni	Annealed	> 0	5.0	NT*	A	1
Au	Annealed	0	4.7	BNT**	B_1	2
AgMg	Annealed	< 0.001	5.7	BNT	B_1	3
Fe – 3.9% Mo		0	> 4.0	BIT***	B_3	15
Al	Annealed	0	< 4.0	NOT****	B_2	4
Al – 3.1% Mg	Annealed	0	3.5	NOT	B_2	5
β – CuZn		< 0.004	3 – 3.4	NOT	B_2	16
Ag_2 Al	Annealed	0	3.6	BIT	B_3	6
Ni	Cold worked to $\varepsilon = 0.034$	0		NOT	C	1
Al	Crept at Higher Stress	0		NOT	C	7
Niomonic Alloy	Annealed	0		NOT	D	14

*NT: Normal Transient Creep.

**BNT: Brief Normal Transient.

***BIT: Brief Inverse Transient.

****NOT: No Transient.

Table II. Summary of analyzed transient creep data.

Metal (Polycrystalline)	Crystal Structure	Grain Size (mm)	ε_T	K	β	Chemical Composition	Ref
Al	F.c.c.	0.21	0.105	82	9.56		8
Ag	F.c.c.	0.017	0.013	99	2.28		31
Alpha-Fe	B.c.c.		0.053	92.5	5.9	C, N <50 p.p.m., total impurity <100 p.p.m.	30
α - Fe	B.c.c.	0.025	0.017	82.5	2.4	C: 0.054, N: 0.0005, O: 0.0005, Si: 0.013, S: 0.0026, P: 0.001, Mn: 0.001, Al: 0.006	32
Cu	F.c.c.	0.03	0.0085	277	3.35	Fe: 40 p.p.m, S: 8, Ag: 6, Ni: 5, Pb: 4	33
Plain C - Steel	B.c.c.	austenitic grain size 0.21-0.104	0.024	116	3.77	C: 0.23, Si: 0.13, Mn: 0.60, S: 0.036, P: 0.036, Cr: 0.092 Ni: 0.17 and Cu: 0.125%	34
Mo [S] Single crystal	B.c.c.		0.052	173	10.00		12
Ni	F.c.c.		0.042	326	12.9		37
low c - Ni	F.c.c.		0.036	330	12.88	C: 0.01, Mn: 0.03, Si: 0.1	35
Nb	B.c.c.	0.17 - 0.36	0.032	173.2	10.2	C: <0.015, O: 0.076, Ta: <0.25	38
Pt	F.c.c.		0.01	138.6	2.39	Traces of Fe, Pd <0.01 wt %	36
Stainless Steel	F.c.c.	0.09	0.065	34.60	3.26		28

Figure Captions

Fig. 1. Common Types of Creep Curves.

Fig. 2. Density of Dislocations within Subgrains vs. Strain in Transient Creep.

Fig. 3. Experimental Data for Equation 11, as $\varepsilon - \varepsilon_o$ versus $\dot{\varepsilon}_s \cdot t$.

Fig. 4. Initial Strain ε vs. σ/G.

Fig. 5. Stress Dependence of Secondary Creep Rate.

Fig. 6. The Universal Creep Curve.

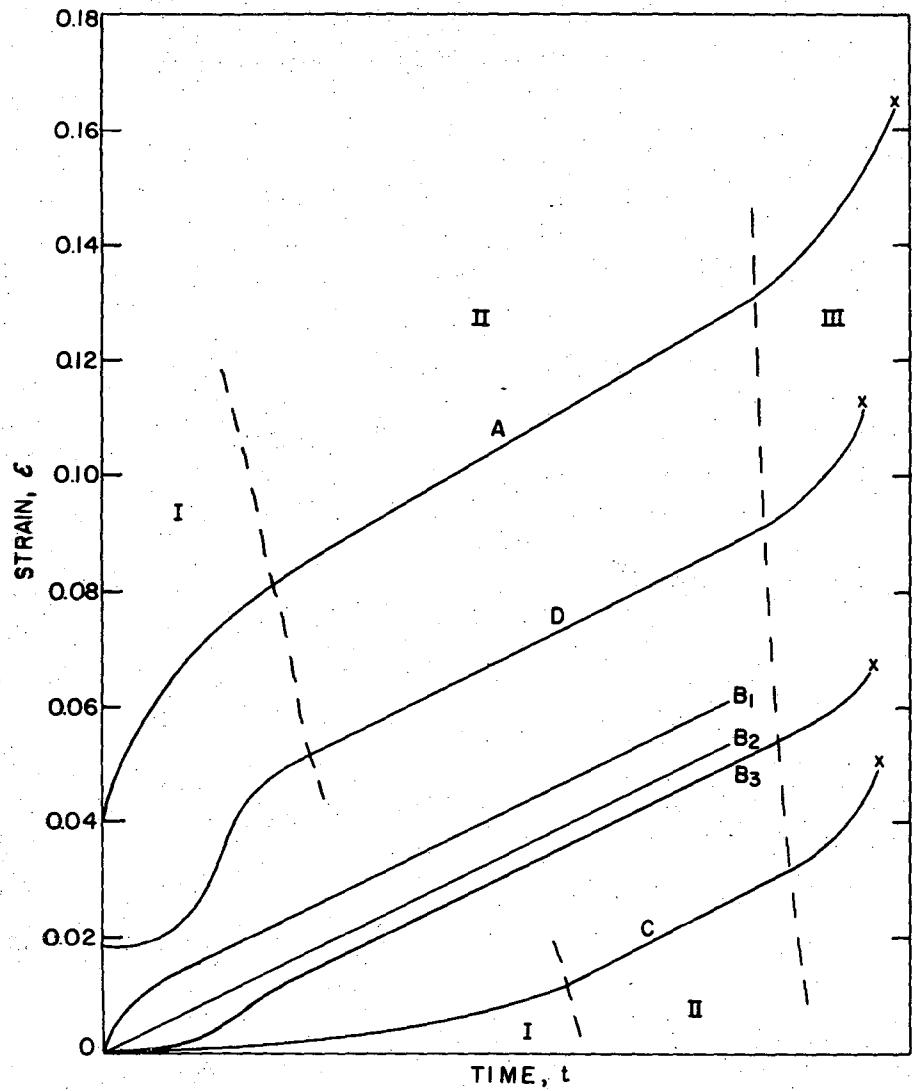Fig. 7. The Stress Dependence of Initial and Secondary Creep Rate for Nb and Ni.

XBL 697-950

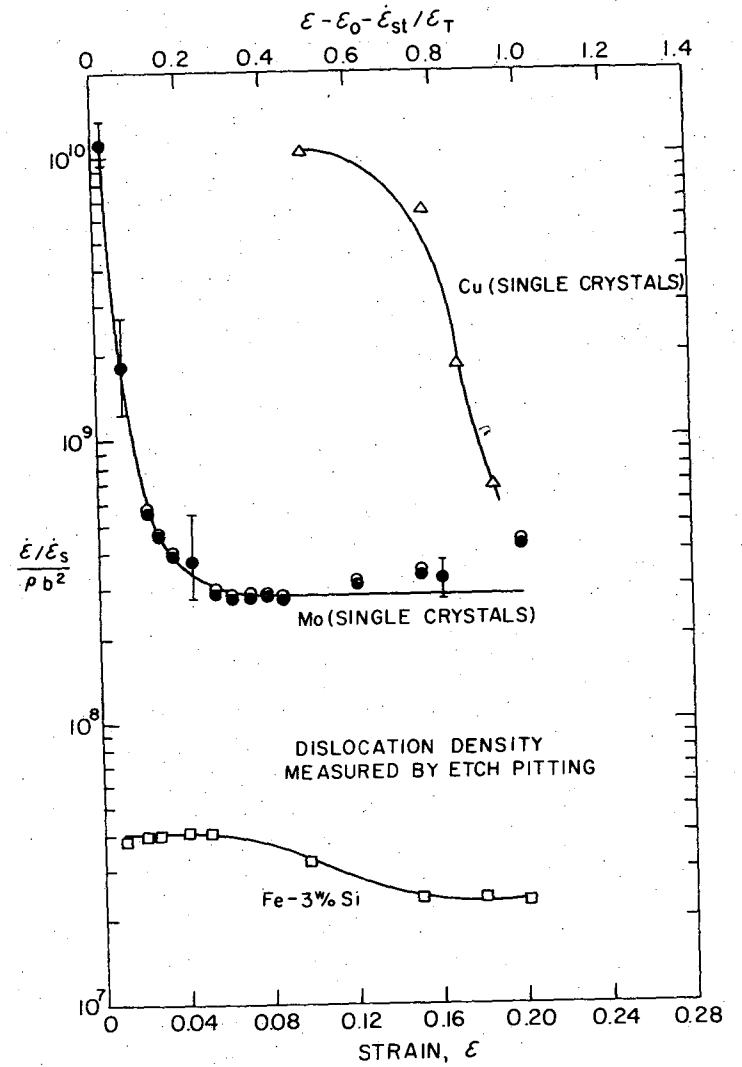FIG. 1 TYPICAL CREEP CURVES.

XBL 697-918

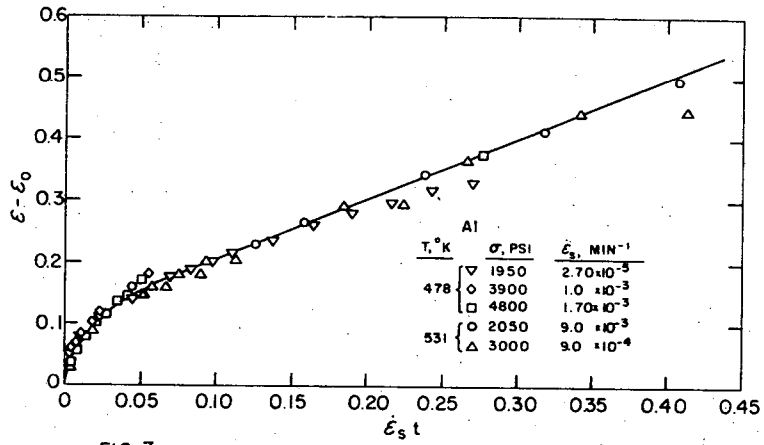FIG. 2 VARIATION OF DENSITY OF DISLOCATIONS WITHIN SUBGRAINS WITH STRAIN IN TRANSIENT CREEP.

FIG. 3a

Al
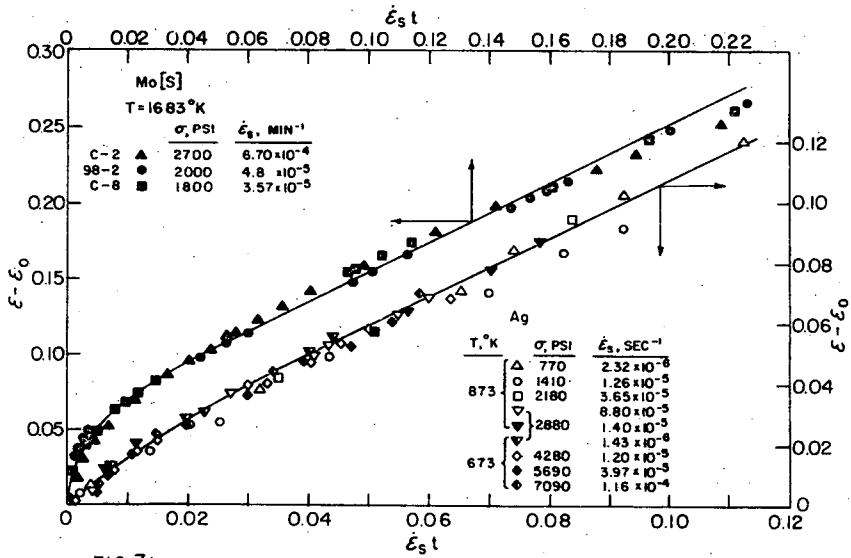
T, °K		σ, PSI	$\dot{\varepsilon}_s$, MIN^{-1}
478	▽	1950	2.70×10^{-5}
	◇	3900	1.0×10^{-3}
	□	4800	1.70×10^{-3}
531	◯	2050	9.0×10^{-3}
	△	3000	9.0×10^{-4}

FIG. 3b

Mo [S]
T = 1683 °K

		σ, PSI	$\dot{\varepsilon}_s$, MIN^{-1}
C−2	▲	2700	6.70×10^{-4}
98−2	●	2000	4.8×10^{-5}
C−8	■	1800	3.57×10^{-5}

Ag

T, °K		σ, PSI	$\dot{\varepsilon}_s$, SEC^{-1}
873	△	770	2.32×10^{-6}
	◯	1410	1.26×10^{-5}
	□	2180	3.65×10^{-5}
	▽	2880	8.80×10^{-5}
	▼		1.40×10^{-5}
673	◇	4280	1.43×10^{-6}
	◆	5690	1.20×10^{-5}
	◆	7090	3.97×10^{-5}
			1.16×10^{-4}

XBL 701-182

FIG. 3c

Fe CONTAINING C

T, °C		σ, t/m²	$\dot{\varepsilon}_s$, HR^{-1}
700	◯	1	1.50×10^{-3}
550	□	1.5	1.76×10^{-2}
	▽	3	2.22×10^{-2}
	△	5.6	1.78×10^{-4}

Fe (PURIFIED)
T = 600 °C

	σ, kg/cm²	$\dot{\varepsilon}_s$, HR^{-1}
●	176	7.0×10^{-4}
▲	500	4.0×10^{-4}

FIG. 3d

Cu, T = 550 °C

	σ, kg/mm²	$\dot{\varepsilon}_s$, SEC^{-1}
■	1.75	1.32×10^{-6}
□	1.38	1.66×10^{-6}
△	2.2	3.16×10^{-6}
▲	2.88	9.23×10^{-6}
▽	2.70	7.60×10^{-6}
◯	3.38	1.9×10^{-5}
●	4.09	4.6×10^{-5}

Pt

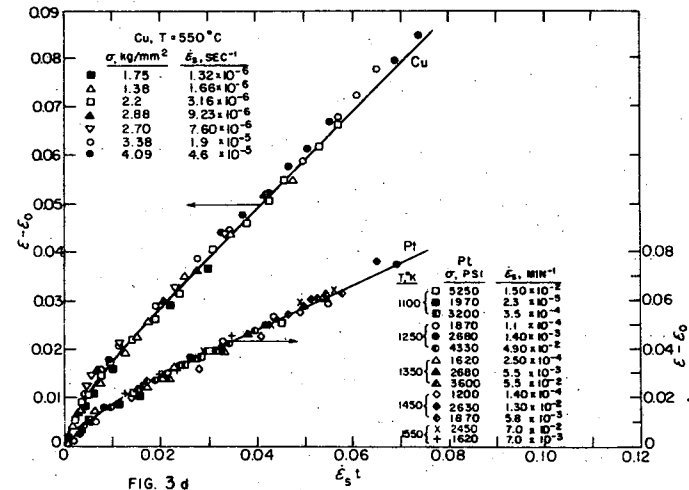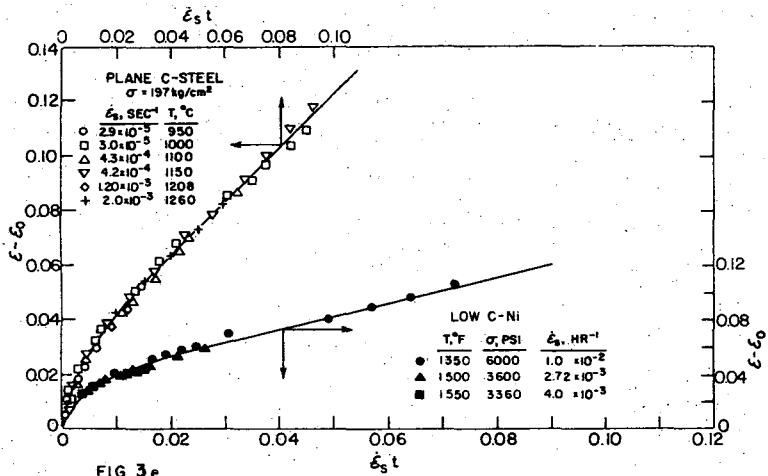| T, °K		σ, PSI	$\dot{\varepsilon}_s$, MIN^{-1}
1100	◻	5250	1.50×10^{-2}
	■	1970	2.3×10^{-5}
	□	3200	3.5×10^{-4}
1250	◇	1870	1.1×10^{-4}
	●	2680	1.40×10^{-4}
	◯	4330	4.90×10^{-4}
1350	△	1620	2.50×10^{-4}
	▲	2680	5.5×10^{-3}
	▲	3600	5.5×10^{-3}
1450	◆	1200	1.40×10^{-4}
	●	2630	1.30×10^{-3}
	●	1870	5.8×10^{-3}
1550	X	2450	7.0×10^{-3}
	+	1620	7.0×10^{-3}

XBL 701-181

FIG. 3e

FIG. 3f

XBL 701-183
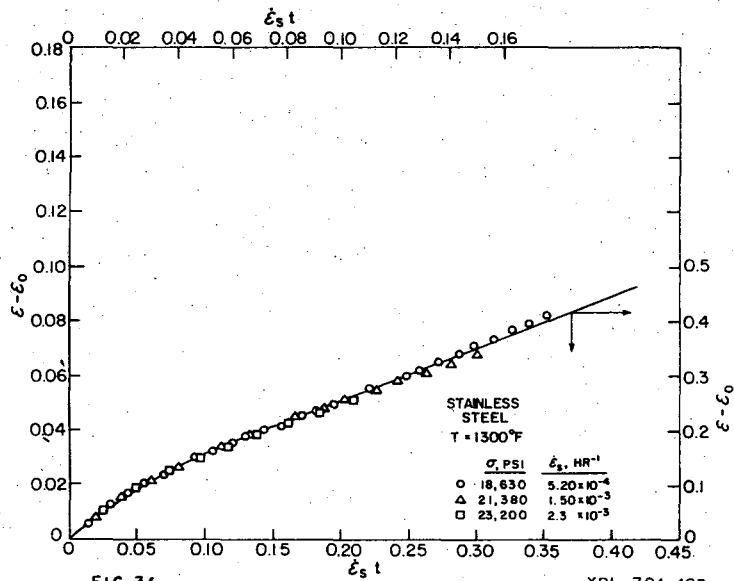
FIG. 3 EXPERIMENTAL DATA FOR EQUATION (5) AS $\epsilon - \epsilon_0$ vs. $\dot{\epsilon}_s t$

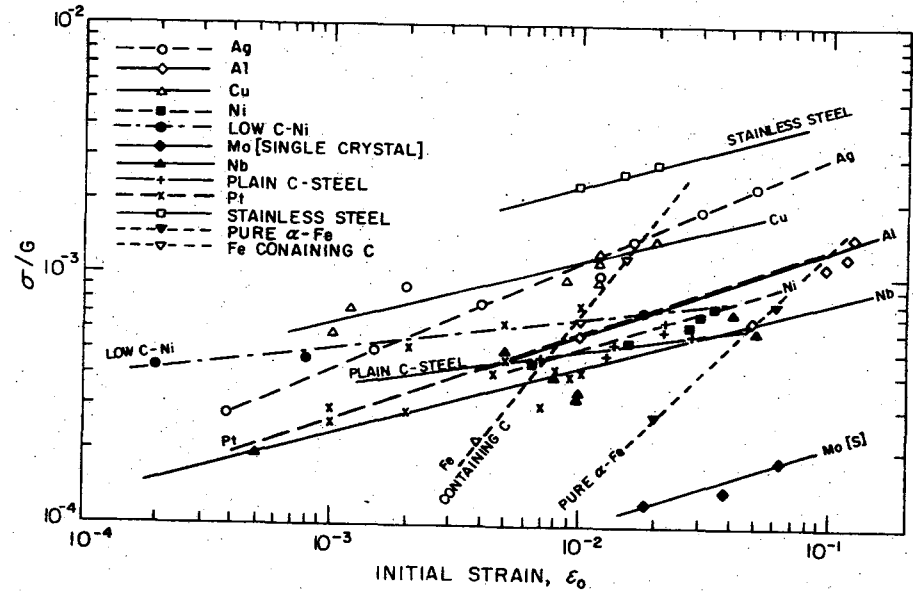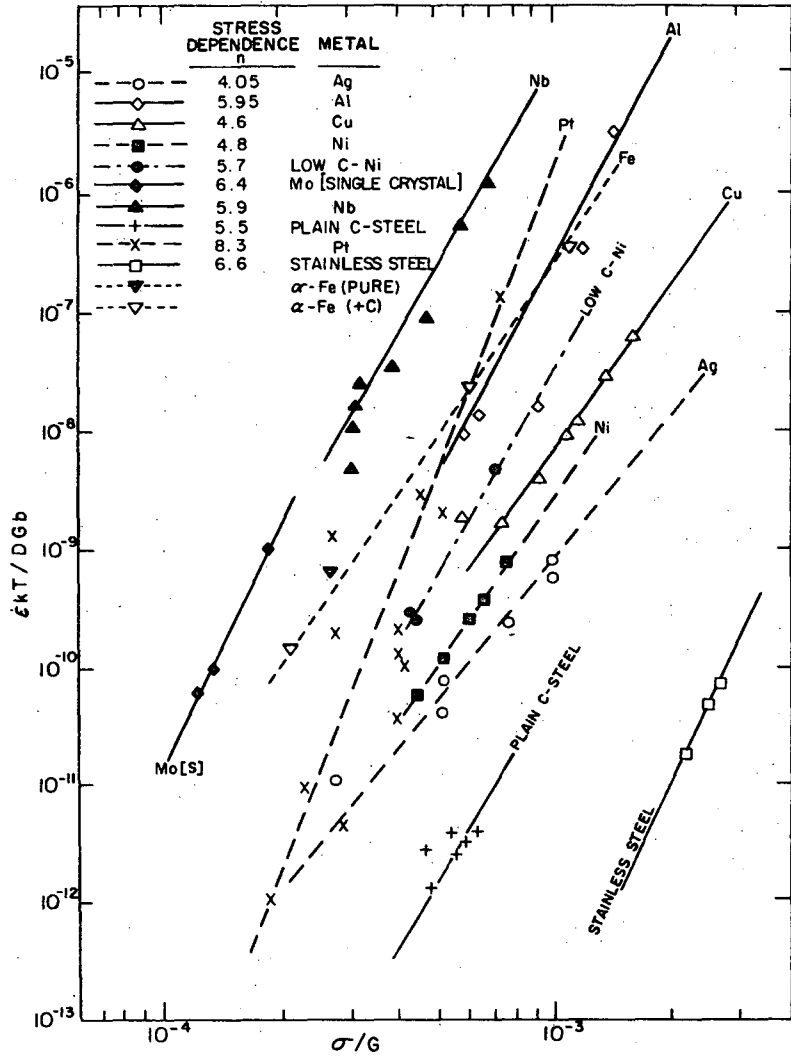FIG 4. INITIAL STRAIN, ϵ, vs. σ/G.

XBL 701-178

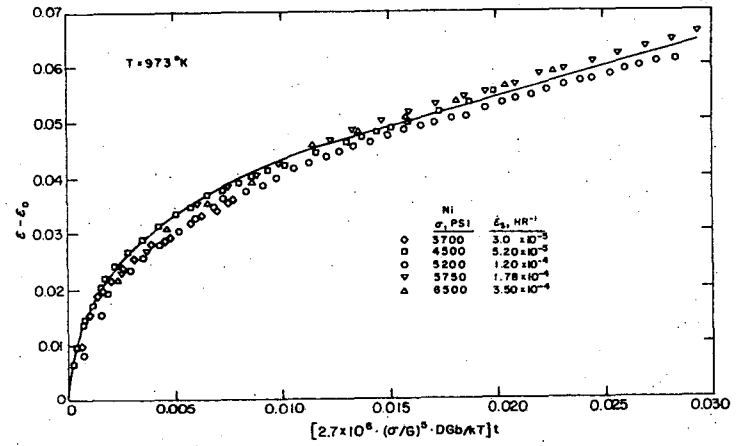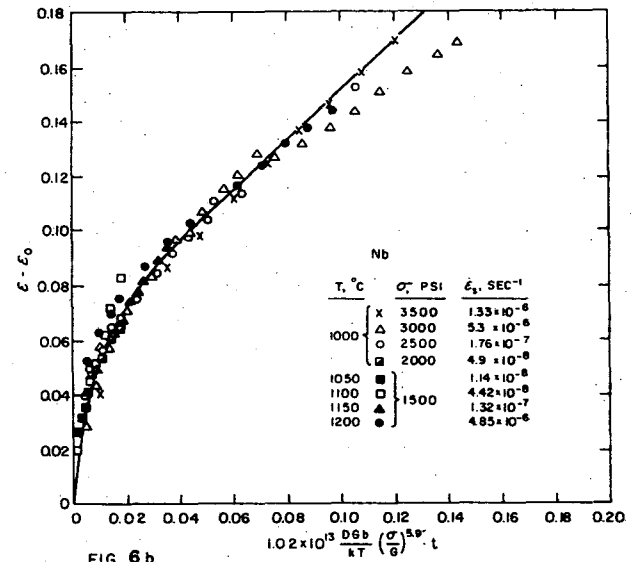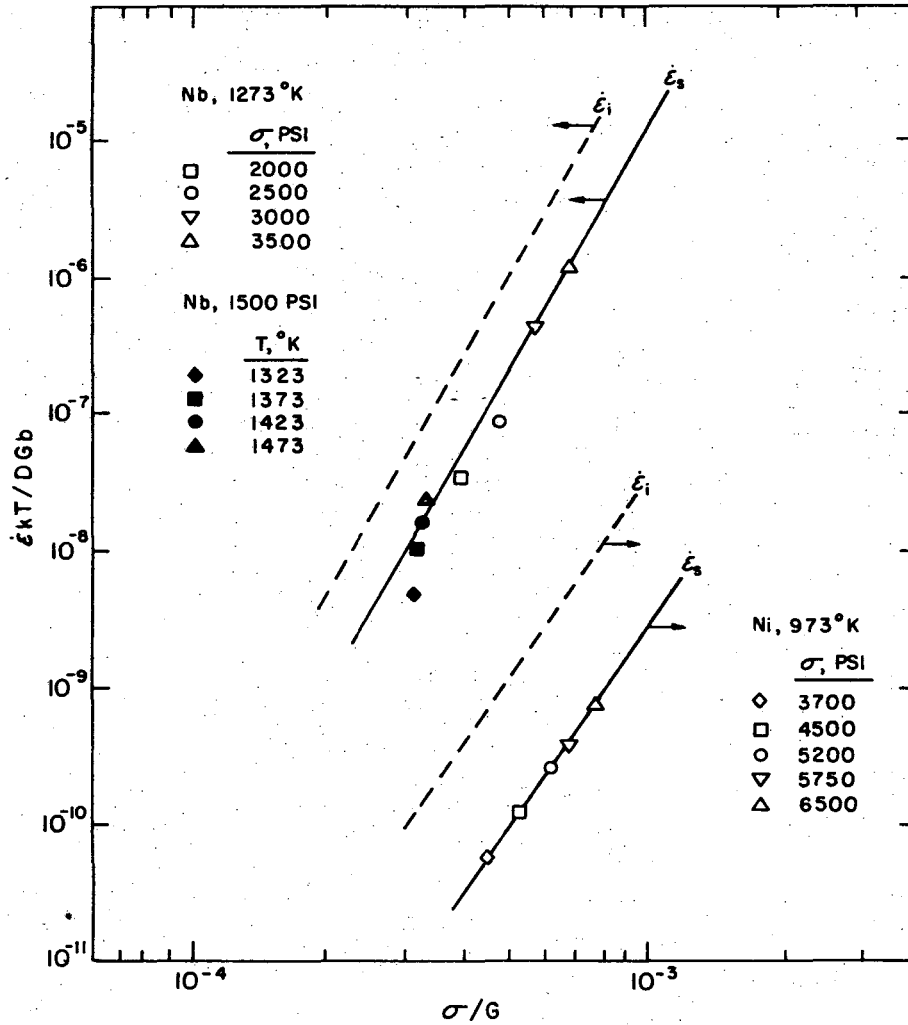FIG. 5 STRESS DEPENDENCE OF SECONDARY CREEP RATE.

XBL 701-180

FIG. 6 a

FIG. 6 b

FIG. 6 THE UNIVERSAL CREEP CURVE.

XBL 701-184

FIG. 7 THE STRESS DEPENDENCE OF INITIAL AND SECONDARY
CREEP RATE FOR Nb AND Ni.

XBL 701-179