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September, 1965

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ABSTRACT

The creep behavior of alpha iron was investigated over the range

of temperatures from 375° to 1150°K. Apparent activation energies for

creep, obtained by the effect of sudden changes in temperature on the

creep rate, revealed the presence of four distinguishable regions:

I. Below 480°K the apparent activation energy decreased
rapidly as the stress was increased.

II. Between 480° and about 775°K the apparent activation
energy decreased slightly with increasing stress and
was insensitive to temperature and strain; and when
extrapolated to zero stress gave the value of 64,000 cal/mole
which agreed well with that for self diffusion in
ferromagnetic iron.

III. Over the range of the Curie transformation (775°
to 1045°K) the apparent activation energy increased
rapidly with increased temperature and reached a maximum
value of over 130,000 cal/mole just below the Curie
temperature (1045°K).

IV. At the Curie temperature the apparent activation
energy decreased abruptly to about 90,000 cal/mole and
increased as the temperature was increased from 1045°
to 1150°K.

The creep behavior of α-Fe in Regions II and III was found to

correlate well with a model where the creep rate is controlled by the

nonconservative motion of jogged screw dislocations. The anomalous increase in the apparent activation energy for creep in Region III was found to be in harmony with the known decrease in the free activation energy for self diffusion over the Curie transformation range. Furthermore, the creep rates in Regions II and III were found to increase with stress due not only to the effect of stress on the activation energy but also to an increase in density of mobile dislocations. The evidence suggests that pipe diffusion along moving dislocations is significant over the lower temperatures of Region II.

INTRODUCTION

Although the creep properties of iron and steel have been of principle interest to metallurgists for some time and an extensive literature is now available, most of the published information has not produced much toward understanding the fundamental dislocation processes controlling the creep behavior of these materials.

Whereas the activation energy for high temperature creep of metals usually agrees well with that for self diffusion, Sherby, Orr and Dorn[1] deduced from the data of Tapsell and Clenshaw[2] that the activation energy for creep of Armco iron is about 78,000 cal/mole and more recently Lytton and Sherby[3] reviewed the data of Jenkins and Mellor[4] and found an activation energy for creep of alpha iron of 80,000 cal/mole, both values being substantially greater than the activation energy for self diffusion in alpha iron. Because of the limited data suitable for purposes of identifying the controlling dislocation mechanisms of creep in alpha iron and the unusually high activation energies for creep quoted above it was considered desirable to reinvestigate in detail the creep of alpha iron by more reliable techniques which would provide sufficient data for determination of the creep mechanisms.

EXPERIMENTAL TECHNIQUE

Creep specimens were prepared from 3/8" x 3/4" iron bar stock of the following composition by weight percent: 0.001 C; 0.0120 O; 0.001 N; 0.004 S and 0.003 P. The as received bars were cold rolled to a thickness of 0.100", annealed under argon for 30 minutes at 1113°K, cold rolled to a final thickness of 0.063", and machined into tensile specimens having a gage length section 0.250" wide and 1.70" long. Finally the specimens were recrystallized under argon at 1113°K. Specimens to be crept above 1060°K were recrystallized at 1173°K. Both recrystallization treatments gave the same reproducible equiaxed ASTM No. 4 grain size (vide Fig. 7). Creep testing was conducted in machines fitted with Andrade-Chalmers arms that maintained a constant stress to within ± 0.2% of the reported values. Deformation over the specimen gage section was sensed with a linear differential transformer and recorded autographically. as a function of time. The strains deduced from these measurements were sensitive to ± 5 x 10^{-5}. During creep the specimens were contained in an argon filled chamber which was immersed in a temperature controlled molten tin bath. Creep temperatures, as measured by thermocouples attached to the specimen, were maintained to within ± 0.25°K of the reported values. For activation energy determinations rapid changes in temperature of about 12°K were obtained within 30 seconds by direct resistance heating of the creep specimen, and maintained to ± 1°K of the reported values. Observation of structural details of specimens before and after creep tests were made by electrolytic polishing and etching in acqueous ammonium persulfate.

EXPERIMENTAL RESULTS

A typical example of the determination of apparent activation

energies, Q, by the effect of small changes in temperature is illustrated

in Fig. 1. Q is defined by

$$Q = \frac{\partial \ln \dot{\varepsilon}}{\partial (-\frac{1}{RT})} \approx \frac{\ln \frac{\dot{\varepsilon}_2}{\dot{\varepsilon}_1}}{\frac{1}{R}(\frac{1}{T_1} - \frac{1}{T_2})} \tag{1}$$

where $\dot{\varepsilon}$ is the tensile strain rate, T is the absolute temperature and

R is the gas constant. Also documented in Fig. 1 are the calculated

values of the apparent activation energies. Apparent activation energies

determined in this manner were found to be independent of strain; for

any given temperature and stress the same activation energy was obtained

throughout the primary and secondary stages of creep. The usual scatter

range of about ± 8% was primarily due to deviations of about ± 1°K in

the measured temperature. The mean values of the apparent activation

energies for various stresses and temperatures are summarized in Table I;

in general they appear to depend sensitively on the temperature and much

less sensitively on the stress.

The apparent activation energies, with their scatter bands identified,

are plotted as a function of temperature alone in Fig. 2. Although these

data scatter somewhat because the effect of stress on the apparent

activation energies was not taken into consideration here, they nevertheless

clearly reveal the existence of four uniquely different temperature

regions of creep:

Region I. Below about 480°K the creep rate decreased rapidly with

increasing strain and finally approached zero creep rate. Whereas the apparent activation energy in this range appeared to increase rapidly with increasing temperature, it is more likely that this observation results from the effect of stress (vide Table I) on the apparent activation energy. The creep mechanism over Region I was not subjected to detailed scrutiny in this investigation.

Region II. For the stresses that were employed, only primary creep took place over Region II; however, the creep rate appeared to be approaching a finite steady state value at the higher strains.

The apparent activation energy over all of Region II appears to increase slightly with an increase in temperature. The higher temperature tests, however, were conducted at lower stresses and, as will be shown later, the observed trend is attributable to the effect of stress and not temperature on the apparent activation energy.

Region III. Near the beginning of the Curie transformation, Q increases ever more rapidly with an increase in temperature and reaches a maximum value at the Curie temperature. It is significant to note that this trend is in nominal agreement with that previously reported for the apparent activation energy for creep of Ni by Landon, Lytton, Shepard and Dorn.[5] It is therefore suggested that the anomalous hump in the apparent activation energy for the creep of Ni is also associated with effects arising from the Curie transformation.

Region IV. Passing through the Curie temperature Q decreases abruptly, following which it then increases somewhat as the temperature is further increased.

In order to ascertain how the apparent activation energies might depend on the stress, the same data employed in Fig. 2 are replotted as Q versus σ in Fig. 3, without regard to the test temperature. As previously mentioned, over Region I, Q decreases very rapidly with an increase in stress. In contrast, however, Q over Region II appears to decrease mildly and linearly with an increase in stress. Although these data exhibit nominal scatter, the trend of Q with σ appears to be real. Using the method of least squares suggests that over Region II

$$Q = 67,000 - \sigma v \qquad \text{cal/mole} \qquad (2)$$

where σ is the applied tensile stress in psi and v has the value of 0.138. The linear decrease in Q with stress, independent of T over Region II will be discussed later in terms of creep as a result of the non-conservative motion of jogged screw dislocations. The data in Fig. 3 seem to indicate that there is no systematic dependency of Q on σ over Region III and IV. In these Regions, however, Q is extremely sensitive to the test temperature so that the dependency of Q on σ, if small (like that over Region II), will be completely masked by the great and variable change in Q with T.

It is proposed that over Regions II and III

$$\dot{\varepsilon} = A \{\sigma, \text{struct.}\} e^{-\frac{H_D}{RT}} \sinh \frac{\sigma v}{RT} \qquad (3)$$

where A might depend principally on stress and substructure. This formulation, which will be confirmed later, agrees with the data for Q over Region II where $H_D \simeq 67,000$ cal/mole, v $\simeq$ 0.138 (for σ in psi) and the sinh $\frac{\sigma v}{RT} \simeq 1/2 \, e^{\frac{\sigma v}{RT}}$. Over Region III, however, H_D is known to be

sensitive to the temperature; furthermore the term $\sinh \frac{\sigma v}{RT}$ has only a small effect on Q, particularly since lower values of σ were employed in this range.

Creep in Region IV appears to arise from some entirely different mechanism than that operative in Regions II and III. Whereas grain boundary migration was prevalent over Region IV, no grain boundary migration could be detected in Region II and only minor amounts of grain boundary migration were observed at the highest temperatures of Region III.

The question arises as to whether the stress enters Eq. (3) for Regions II and III exclusively as a result of its presence in the hyperbolic sine term or whether the stress also influences the creep rate as a result of its possible effect on A. To check this issue a number of tests were conducted over Regions II and III to determine the effect of abrupt changes in stress on the creep rate. Typical examples of the results of such tests are given in Figs. 4 and 5. Since H_D and v were previously shown to be independent of the strain, the decreasing value of $\dot{\epsilon}/\sinh \left(\frac{\sigma v}{RT}\right)$ over the primary stage of creep must be attributed to a decreasing value of A (Eq. 3). Therefore, the abrupt change in $\dot{\epsilon}/\sinh \left(\frac{\sigma v}{RT}\right)$ with a change in stress clearly reveals that increases in the stress not only decreases the apparent activation energy but that it also increases the values of A.

On both increases and decreases in stress, transients were obtained in the creep rate. In the example of Fig. 4 these transients existed over a range of creep strain of about 1%. Greater changes in stresses resulted in larger transients over greater ranges in strain; this is clearly seen

in Fig. 5. These transients were more pronounced at the lower test
temperatures and much less pronounced at the higher temperatures. It
was also observed that these transients were affected by the strain
rate in such a way that higher strain rates resulted in smaller transient
ranges. The transient region following a change in stress appeared to
be greater in the secondary than in the primary stage of creep.

The effect of stress on the secondary creep rate over Region III
is shown in Fig. 6. These data reveal that the same stress law applies
regardless of the wide variations in the apparent activation energies
over this region. The secondary creep rate is given by $\dot{\epsilon}_s = K\{T\} \, \sigma^n$ where
n = 6.9.

The microstructure of specimens before and after creep tests at
various temperatures is shown in Figs. 7a-f. Below 600°K (Fig. 7c)
heavy slip deformation of the grain took place. Grain boundaries were
slightly roughened due to the heterogenity of deformation. At 700°K
(not shown) the deformation bands began to polygonize into subgrain
networks. The subgrain networks were so extensive at 811°K (Fig. 7d)
that grain boundaries were severely serrated. At large strains extensive
void formation was observed at the grain boundaries. The subgrain size
was larger and more sharply defined at higher temperatures (Fig. 7e).
Grain boundaries were not serrated above 1050°K (Fig. 7f) and the grain
size seems to have increased slightly attended with larger subgrain sizes.

DISCUSSION

The creep behavior of α-Fe over both Regions II and III will be correlated in terms of a single dislocation mechanism, namely that for the thermally activated motion of jogged screw dislocations. This mechanism has already been discussed in some detail by several investigators.[6-10]

The formulation and interpretation employed by Raymond and Dorn[11] will be adopted here. Letting the shear strain rate to be approximated by $\dot{\gamma} = 3/4 \: \dot{\varepsilon}$ and the applied shear stress by $\tau = \sigma/2$, gives

$$\dot{\varepsilon} = \frac{8}{3} \rho \nu b^2 (z-1) \: e^{-\frac{G_D}{RT}} \sinh \frac{N\sigma\ell b^2}{2RT} \qquad (4a)$$

$$\dot{\varepsilon} = \frac{8}{3} \rho \nu b^2 (z-1) \: e^{\frac{S_D}{R}} \: e^{-\frac{H_D}{RT}} \sinh \frac{N\sigma\ell b^2}{2RT} \qquad (4b)$$

where ρ is the density of mobile screw dislocation, z the coordination number, ν the Debye frequency, b the Burgers' vector, ℓ the mean distance between jogs, S_D, H_D and G_D are the entropy, enthalpy and Gibbs' free energy per mole for self diffusion, R the gas constant, and N Avogadro's number. The self diffusivity can be given by

$$D = D_o' \: e^{-\frac{G_D}{RT}} = D_o \: e^{-\frac{H_D}{RT}} \qquad (5)$$

where

$$D_o = D_o' \: e^{\frac{S_D}{R}} = f\alpha\lambda^2\nu e^{\frac{S_D}{R}} = 0.72 \times 4 \frac{b^2}{3} \nu e^{\frac{S_D}{R}} \simeq 8.11 \times 10^{-3} \: e^{\frac{S_D}{R}} \quad cm^2/sec \quad (6)$$

f being the correlation factor, α the jump factor and λ the jump distance which are given for the BCC lattice following the third equality sign.

Eq. (4b) then reduces to

$$\dot{\varepsilon} \simeq \frac{8}{3} \rho(z-1) \, D \, \sinh \frac{N\sigma\ell b^2}{2RT} \tag{7}$$

Accordingly the apparent activation energy for creep by the motion of jogged screw dislocations is given by

$$Q = \frac{\partial \ln \dot{\varepsilon}}{\partial (-\frac{1}{RT})} = \frac{\partial \ln \rho}{\partial (-\frac{1}{RT})} + G_D - T \frac{\partial G_D}{\partial T} - \frac{N\sigma\ell b^2}{2} \coth \frac{N\sigma\ell b^2}{2RT} \tag{8a}$$

If H_D is, in fact, temperature dependent, from thermodynamic requirements S_D must also be temperature dependent. Using the thermodynamic identity $(\frac{\partial H_D}{\partial T}) = T (\frac{\partial S_D}{\partial T})$, Eq. (8a) can be written as

$$Q = \frac{\partial \ln \dot{\varepsilon}}{\partial (-\frac{1}{RT})} = \frac{\partial \ln \rho}{\partial (-\frac{1}{RT})} + H_D - \frac{N\sigma\ell b^2}{2} \coth \frac{N\sigma\ell b^2}{2RT} \tag{8b}$$

Although the physical interpretation of the various terms in Eqs. (8a) and (8b) is clear, it is nevertheless desireable to discuss, at this point, the order of magnitude of the first term to the right of the equality sign since it is frequently neglected as small. It arises from the fact that dislocations are released from some barriers as the temperature is increased since the barrier strength decreases as does the shear modulus of elasticity. Assume, empirically as a first approximation, that

$$\rho = A_o\{\varepsilon\} \left(\frac{\tau}{G}\right)^m \tag{9a}$$

during creep under a constant stress where G is the shear modulus of elasticity. The term $A_o\{\varepsilon\}$ decreases during the primary stage and achieves a steady-state value over the secondary stage of creep.

Accordingly the change in the density of dislocations with a change in temperature is independent of the stress and is given by

$$\frac{\partial \ln \rho}{\partial (-\frac{1}{RT})} = -\frac{mRT^2}{G}\frac{dG}{dT} \qquad (9b)$$

This contribution to the apparent activation energy for creep of α-Fe for m = 1 is shown in Fig. 8 as estimated from data presented by Lytton.[12] The increase in this quantity over Region III arises from the effect of the Curie transformation on the elastic moduli.

The correct value of m in Eq. (9b) can be obtained from the effect of stress on the secondary creep rate over Region III. Comparison of Eqs. (3) and (4) shows $\sinh\frac{N\tau\ell b^2}{RT} = \sinh\frac{2\tau v}{RT} \simeq \frac{2\tau v}{RT}$ for the smaller stresses used over this region. Using this relationship and Eqs. (7) and (9a) the secondary creep rate, $\dot{\varepsilon}_s$, can be approximated by

$$\dot{\varepsilon}_s = A_o\{\varepsilon\}\frac{8(z-1)N\ell b^2 D}{3RTG^m}\tau^{m+1} \qquad (10)$$

This was indeed observed in Region III as shown in Fig. 6 where m = 5.9.

Assuming m remains unchanged throughout the entire temperature range, the data in Fig. 2 are replotted in Fig. 9. The contribution from the moduli of elasticity to the apparent activation energy for creep is generally small except in the Curie transformation range. Similarly, the data in Fig. 3 are replotted in Fig. 10 which reveals that over the Region II

$$Q + \frac{mRT^2}{G}\frac{dG}{dT} = 64,000 - \sigma v \qquad \text{cal/mole} \qquad (11)$$

where σ is in psi and v has the value of 0.147. This value of v

and H_D = 64,000 cal/mole will be accepted in this analysis.

Over Region II the apparent activation energy for creep appears

to depend only on the stress. For the rather high stresses used in

this range $\coth \frac{N\sigma \ell b^2}{2RT} = \coth \frac{\sigma v}{RT} \simeq 1$. Thus Eq. (8b) reduces to Eq. (11)

when H_D is taken to be 64,000 cal/mole and ℓ = 19b and agrees well

with the data over Region II of Fig. 2.

In contrast, the apparent activation energy for creep over the

Curie range of Region III increases rapidly as the Curie temperature

is approached and seems to be rather insensitive to the stress.

Nevertheless, there appears to be no justification for assuming that

the creep mechanism would necessarily change as Fe goes through the

Curie transformation range. If, however, it is assumed that the non-

conservative motion of jogged screw dislocations is operative over

Region III as well as II, the large increase in Q with temperature

must be ascribed to the possibility that G_D depends on the temperature

in a way paralleling the magnetic transformation. In this event, as

shown by Eq. (8a), G_D must decrease as the temperature increases

over the Curie range.

The available data on the activation energy for self diffusion

in Fe as obtained by tracer techniques are given in Table II.

Unfortunately the results of different investigators exhibit appreciable

scatter. The value of H_D = 64,000 cal/mole, as deduced here from creep

data on α-Fe (ferromagnetic) over Region II, is, however, in nominal agreement with the average value reported from diffusion data. Since the scatter in H_D from diffusion data is so large due to differences in accuracy and techniques, it is not possible to deduce from the average of the data obtained by the various investigators whether or not the diffusivity changes throughout the Curie range. Although the absolute values of their H_D data might be questioned, the results of Buffington, Hirano, and Cohen[18] suggest that when using the same technique G_D decreases as the temperature increases over the Curie range. The value of H_D that they report for ferromagnetic Fe, however, may be too low since the lowest temperature that they employed (973°K) was well within the range of the Curie transformation.

In order to ascertain how G_D might vary throughout the Curie range (Region III) let

$$H_D = Q - \frac{\partial \ln \rho}{\partial(-\frac{1}{RT})} + \frac{N\sigma \ell b^2}{2} \coth \frac{N\sigma \ell b^2}{2RT} = G_D - T \frac{\partial G_D}{\partial T} \qquad (12)$$

as deduced from Eqs. (8a) and (8b). The trend of H_D with temperature is shown in Fig. 11. Over Region II

$$H_D = 64,000 \qquad \text{cal/mole} \qquad 480° \leq T \leq 775°K \qquad (13a)$$

Above 775°K, however, H_D increases and, as shown by integration of Eq. (12)

$$G_D\{T\} = 45.68\ T + T \int_T^{1043} \frac{H_D}{T^2}\ dT \qquad \text{cal/mole} \qquad 775°K \leq T \leq T_c = 1043 \qquad (13b)$$

The values of D thus deduced from Eqs. (5) and (13b) are compared with those obtained by other investigators using tracer techniques in Fig. 12. The agreement is nominally good and confirms the concept that the creep of Fe over Regions II and III (480° to 1050°K) is diffusion controlled and obeys the relationship demanded by the non-conservative motion of jogged screw dislocations. The anomalous increase in the apparent activation energy for creep over Region III must be attributed to a decrease in the free activation energy for diffusion as the temperature increases over the range of the Curie transformation.

If the same mechanism of creep applies over Regions II and III, it necessarily follows that the form of the relationship between creep rate and stress should also be the same over both regions. The results given in Figs. 4 and 5 clearly demonstrate that the creep rate is increased more rapidly upon increasing the stress than suggested solely by its influence on the hyperbolic sine term. This demands that ρ also increases as the stress increases. Using Eq. (7), the previously determined values of D, (Fig. 12) and $\ell = 19b$, typical creep data for Region III are replotted in Fig. 13 in terms of ρ versus σ for several different temperatures. These data reveal that the density of dislocations increases with increasing stress during secondary creep over Region III. The effect of stress on the density of the dislocations is practically independent of the test temperature provided the very small apparent systematic deviation is neglected. It appears, however, that the so-calculated density of

mobile dislocations might be slightly too high especially at the higher stresses. Using the same proceedure, the density of dislocations were calculated for the primary stage of creep over Region II, as shown in Fig. 14. These densities are unacceptably high suggesting, because of the excellent agreement between experiment and all other features of the theory, that some further modification of the theory is required.

It is indeed surprising that volume diffusion controlled creep in α-Fe might extend to as low as 475°K. Undoubtedly diffusion at such low temperatures must, in large measure, be modified by various short-circuiting processes such as, for example, pipe diffusion along moving dislocations. The effect of strain-rate on diffusivity has been observed by a number of investigators,[19-22] and recently Balluffi and Ruoff[23-25] presented a rather detailed analysis of this phenomenon.

Hirano, Cohen, Averbach and Ujiiye[21] have shown that the diffusivity, D_d, in α-Fe under compressive straining increases with the strain rate, $\dot{\varepsilon}$, according to the relationship

$$\frac{D_d}{D} = 1 + Be^{\frac{H_p}{RT}} \dot{\varepsilon} \tag{14}$$

where D is the diffusivity when $\dot{\varepsilon} = 0$ and $B = 3 \times 10^{-3}$ sec and $H_p = 39,200$ cal/mole. Assuming, then that the diffusivity is strain rate sensitive, D of Eq. (7) should be replaced by D_d as given in Eq. (14). Accordingly, Eqs. (7) and (8a) can be rewritten as

$$\dot{\varepsilon} \simeq \frac{8}{3} \rho(z-1) D_d \sinh \frac{N\sigma\ell b^2}{2RT} \tag{15}$$

and

$$Q = \frac{\partial \ell n \dot{\varepsilon}}{\partial(-\frac{1}{RT})} = \frac{\partial \ell n \rho}{\partial(-\frac{1}{RT})} + G_D - T \frac{\partial G_D}{\partial T} - \frac{N\sigma\ell b^2}{2} \coth \frac{N\sigma\ell b^2}{2RT}$$

$$+ \frac{\rho[\frac{\partial \ell n \rho}{\partial(-\frac{1}{RT})} + G_D - T \frac{\partial G_D}{\partial T} - H_p - \frac{N\sigma\ell b^2}{2} \coth \frac{N\sigma\ell b^2}{2RT}]}{\frac{3e^{\frac{G_D - H_p}{RT}}}{8(z-1) \, B \, D_o' \, \sinh \frac{N\sigma\ell b^2}{2RT}} - \rho} \tag{16}$$

For a test run at constant temperature in Region III, Eq. (15) gives the secondary creep rate

$$\dot{\varepsilon}_s \simeq \frac{8}{3} \rho(z-1) \, D_o' \, e^{-\frac{G_D}{RT}} (1 + B e^{\frac{H_p}{RT}} \dot{\varepsilon}_s) \frac{N\sigma\ell b^2}{2RT} \tag{17}$$

or

$$\ell n \frac{\dot{\varepsilon}_{s2}}{\dot{\varepsilon}_{s1}} = \ell n \frac{\rho_2(1 + Be^{\frac{H_p}{RT}} \dot{\varepsilon}_{s2})}{\rho_1(1 + Be^{\frac{H_p}{RT}} \dot{\varepsilon}_{s1})} + n \frac{\sigma_2}{\sigma_1} \tag{18}$$

Thus Eq. (18) agrees well with the results shown in Fig. 6 when

$$\ell n \frac{\rho_2(1 + Be^{\frac{H_p}{RT}} \dot{\varepsilon}_{s2})}{\rho_1(1 + Be^{\frac{H_p}{RT}} \dot{\varepsilon}_{s1})} = 5.9 \, \ell n \frac{\sigma_2}{\sigma_1} \tag{19}$$

Furthermore, if Eq. (19) is compared with Eq. (7) it clearly implies that

$$\ell n \left(\frac{\rho_2}{\rho_1}\right) \text{(with no strain-rate enhanced D)} = \ell n \left(\frac{\rho_2}{\rho_1}\right) \text{(with enhanced } D_d)$$

$$+ \ln \frac{(1 + Be^{\frac{H_p}{RT}} \dot{\varepsilon}_{s2})}{(1 + Be^{\frac{H_p}{RT}} \dot{\varepsilon}_{s1})} \qquad (20)$$

Therefore, using Eq. (15), the data in Fig. 13 are replotted in Fig. 15 by taking H_p = 39,200 cal/mole and B = 3 x 10^{-5} sec. The B value of Hirano et al was found to be about two orders of magnitude too high and gave strange results. Figure 15 gives about the right order of magnitude for ρ and shows about the same stress dependency as in Fig. 13.

Similarly, the data in Fig. 14 are recalculated by using Eq. (15) and the density of mobile dislocations, ρ, corresponding to the secondary stage is extrapolated and summarized in Table III. These results show that a higher value of ρ is obtained at the higher stress level.

Although the absolute values of ρ in Table III might still be slightly too high, they are not unacceptable; particularly when the detailed mechanism of the strain-rate enhancement of diffusivity is not well understood in this region. Certainly the results in Table III seem to indicate that the unacceptably high ρ in Fig. 14 arises from neglecting the effect of plastic deformation on the diffusivity in the initial analysis.

Now the question arises whether the introduction of D_d, instead of D, into Eq. (7) will change the experimental results of apparent activation energies for creep. It can be shown easily from Eq. (16) that the last term beyond the second equality sign is always negligible.

For example, for T = 606°K, σ = 13,390 psi and ρ = 4 x 10^9 cm^{-2} in

Region II, Eq. (16) gives

$$Q = 1560 + 64,000 - 2120 + \frac{\rho(1560 + 64,000 - 39,200 - 2120)}{1.55 \times 10^{11} - \rho}$$

$$= 63,440 + \frac{4 \times 10^9 \times 24,240}{1.55 \times 10^{11} - 4 \times 10^9}$$

$$\simeq 63,440 \qquad \text{cal/mole.}$$

The value of ρ is estimated from the results in Table III based on the applied stress.

Again for $T = 1006°K$, $\sigma = 1400$ psi and $\rho = 1.5 \times 10^6$ cm^{-2} (taken from Fig. 15) in Region III.

$$Q = 14,960 + 97,100 - 2000 + \frac{\rho(14,960 + 97,100 - 39,200 - 2000)}{3.9 \times 10^8 - \rho}$$

$$= 110,060 + \frac{1.5 \times 10^6 \times 70,860}{3.9 \times 10^8 - 1.5 \times 10^6}$$

$$\simeq 110,060 \qquad \text{cal/mole.}$$

Therefore, the values of Q, the apparent activation energy for creep, remain unchanged with the introduction of D_d into Eq. (7). Thus all of the data over Regions II and III are in good agreement with the assumed model for creep.

The transients that were noted in the creep rate upon rapid changes in stress, as documented in Figs. 4 and 5, are primarily due to abrupt changes in the densities of the mobile dislocations. When the stress is decreased from a high value, dislocations are temporarily blocked by barriers which then recover causing a slight increase in creep rate before the usual decrease again sets in. Upon an increase in stress, a larger number of mobile dislocations are produced causing a substantially

higher instantaneous creep rate which soon levels off to the usual trend as more barriers are produced.

Over Region IV a major contribution to creep occurs by some alternate mechanism than that ascribed to Regions II and III. Although no recrystallization was noted during creep over Regions II, III, and IV (vide Fig. 7) the grain boundaries were more regular and the subgrain sizes were larger after creep at temperatures above T_c. Subgrain and grain boundary migration were appreciable over Region IV and may have started at the higher temperatures of Region III. As previously discussed by Raymond and Dorn,[11] when grain boundary and subgrain boundary migration occurs, the regions swept out are rather free of dislocations and higher average creep rates are obtained. Furthermore the activation energy for this process has been shown to be greater than that for self diffusion. It is therefore proposed that this may be the reason for the high activation energies noted in Region IV.

CONCLUSIONS

(1) The apparent activation energy for creep from 480° to 775°K is insensitive to temperature and strain and decreases linearly with increasing stress; over the Curie transformation range, 775° to 1045°K, the apparent activation energy increases steeply with increasing temperature and decreases precipitously at the Curie temperature, and increases again at higher temperatures.

(2) The anomalous increase in activation energy with increased temperature over the range of the Curie transformation is in complete harmony with the decrease in the free activation energy for self diffusion.

(3) The activation energy for self diffusion in ferromagnetic iron was found to be about 64,000 cal/mole.

(4) Over the entire range of temperatures from 480° to 1045°K the creep behavior was found to be in agreement with the mechanism of non-conservative motion of jogged screw dislocations.

(5) The effect of stress on the creep rate from 480° to 1045°K was found to arise not only from the effect of stress in the activation energy but also because the density of mobile dislocations increased as the stress was increased.

(6) The evidence suggests that the creep strain rate is enhanced by pipe diffusion along moving screw dislocations particularly over the temperature range from 480° to 775°K.

ACKNOWLEDGMENT

REFERENCES

1. O. D. Sherby, R. L. Orr and J. E. Dorn, J. Metals, 71 (Jan. 1954).

2. H. J. Tapsell and W. J. Clenshaw, Properties of Materials at High
   Temperatures--I. Great Britain Dept. of Science and Industrial
   Research. Special Report No. 1 (1927).

3. O. D. Sherby and J. L. Lytton, J. Metals, 928 (Aug. 1956).

4. C. H. M. Jenkins and G. A. Mellor, J. Iron and Steel Inst., 132,
   179 (1935).

5. P. R. Landon, J. L. Lytton, L. A. Shepard and J. E. Dorn, Trans. ASM,
   51, 900 (1959).

6. N. F. Mott, Phil. Mag., 43, 1151 (1952).

7. A. Seeger, Phil. Mag., 46, 1194 (1955).

8. J. Friedel, Phil. Mag., 46, 1169 (1956).

9. P. B. Hirsch and D. H. Warrington, Phil. Mag., 6, 735 (1961).

10. J. E. Dorn and J. D. Mote, "Phys. Aspects of Creep", High Temp. Str.
    and Mat'ls., 95, Pergamon Press, New York (1963).

11. L. Raymond and J. E. Dorn, Trans. AIME, 230, 560 (1964).

12. J. L. Lytton, J. Appl. Phys., 35, 239 (1964).

13. C. E. Birchenall and R. F. Mehl, Trans. AIME, 188, 144 (1950).

14. F. S. Buffington, I. D. Bakelar and M. Cohen, The Phys. of Powder
    Metallurgy, (edited by W. E. Kingston) Ch. 6, McGraw-Hill, New York (1951).

15. V. M. Golikov and V. T. Borisov, 4th Symp. on the Problem of Metallography
    and the Physics of Metals, (edited by B. Ya. Lubov) U. S. Atomic
    Energy Commission 2924, 365 (1955).

16. C. Leymonie and P. Lacombe, Rev. Metall., 6, 524 (1958).

17. R. J. Borg and C. E. Birchenall, Trans. AIME, 218, 980 (1960).

18. F. S. Buffington, K. Hirano and M. Cohen, Acta. Met., 9, 434 (1961).

19. C. H. Lee and R. Maddin, J. Appl. Phys., 32, 1846 (1961).

20. A. F. Forestien and L. A. Girifalco, J. Phys. Chem. Solids, 10, 99 (1959).

21. K. Hirano, M. Cohen, B. L. Averbach and N. Ujiiye, Trans. AIME, 227, 950 (1963).

22. A. R. Wazzan and J. E. Dorn, J. Appl. Phys., 36, 222 (1965).

23. R. W. Balluffi and A. L. Ruoff, J. Appl. Phys., 34, 1634 (1963).

24. A. L. Ruoff and R. W. Balluffi, J. Appl. Phys., 34, 1848 (1963).

25. A. L. Ruoff and R. W. Balluffi, J. Appl. Phys., 34, 2862 (1963).

Table I

Apparent Activation Energies

$\bar{T} \simeq \dfrac{2T_1 T_2}{T_1 + T_2}$ °K	σ psi	Q Kcal/mole
1147.8	860	98.3
1127.7	1,170	93.7
1112.5	1,380	91.8
1094.0	1,400	90.6
1056.7	2,700	101.8
1054.7	1,796	101.4
1054.3	956	95.3
1033.2	2,280	129.6
1008.1	2,670	105.1
1006.2	1,400	110.1
950.6	4,420	84.6
920.7	4,912	73.3
914.2	2,670	73.9
811.2	6,970	69.2
806.8	5,770	67.2
698.8	9,230	65.6
698.9	14,500	65.2
608.7	21,970	62.8
605.6	13,390	63.7
605.5	17,700	64.8
524.5	20,330	61.8
481.6	26,980	63.2
481.5	24,830	62.1
481.4	28,270	57.5
425.5	25,930	55.4
421.5	28,060	54.6
420.2	31,450	40.3
377.4	30,490	40.8

Table II

Comparison of Self-diffusion Results for BCC Iron

Investigator	D_o (cm^2/sec)	H_D $(kcal/mole)$	Phase	Temperature Range $(^\circ K)$
Birchenall and Mehl[13] (1950)	2300	73.2	α,β (not differentiated)	993-1173
Buffington, Bakalar and Cohen[14] (1951)	5.8	59.7	β (not differentiated)	1082-1178
Golikov and Borison[15] (1955)	530	67.1	α,β	923-1123
Leymonie and Lacombe[16] (1958)	18	64.1	β	1048-1165
Borg and Birchenall[17] (1960)	118	67.2	β	1069-1168
Buffington, Hirano and Cohen[18] (1960)	1.9	57.2	β	1082-1178
Buffington, Hirano and Cohen[18] (1960)	2.0	60.0	α	973-1023

β-Fe was paramagnetic; but for the temperatures that were employed, the α-Fe was always in the Curie range and therefore only partially ferromagnetic.

Table III

Density of Mobile Dislocations in Region II
Corrected for Strain-rate Enhancement of Diffusivity

T, °K	σ, psi	ρ, cm^{-2}
601	24,000	4.7×10^{10}
773	12,250	2.9×10^{9}
773	11,750	1.4×10^{8}
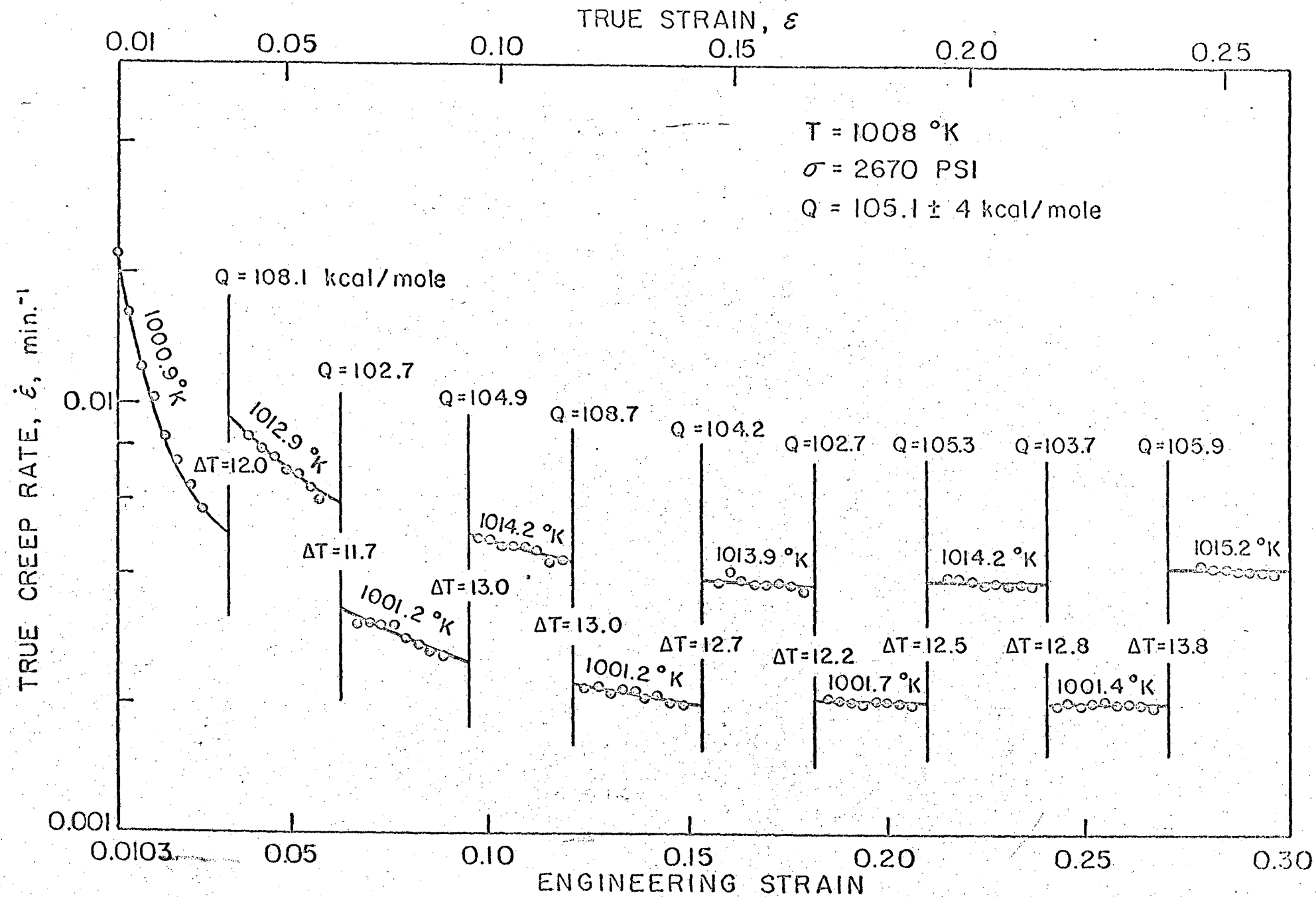
FIG. 1 TYPICAL EVALUATION OF THE APPARENT ACTIVATION ENERGIES.
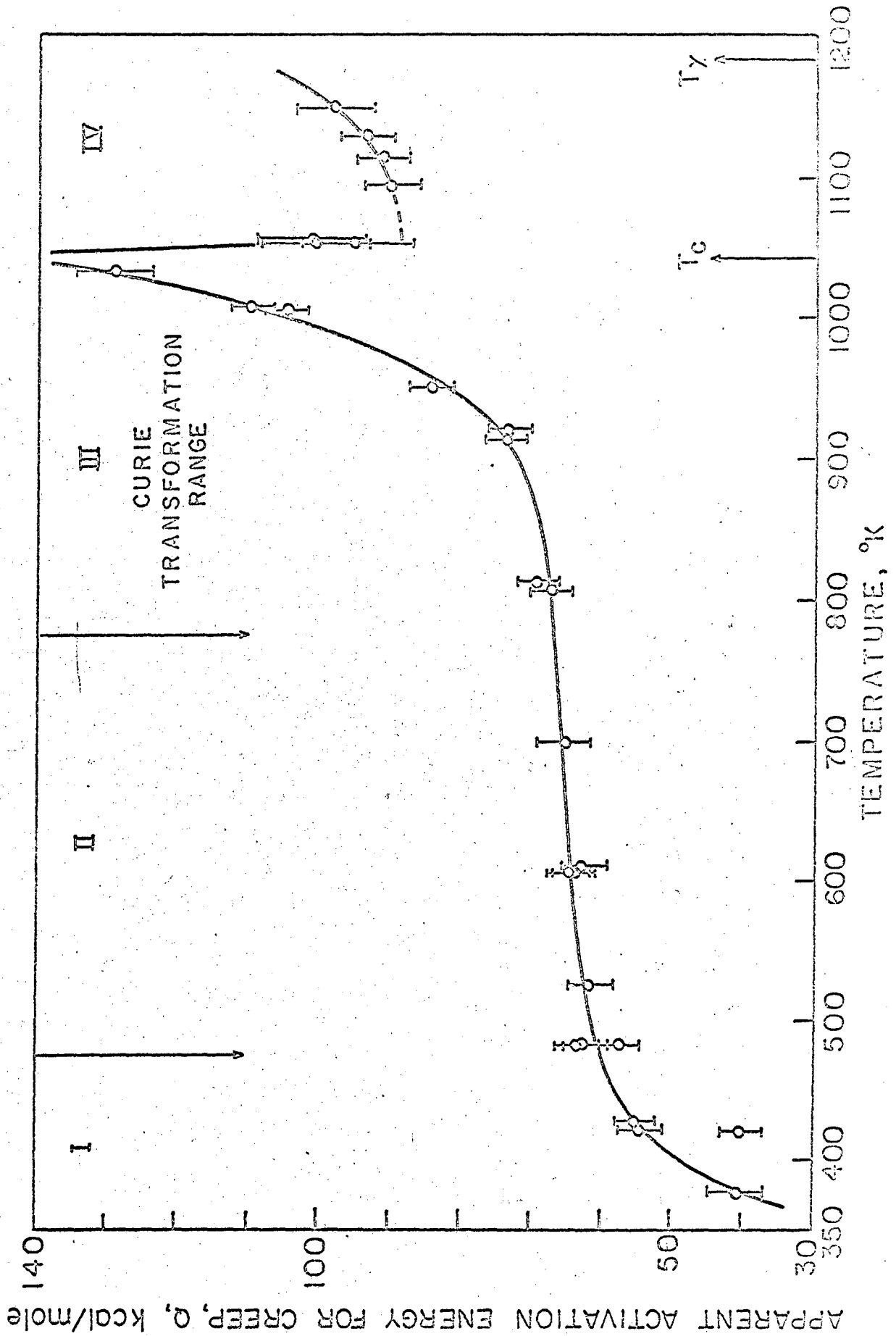
FIG.2 EFFECT OF TEMPERATURE ON THE APPARENT ACTIVATION ENERGY
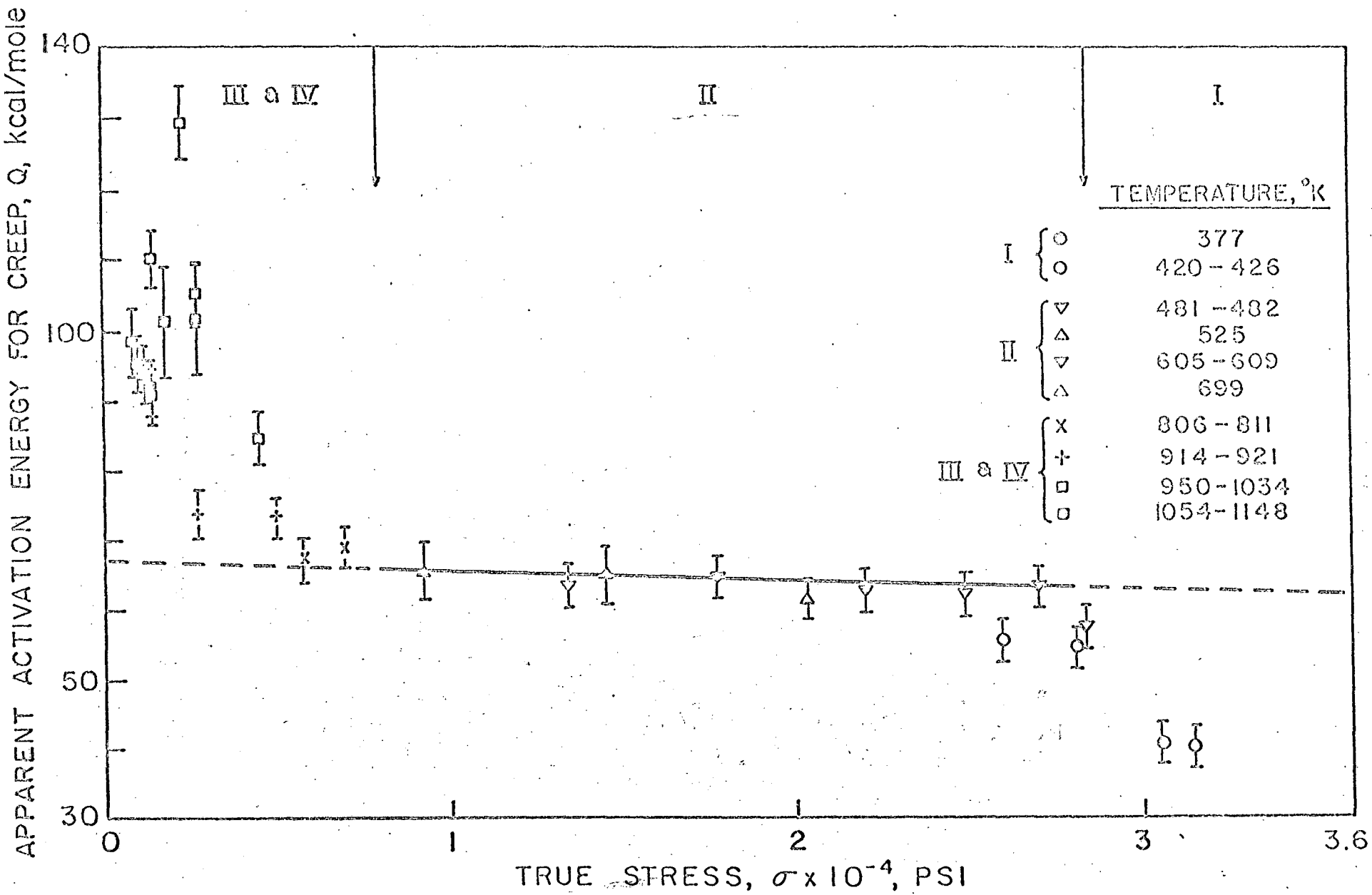
FIG. 3 EFFECT OF STRESS ON THE APPARENT ACTIVATION ENERGY FOR CREEP.
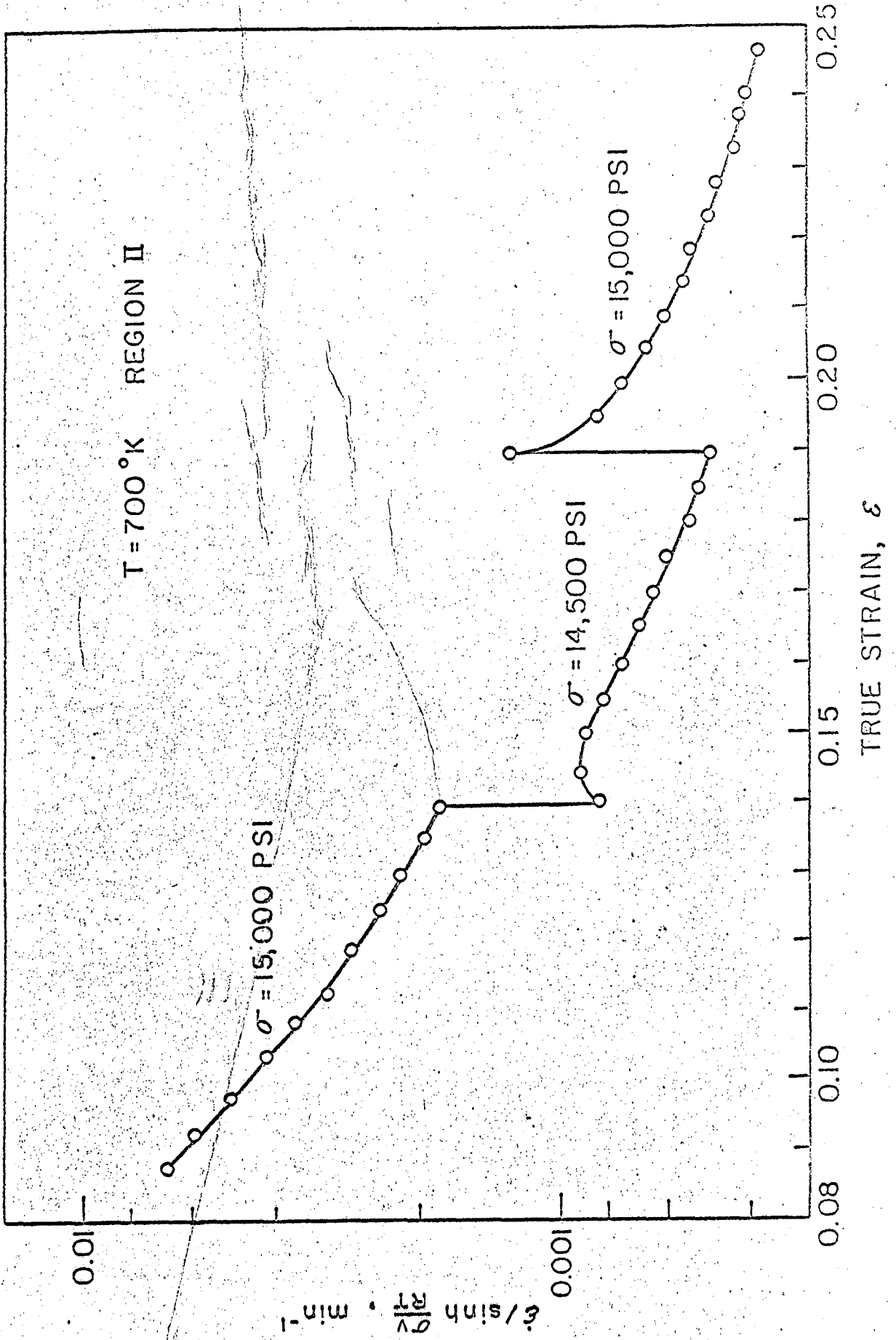
FIG. 4 EFFECT OF CHANGES IN STRESS ON THE CREEP RATES.
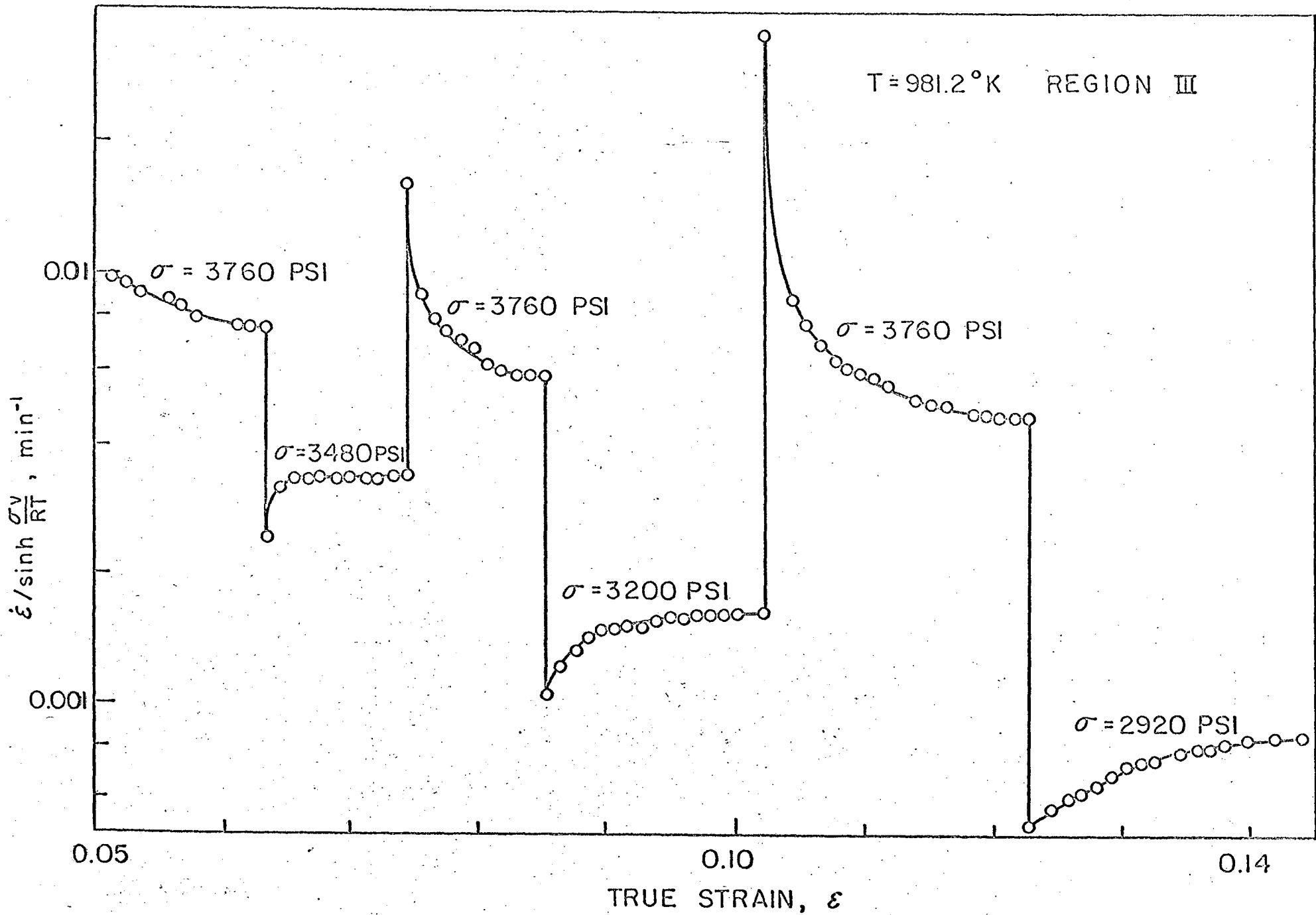
FIG. 5 EFFECT OF CHANGES IN STRESS ON THE CREEP RATES.
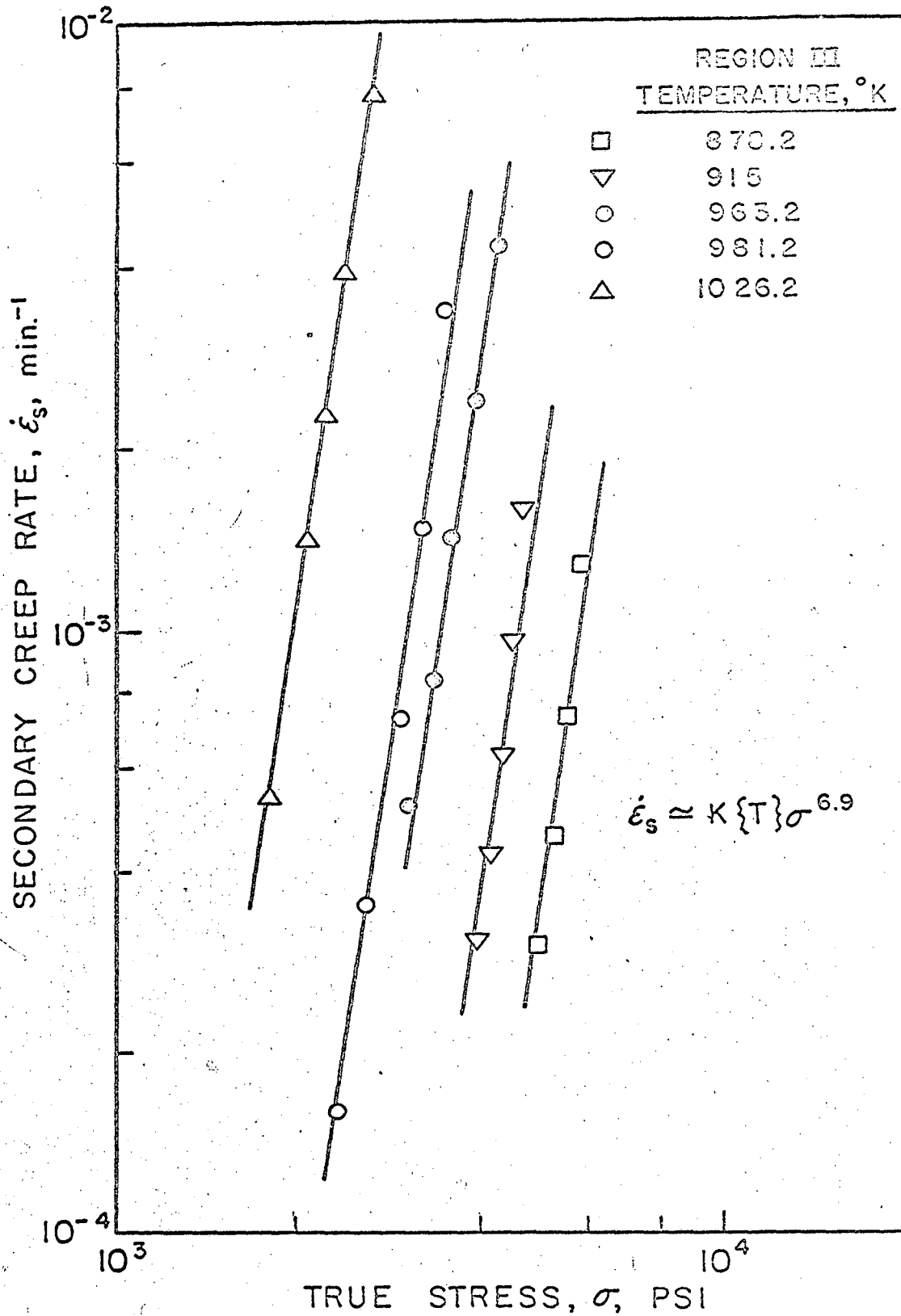
FIG. 6 EFFECT OF STRESS ON THE
SECONDARY. CREEP RATE.
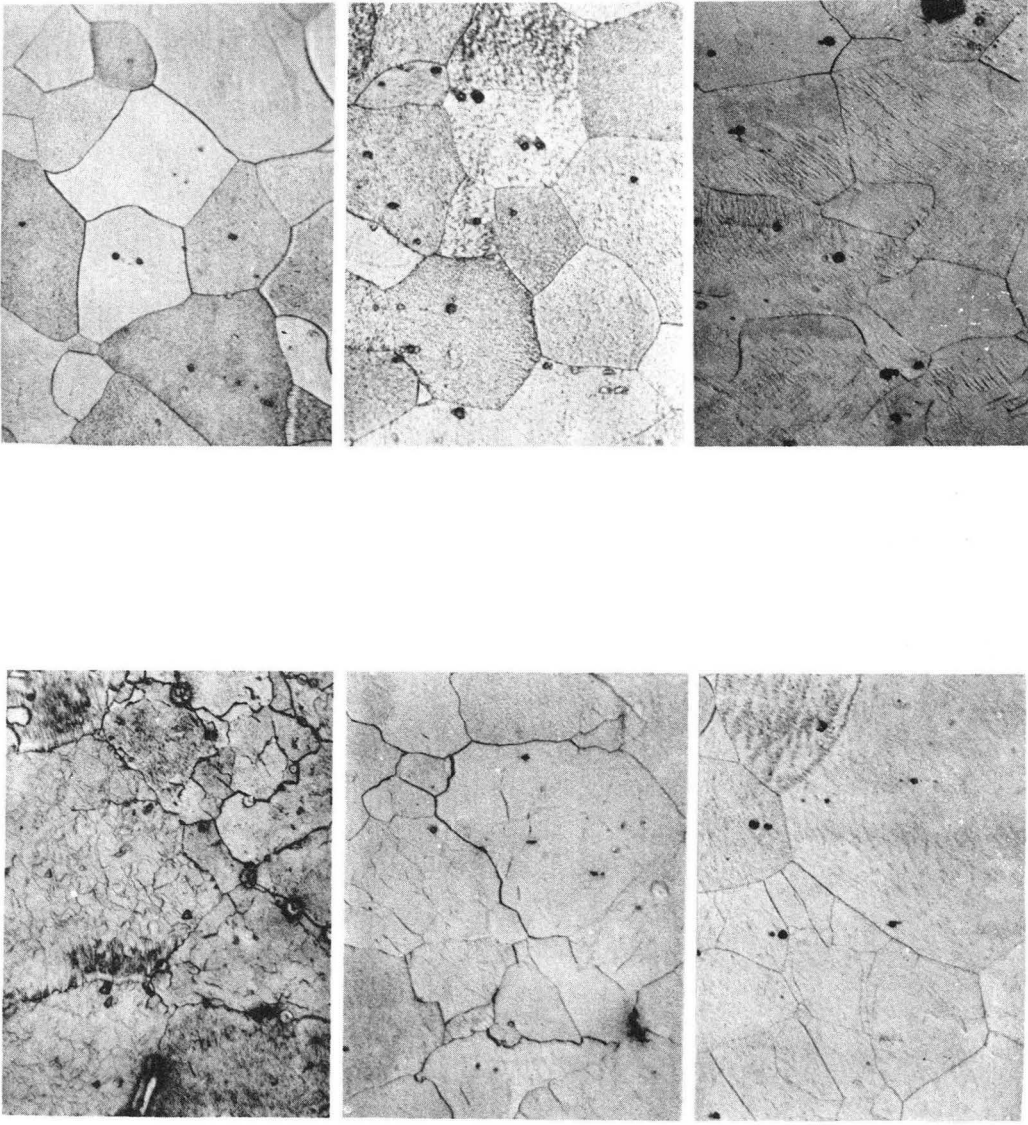
ZN-5267

Fig. 7
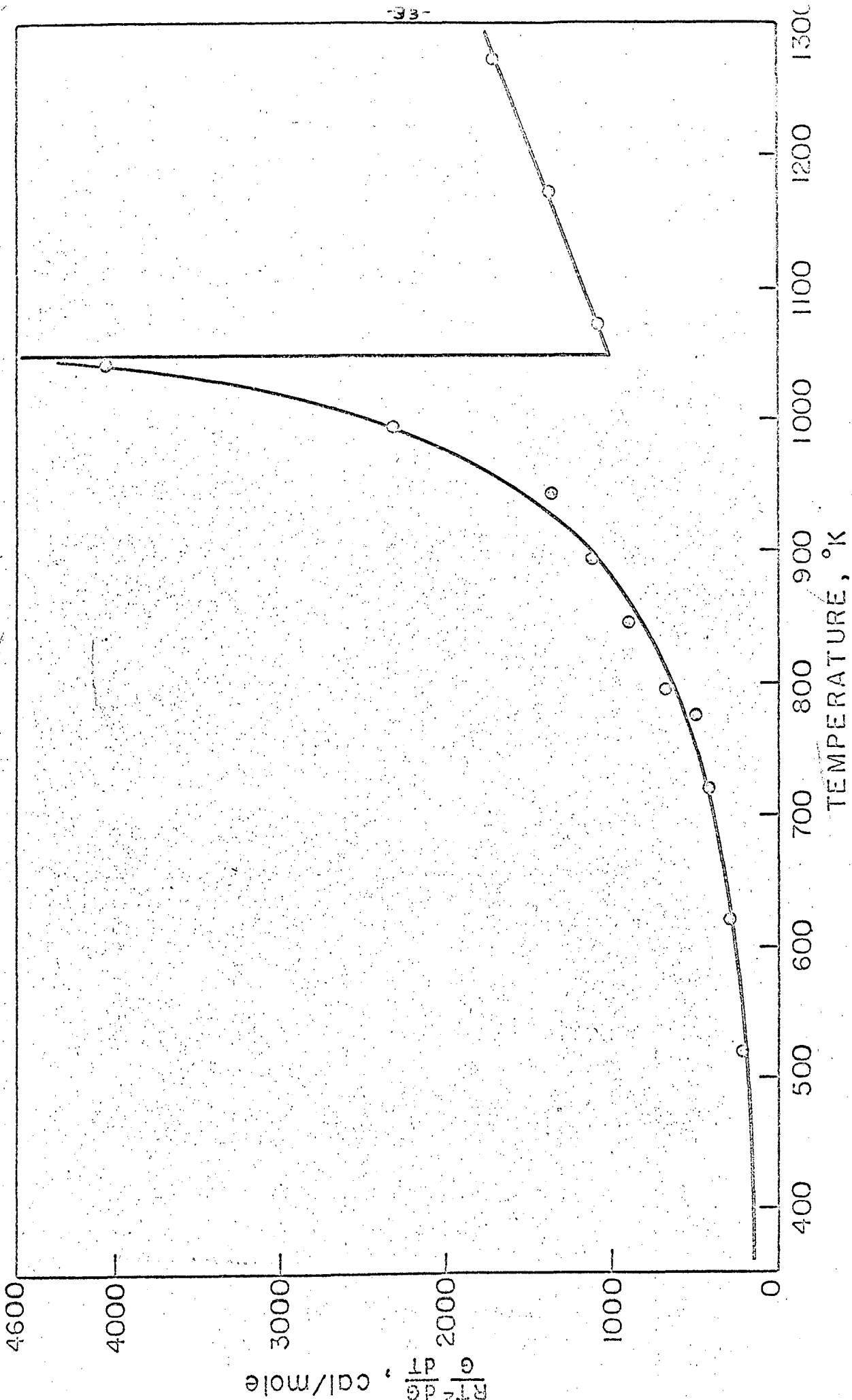
FIG. 8 CONTRIBUTION TO THE APPARENT ACTIVATION ENERGY FOR CREEP DUE
TO THE CHANGE OF SHEAR MODULUS WITH TEMPERATURE. (DATA FROM REF. 12).
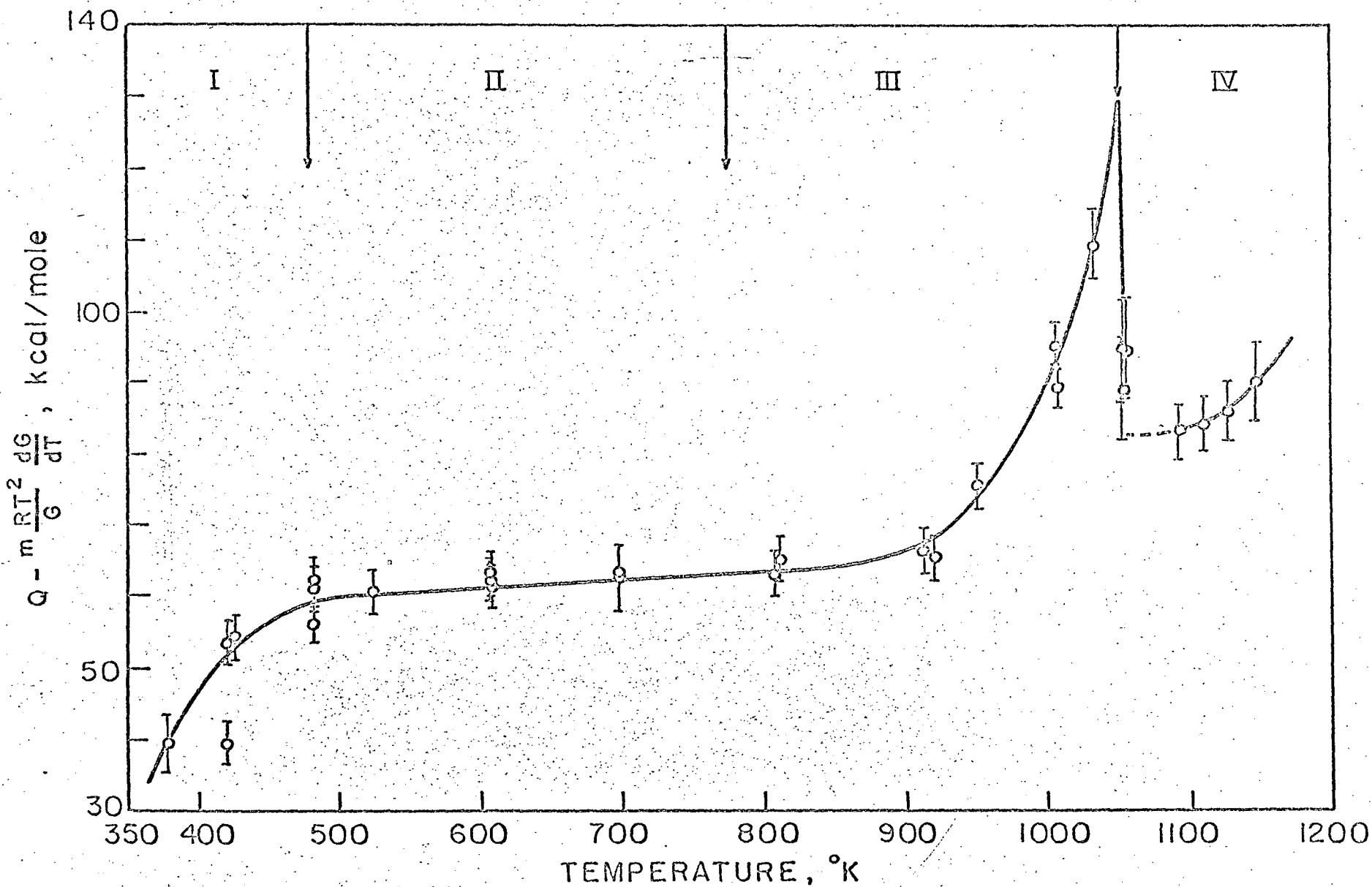
FIG. 9 APPARENT ACTIVATION ENERGIES CORRECTED FOR
CHANGES OF SHEAR MODULUS.

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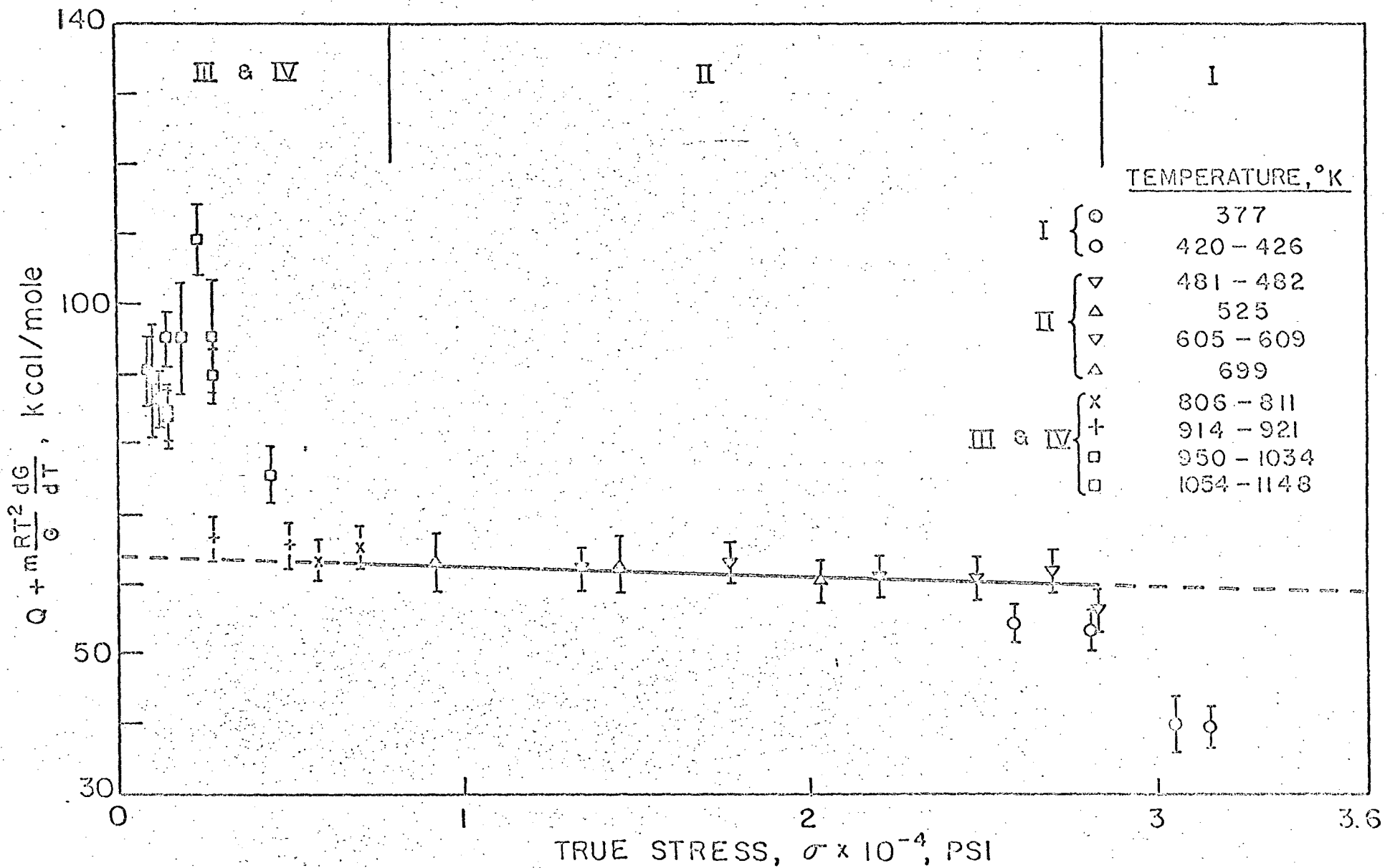
FIG. 10 EFFECT OF STRESS ON CREEP ACTIVATION ENERGY.
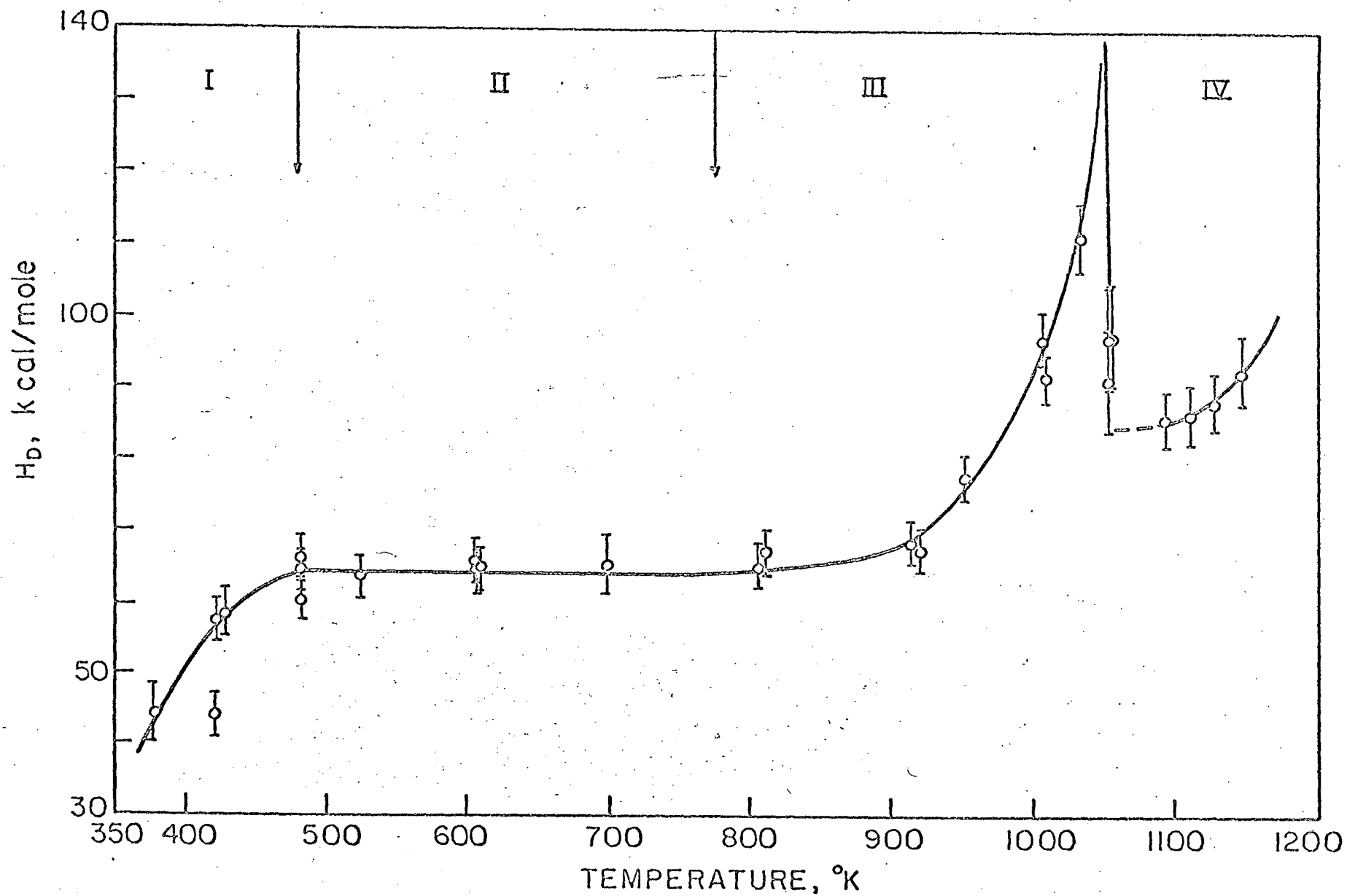
FIG. II. VARIATION OF H_D WITH TEMPERATURE.

-36-
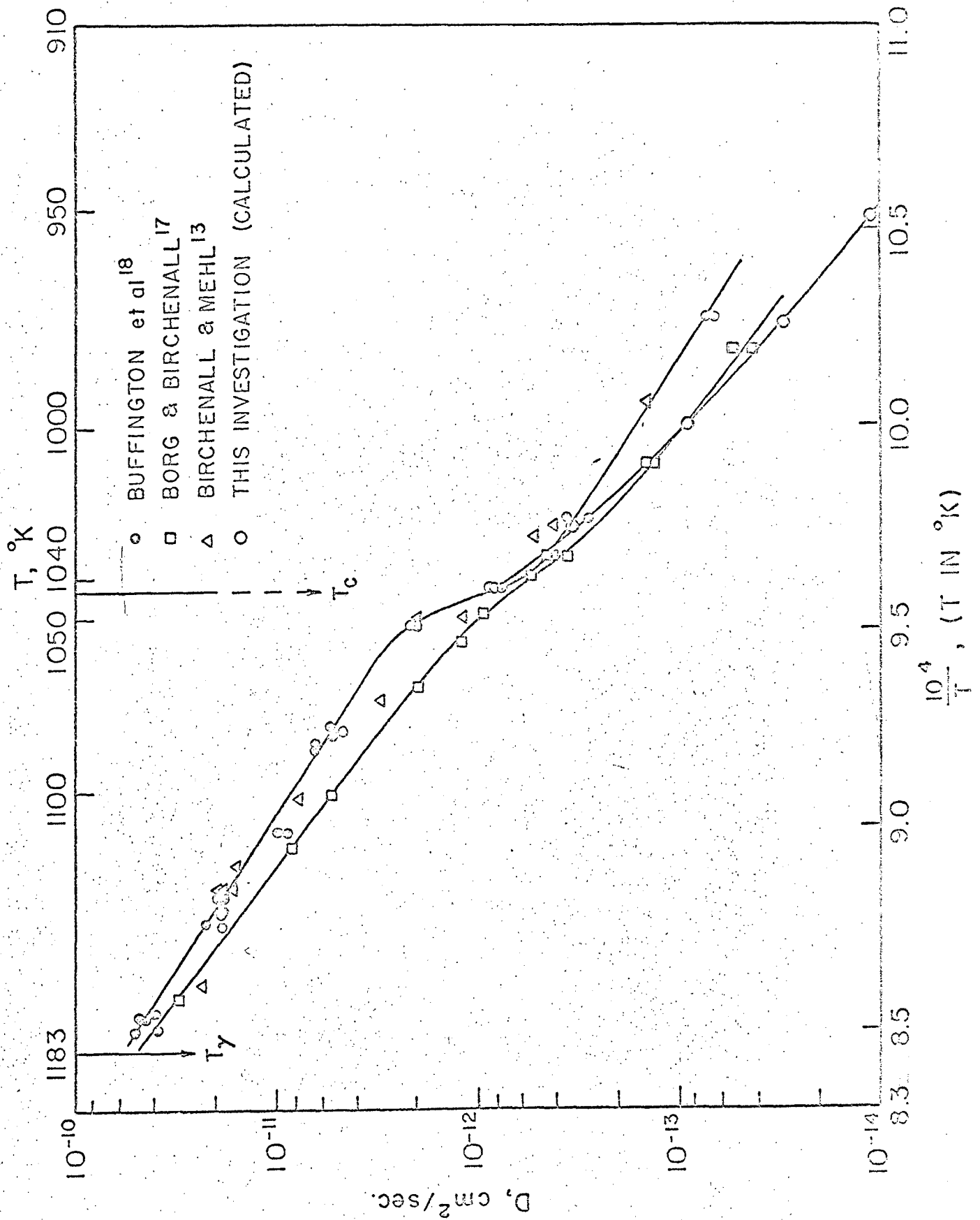
FIG. 12. THE DIFFUSION COEFFICIENT OF α AND β IRONS AS A FUNCTION OF TEMPERATURE.
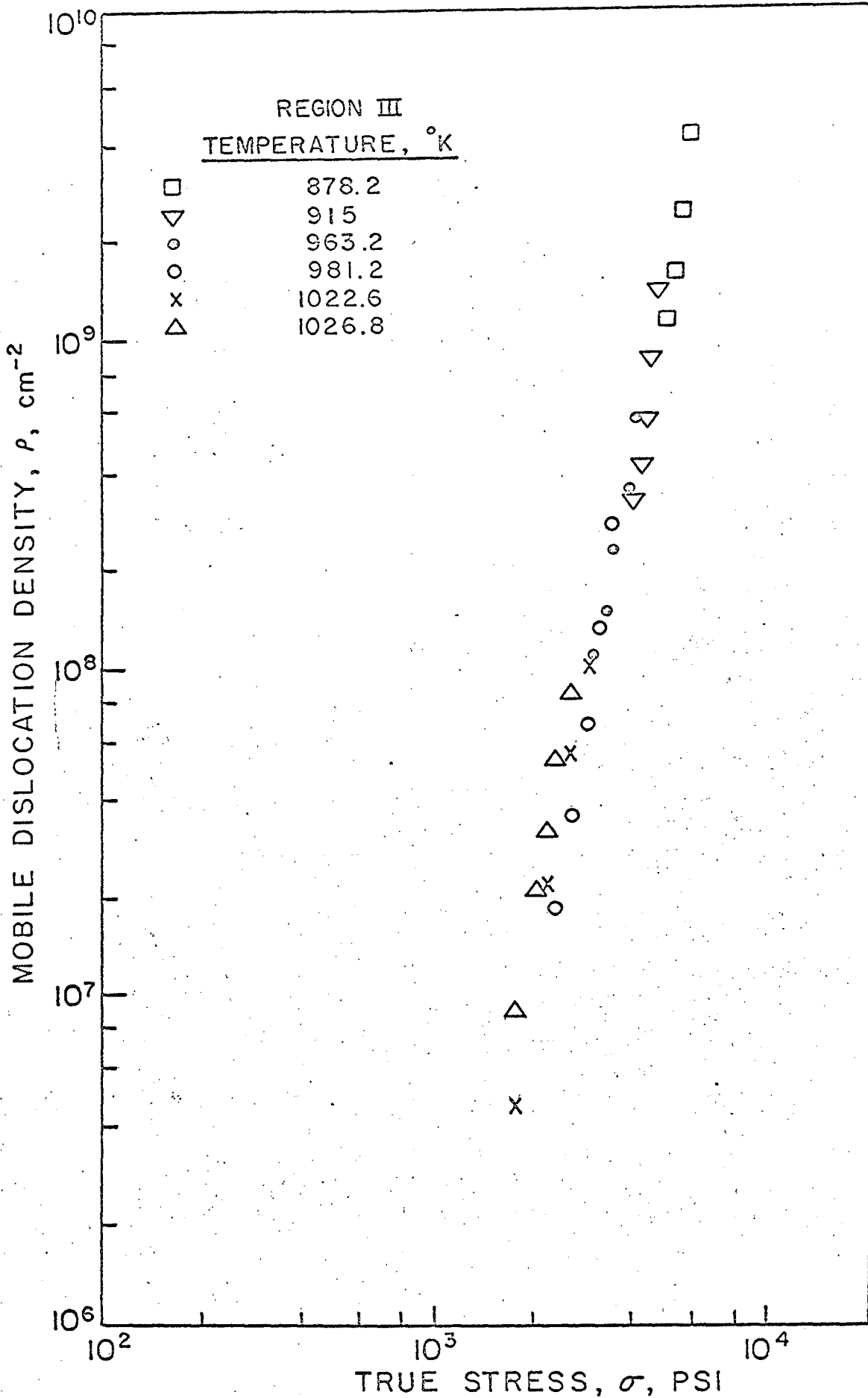
FIG. 13 EFFECT OF STRESS ON THE DENSITY OF
MOBILE DISLOCATIONS DURING SECONDARY CREEP.

	T,°K	σ, PSI	REGION
▽	601	24,000	II
□	773	12,250	II
△	773	11,750	II
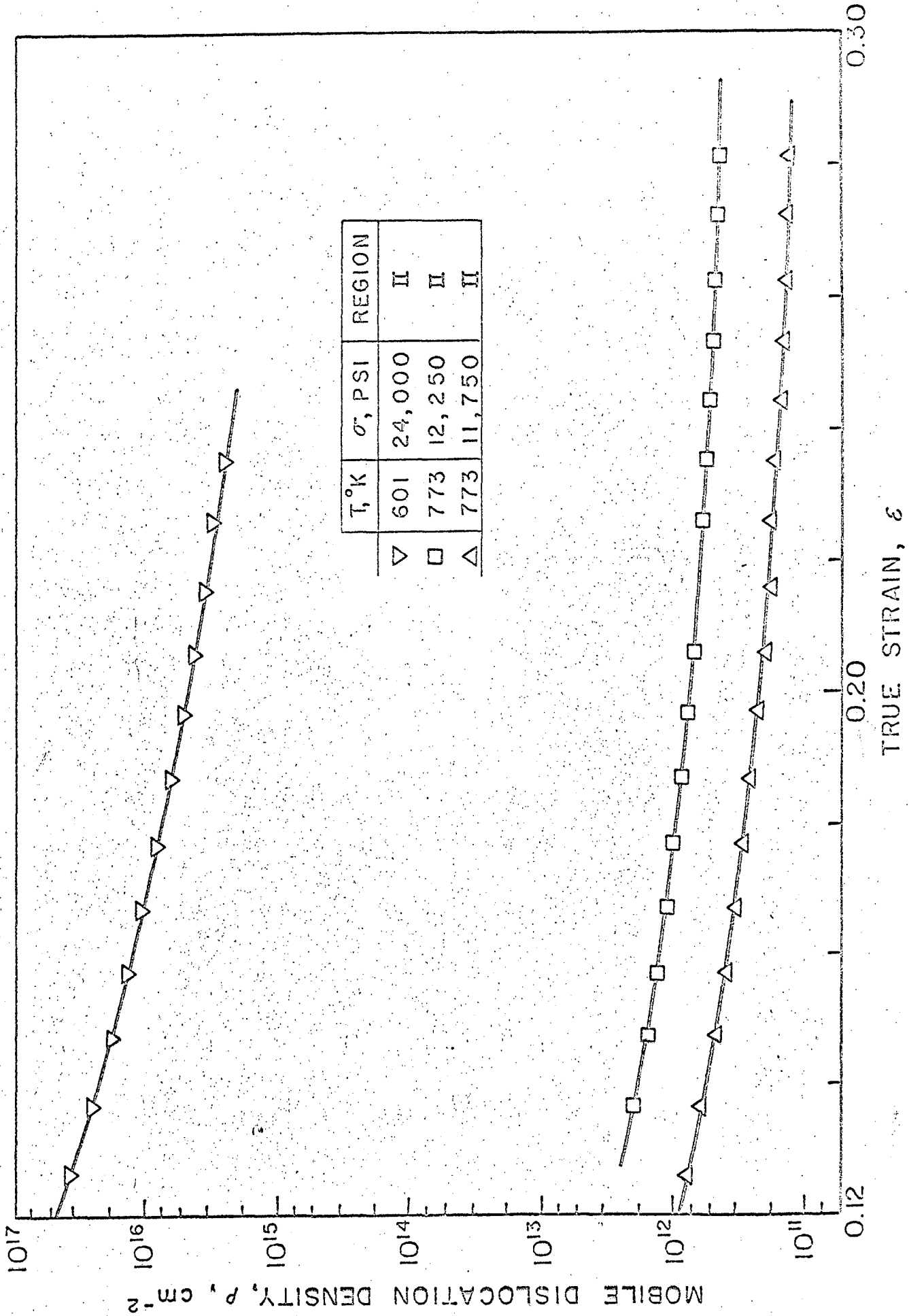
FIG. 14 EFFECT OF STRESS AND STRAIN ON THE DENSITY OF DISLOCATIONS.
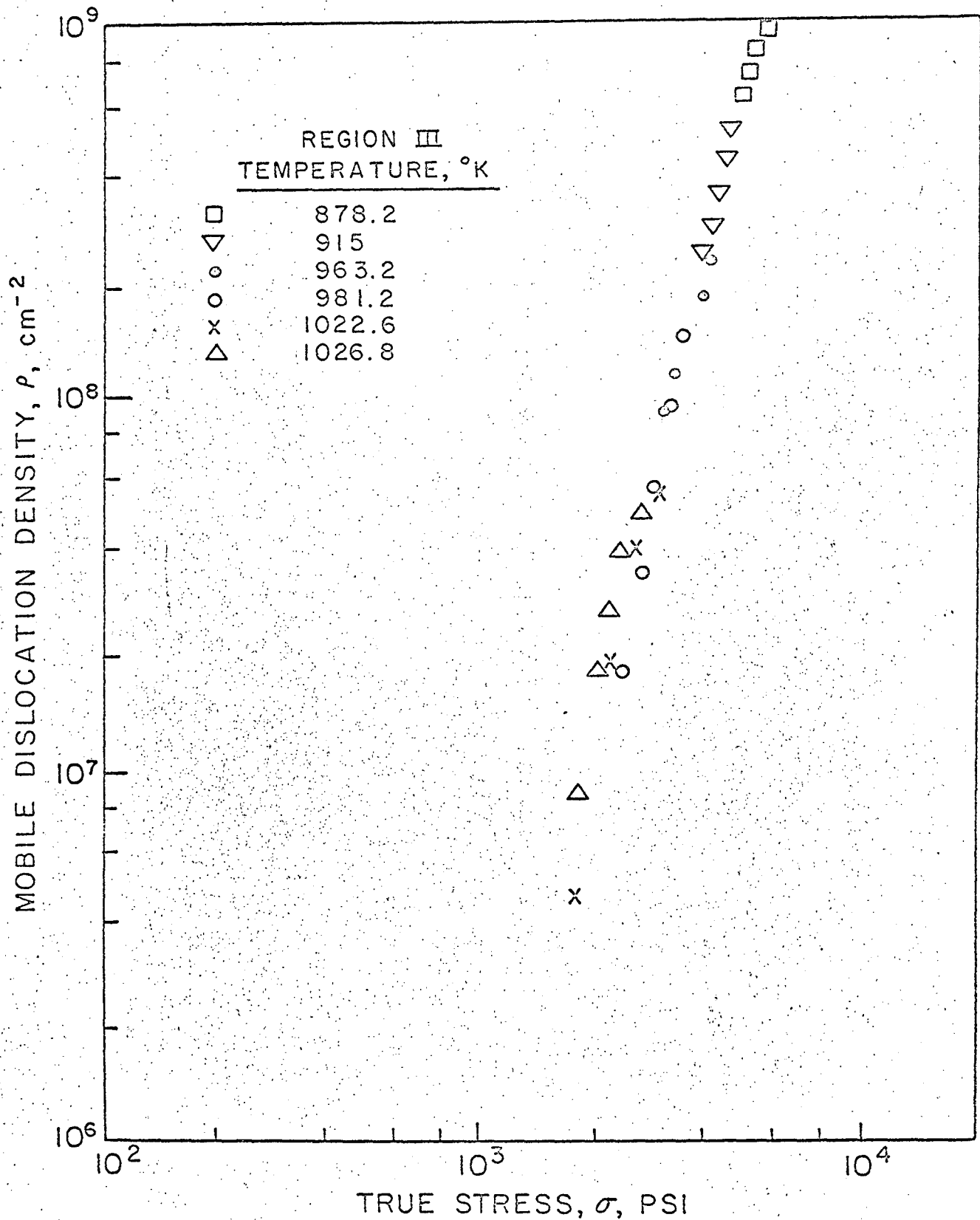
FIG. 15 EFFECT OF STRESS ON THE DENSITY OF MOBILE.
DISLOCATIONS DURING SECONDARY CREEP CORRECTED FOR
THE STRAIN-RATE ENHANCED DIFFUSIVITY.

FIGURE CAPTIONS

Figure 1. Typical Evaluation of the Apparent Activation Energies.

Figure 2. Effect of Temperature on the Apparent Activation Energy.

Figure 3. Effect of Stress on the Apparent Activation Energy for Creep.

Figure 4. Effect of Changes in Stress on the Creep Rates.

Figure 5. Effect of Changes in Stress on the Creep Rates.

Figure 6. Effect of Stress on the Secondary Creep Rate.

Figure 7. Photomicrographs (x250) of α-Fe Before and After Creep Test at Various Temperatures.

        a. Before test, as annealed at 1113°K for 30 minutes.

        b. Before test, as annealed at 1173°K for 30 minutes.

        c. Crept to 0.30 strain at 482°K.

        d. Crept to 0.29 strain at 811.2°K.

        e. Crept to 0.26 strain at 1008°K.

        f. Crept to 0.32 strain at 1147°K.

Figure 8. Contribution to the Apparent Activation Energy for Creep Due to the Change of Shear Modulus with Temperature. (Data from Reference 12.)

Figure 9. Apparent Activation Energies Corrected for Changes of Shear Modulus.

Figure 10. Effect of Stress on Creep Activation Energy.

Figure 11. Variation of H_D with Temperature.

Figure 12. The Diffusion Coefficient of α and β Irons as a Function of Temperature.

Figure 13. Effect of Stress on the Density of Mobile Dislocations During

Secondary Creep.