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HEXAGONAL INTERMEDIATE PHASE

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Authors

Howard, Eugenia M.

Barmore, Willis L.

Mote, Jim D.

et al.

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Berkeley, California

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ON THE THERMALLY-ACTIVATED MECHANISM
OF PRISMATIC SLIP IN THE Ag-Al
HEXAGONAL INTERMEDIATE PHASE

Eugenia M. Howard, Willis L. Barmore, Jim D. Mote and John E. Dorn

December 10, 1962

"On the Thermally-Activated Mechanism
of
Prismatic Slip in the Ag-Al
Hexagonal Intermediate Phase"

by

Eugenia M. Howard,¹ Willis L. Barmore,²

Jim D. Mote³ and John E. Dorn⁴

December 10, 1962

¹ Formerly Research Assistant, Inorganic Materials Research Division, Lawrence Radiation Laboratory, University of California, Berkeley, California; now, Chemist, U.S. Bureau of Mines, University of California, Berkeley, California.

² Formerly Metallurgist, Institute of Engineering Research, University of California, Berkeley; now, Research Metallurgist, Lawrence Radiation Laboratory, Livermore, California.

³ Research Metallurgist, Inorganic Materials Research Division, University of California, Berkeley, California.

⁴ Miller Professor of Materials Science in the Department of Mineral Technology, and Research Metallurgist, Inorganic Materials Research Division, Lawrence Radiation Laboratory, University of California, Berkeley.

ABSTRACT

The effect of stress and temperature on the creep rate of Ag-33 atomic % Al and Ag-33 atomic % Al-1 atomic % Zn for prismatic slip was determined over a range of temperatures from 440° to 700°K. The alloys exhibited a brief initial inverted transient creep over which the creep rate increased slightly to a steady state value which depended only on the stress and temperature, regardless of the previous strain history. The steady state creep rates were found to be given by:

$$\dot{\gamma} = (1.4 \pm 0.3) \tau^{3.6 \pm 0.1} e^{-\frac{33,000}{RT}} \quad \text{for Ag-Al}$$

and

$$\dot{\gamma} = (0.08 \pm 0.02) \tau^{4.0 \pm 0.5} e^{-\frac{33,000}{RT}} \quad \text{for Ag-Al-Zn}$$

where the activation energy is independent of the stress and strain and agrees fairly well with that estimated for diffusion.

The observed creep behavior was shown to be in qualitative agreement with a model based on the stress-induced local disordering in the vicinity of a dislocation.

INTRODUCTION

Over the past fifteen years considerable progress has been made in correlating the experimentally observed plastic behavior of pure metals over each temperature range with one or more of a series of theoretical strain-rate controlling dislocation mechanisms. Despite the current interest in the plastic behavior of intermediate phases and intermetallic compounds, however, very little has been done in identifying the operative dislocation mechanisms in these materials. It is anticipated that the plastic behavior of intermediate phases can be rationalized in terms of the same dislocation mechanisms that are appropriate to pure metals and alpha solid solutions, provided the additional complexities arising from crystal structure, bonding, long-range order, short-range order and defect lattices are taken into account.

As part of a more extensive investigation on the plastic behavior of intermediate phases, Mote, Tanaka, and Dorn⁽¹⁾ recently reported on the effect of temperature on the yield strength for prismatic slip in the hexagonal Ag-33 atomic percent Al intermediate phase as recorded in Fig. 1: Over Region I the deformation was observed to take place as a result of the thermally activated Peierls mechanism, and over Region II the athermal disordering mechanism applied. The experimental data at that time, however, were not sufficiently extensive to identify the thermally activated mechanisms operative over Regions III and IV. It is the purpose of this investigation to attempt to identify the mechanisms that are operative over these high temperature regions. It will be shown that the same mechanism

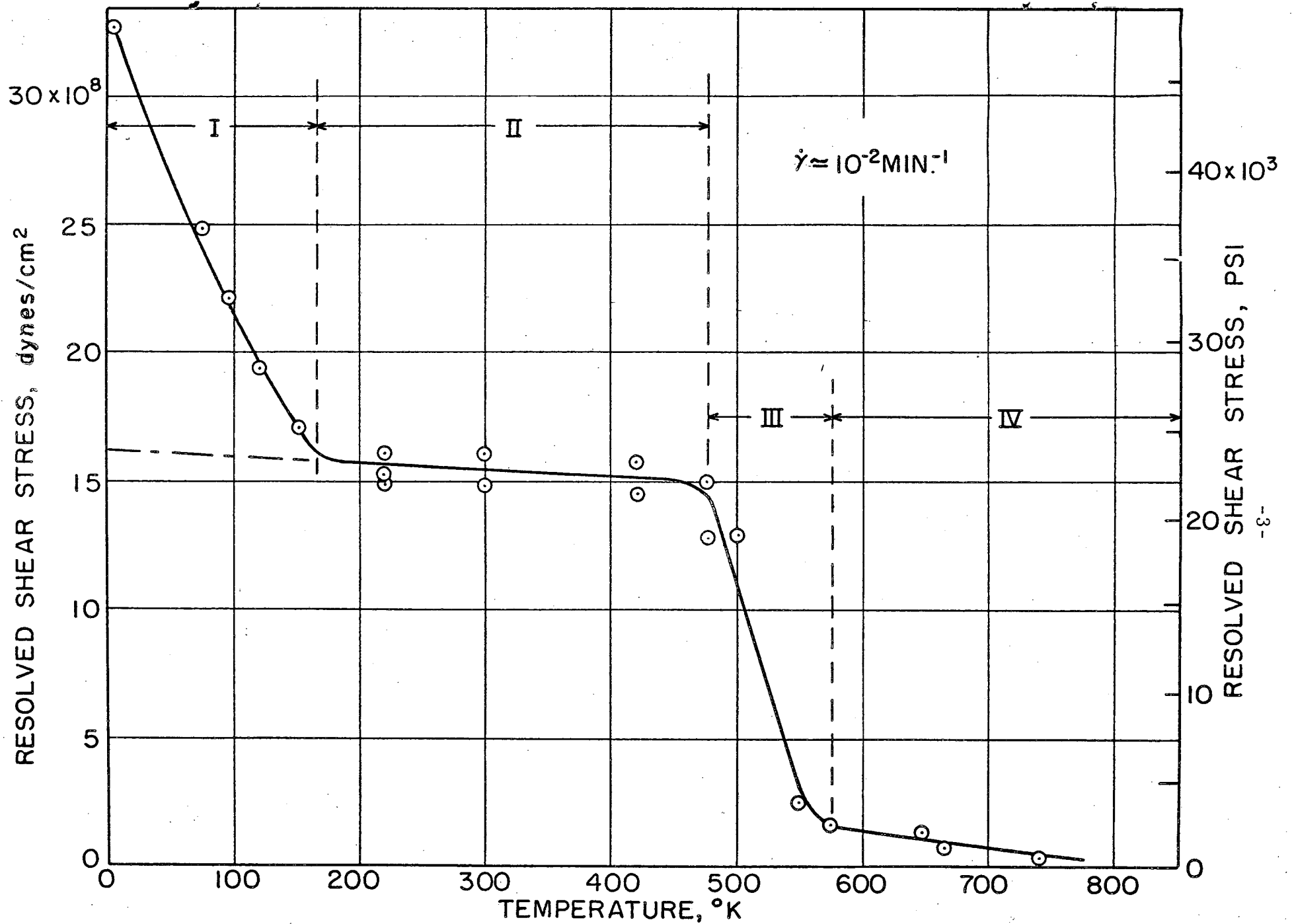


FIG. I. EFFECT OF TEMPERATURE ON THE CRITICAL RESOLVED SHEAR STRESS FOR PRISMATIC SLIP OF Ag - 33 AT. % Al SINGLE CRYSTALS.

operates over Region III and IV, although the activation energy for this mechanism is in good agreement with that for diffusion, the process cannot be attributed either to the motion of jogged screw dislocations or the climb of edge dislocations. The results, however, are in good qualitative agreement with stress induced disordering localized at dislocations.

EXPERIMENTAL PROCEDURE

The alloys studied in this investigation were prepared from silver, aluminum, and zinc (each of purity greater than 99.995%) by induction melting under an argon atmosphere and chill casting in a copper mold. Rectangular bar single crystal specimens (1/4 x 1/8 inch in cross-section) of Ag-33 at. % Al were grown in vacuum by modified Bridgman technique. Similar Ag-33 at. % Zn specimens were produced by an identical technique employing an argon atmosphere. The specimens were so oriented that the angle $90 - \chi_0$ between the tensile axis and the normal to the prismatic plane and the angle λ_0 between the tensile axis and the slip direction were 45 ± 1 deg.

The effect of a change in temperature on the creep rate and a change in stress on the creep rate were determined over the temperature range $440^\circ - 700^\circ\text{K}$. The temperature of the specimen was controlled by means of a constant temperature bath to within ± 0.5 deg. K. of the reported values. The bath, a molten mixture of sodium nitrate and sodium nitrite, was stirred vigorously in order to maintain a uniform temperature throughout. Temperature control was effected

by means of the signal from a ten-unit thermopile in conjunction with a Leeds and Northrup Micromax recorder-controller. Temperature changes were made by quickly removing one bath and immersing the specimen in another bath at a preset temperature.

The extension of the specimen was autographically recorded by monitoring the signal generated by the relative displacement between the body and the core of a linear variable differential transformer. The body of the transformer was affixed to the upper extremity of the gage section and the core to the lower extremity of the gage length by means of vycor tubes. Strains were measured to the nearest ± 0.0001 .

EXPERIMENTAL RESULTS

Unlike pure metals which usually exhibit three regions of creep, (i. e. an initial transient region characterized by a decelerating creep rate as the material work hardens, a steady state region wherein the creep rate is constant, and a tertiary region in which the creep rate accelerates until fracture), the present alloys exhibited only two regions of creep, i. e. an inverse transient characterized by an accelerating creep rate and a secondary region over which creep rate remained constant. A typical creep curve for these alloys is shown in Fig. 2.

In pure metals a decrease in stress produces a transient characterized by recovery, as indicated by an increasing creep rate, whereas in this alloy a decrease in stress produces a transient characterized by a decreasing creep rate. An example of this phenomena is shown in Fig. 3. After a stress decrease of approximately two thirds the initial

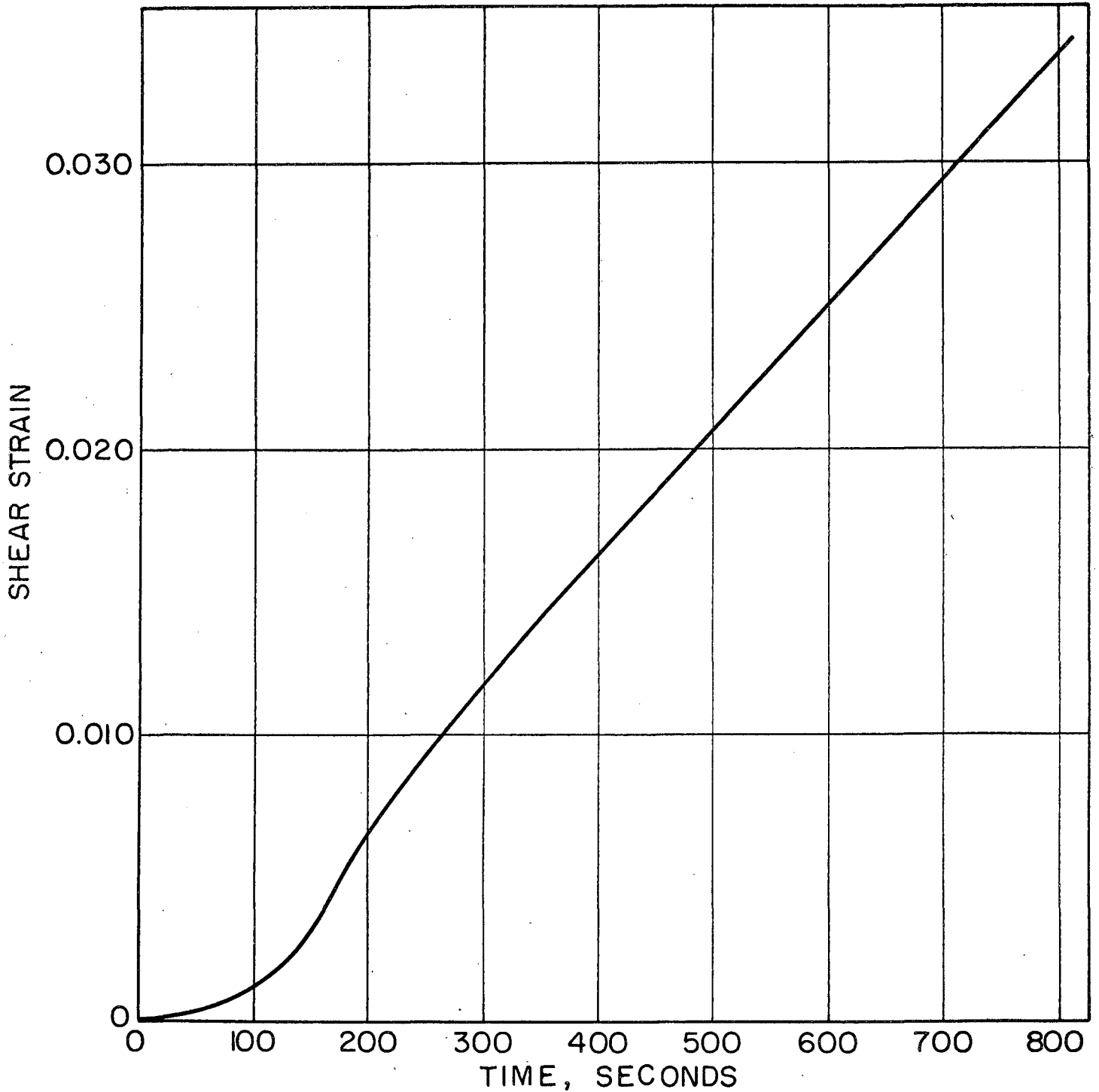


FIG. 2. TYPICAL CREEP FOR Ag-33 AT. % SINGLE CRYSTALS ORIENTED FOR PRISMATIC SLIP.

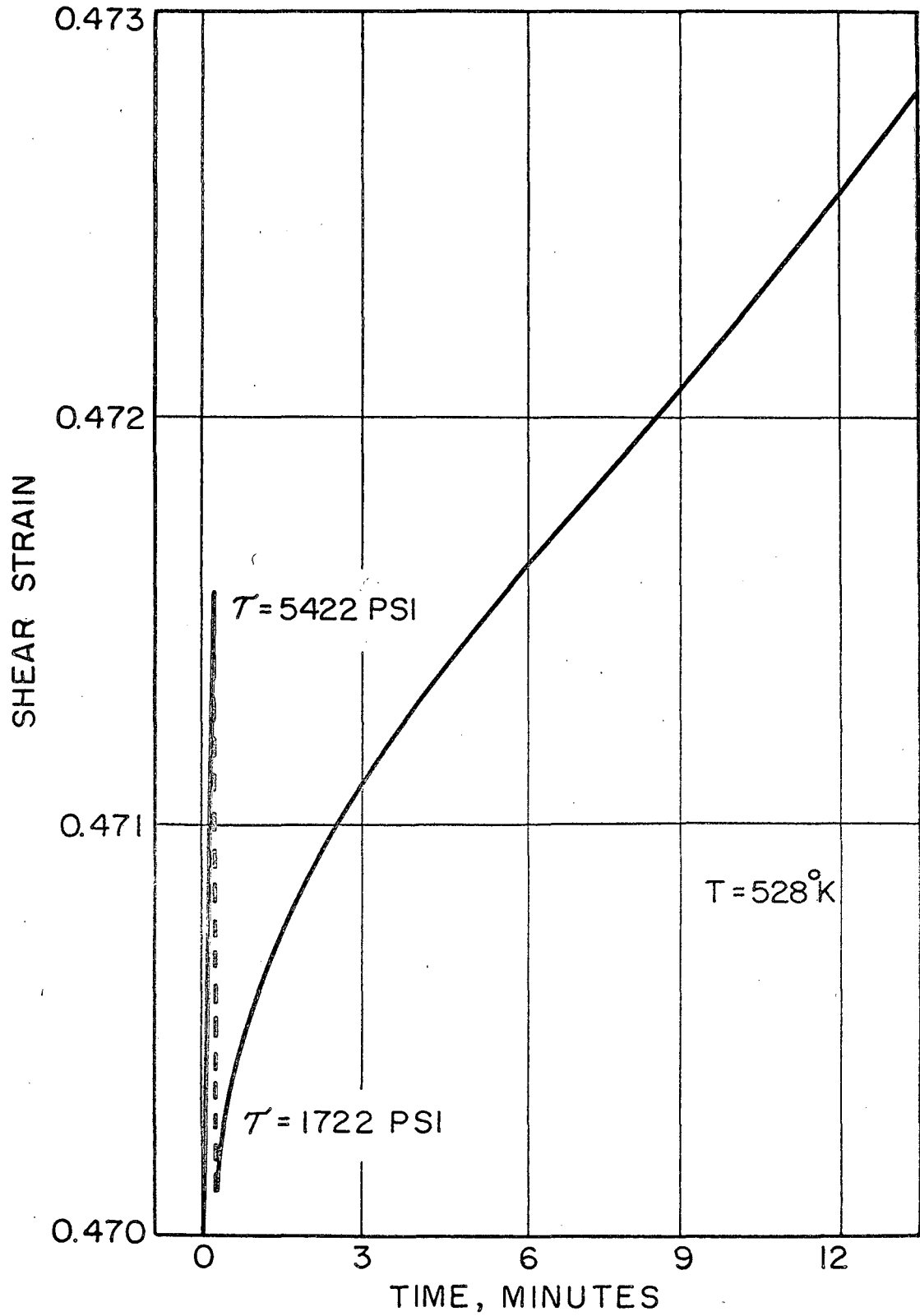


FIG. 3. DECELERATING CREEP OF Ag-33 AT. % Al AFTER STRESS-DECREASE.

value, the strain rate decreased to a steady-state value over a period of about 10 minutes. Increases in stress produce transients characterized by a transient increase in creep rate.

A log-log plot of the effect of the resolved shear stress, τ , on the steady state creep rate, $\dot{\gamma}$, of Ag-33 at. % Al single crystals oriented for prismatic slip is shown in Fig. 4, for two different temperatures. These data were obtained for both decreases and increases of stress and were found to be independent of strain. Similar data are shown in Fig. 5 for both Ag-Al-Zn single crystals and Ag-Al single crystals for various temperatures. Fig. 5 reveals that the data for the Ag-Al-Zn single crystals gives lines of slightly greater slope than the data for Ag-Al single crystals. One possible reason for this difference is that the Ag-Al-Zn single crystals, although produced under an argon atmosphere, contained inclusions whereas the Ag-Al single crystals, produced under vacuum, were free from any inclusions.

The data of figs. 4 and 5 reveal a power dependence of the stress on the strain rate. Fig. 6 is a plot of the slopes of curves of Fig. 4 and 5 vs. the absolute temperature, T. These data reveal that

$$\dot{\gamma} \propto \tau^{3.6 \pm 0.1}$$

for Ag-Al, and

$$\dot{\gamma} \propto \tau^{4.0 \pm 0.5}$$

for Ag-Al-Zn, both relationships being independent of the test temperature.

The temperature dependence of the strain rate is shown in Fig. 7 where, $\ln \dot{\gamma}$ is plotted vs. $\frac{1}{RT}$ where R is the gas constant. The fact that this plot results in a straight line reveals that $\dot{\gamma}$ is proportional to $\exp^{-Q/RT}$ where Q, the activation energy, is independent

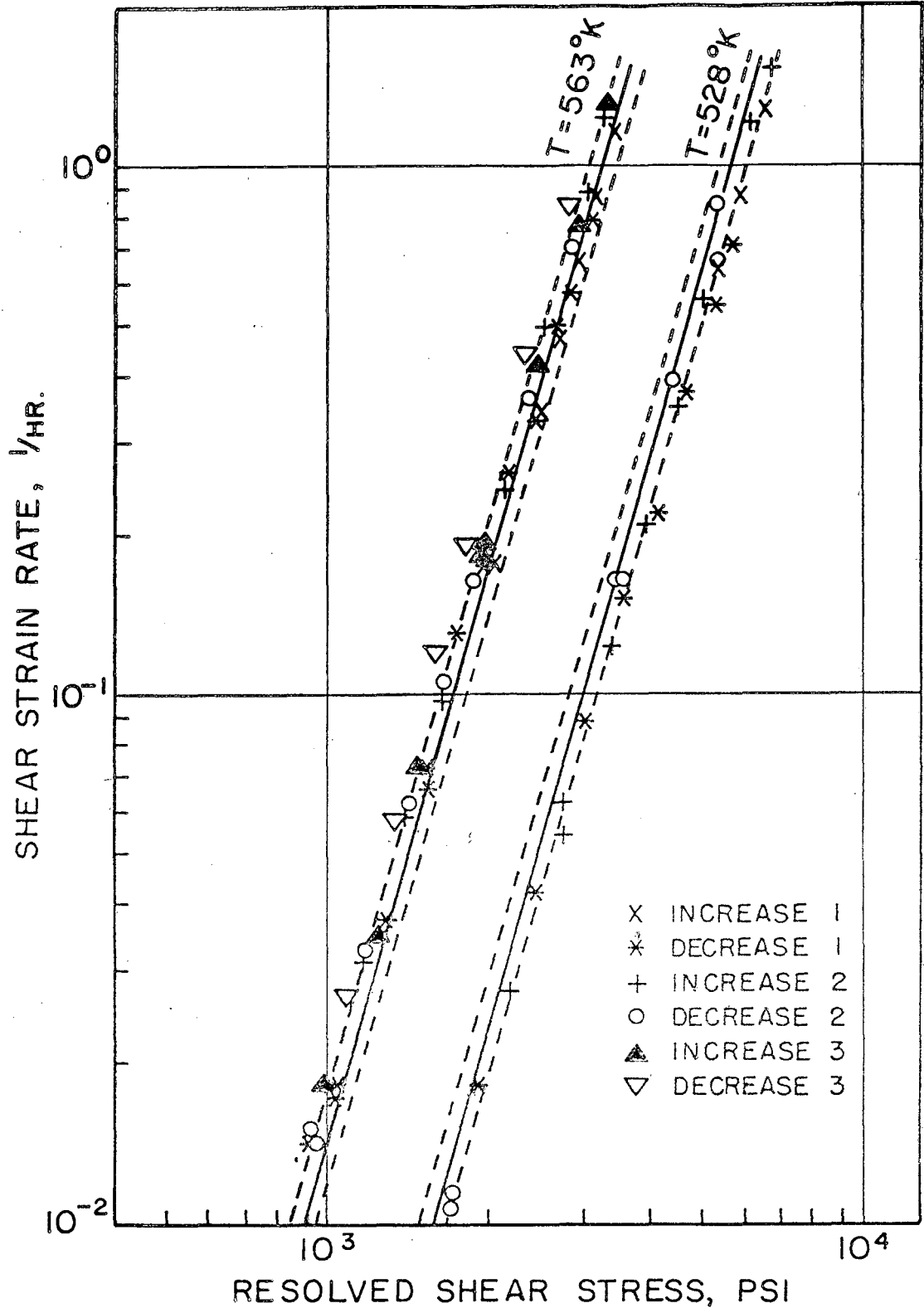


FIG. 4. TWO SETS OF DATA POINTS FOR PRISMATIC SLIP OF Ag-Al ALLOY AT CONSTANT TEMPERATURES OF 528°K AND 563°K . THE SOLID LINE IS THE EMPIRICAL EQUATION DERIVED FROM THE DATA, AND THE DOTTED LINES SHOW THE MAGNITUDE FOR THE ERROR IN THE CONSTANT FOR THIS EQUATION.

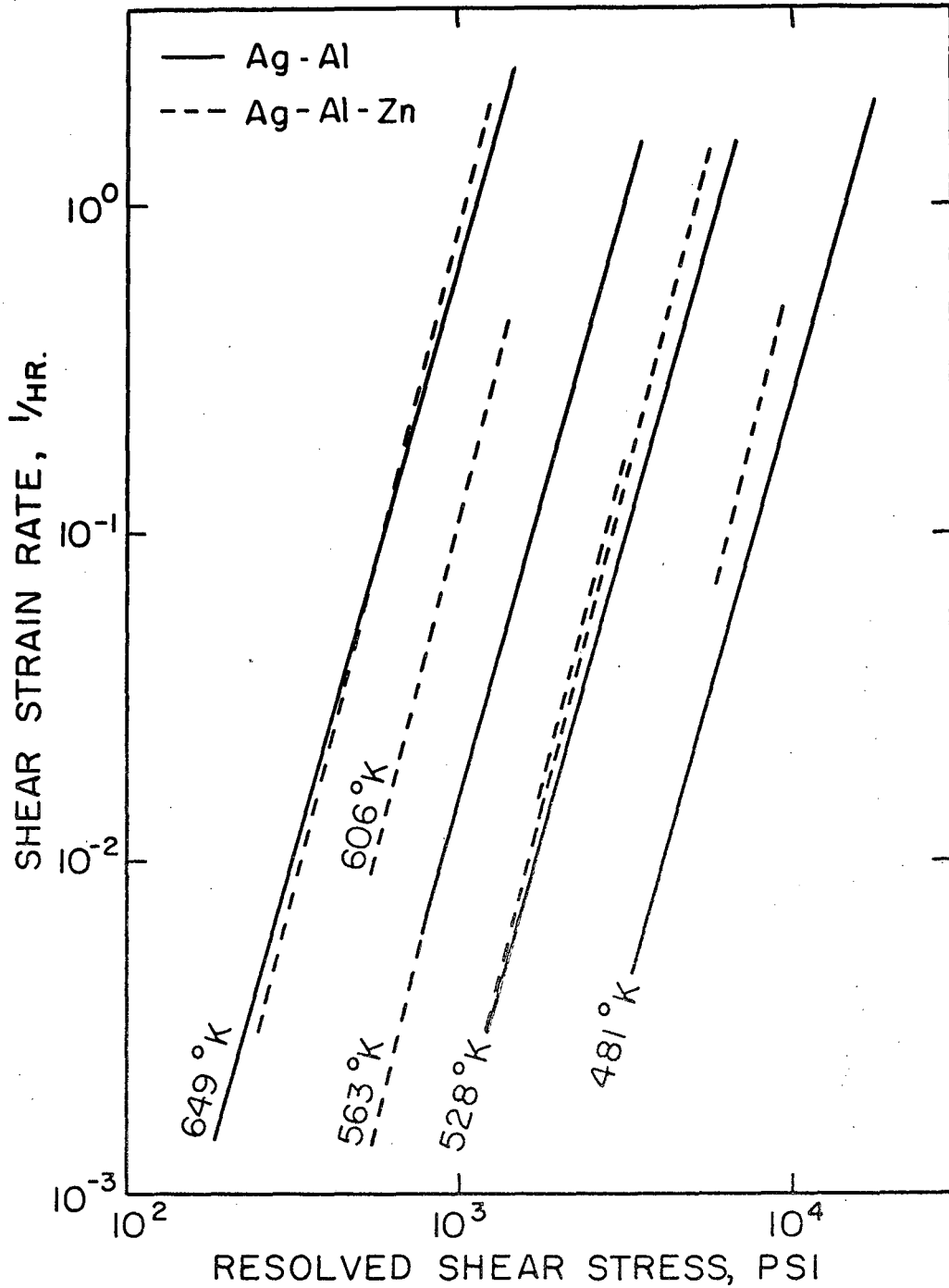


FIG. 5. FAMILY OF LINES REPRESENTING DATA DETERMINED IN THE STEADY STATE REGION OF CREEP.

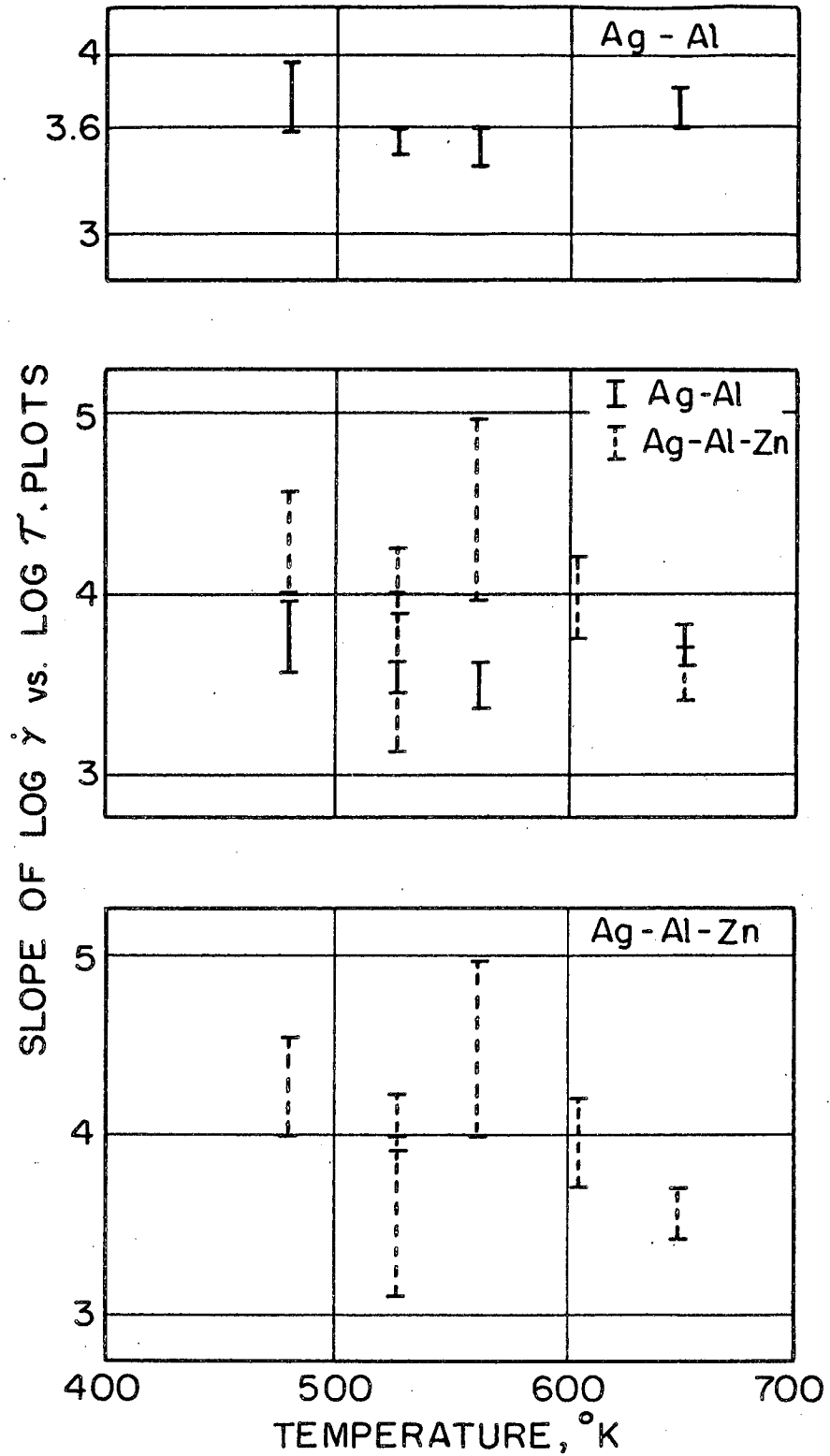


FIG. 6. COMPARISON OF SLOPES OF LOG $\dot{\gamma}$ vs LOG τ PLOTS FOR BOTH ALLOYS AS A FUNCTION OF TEMPERATURE.

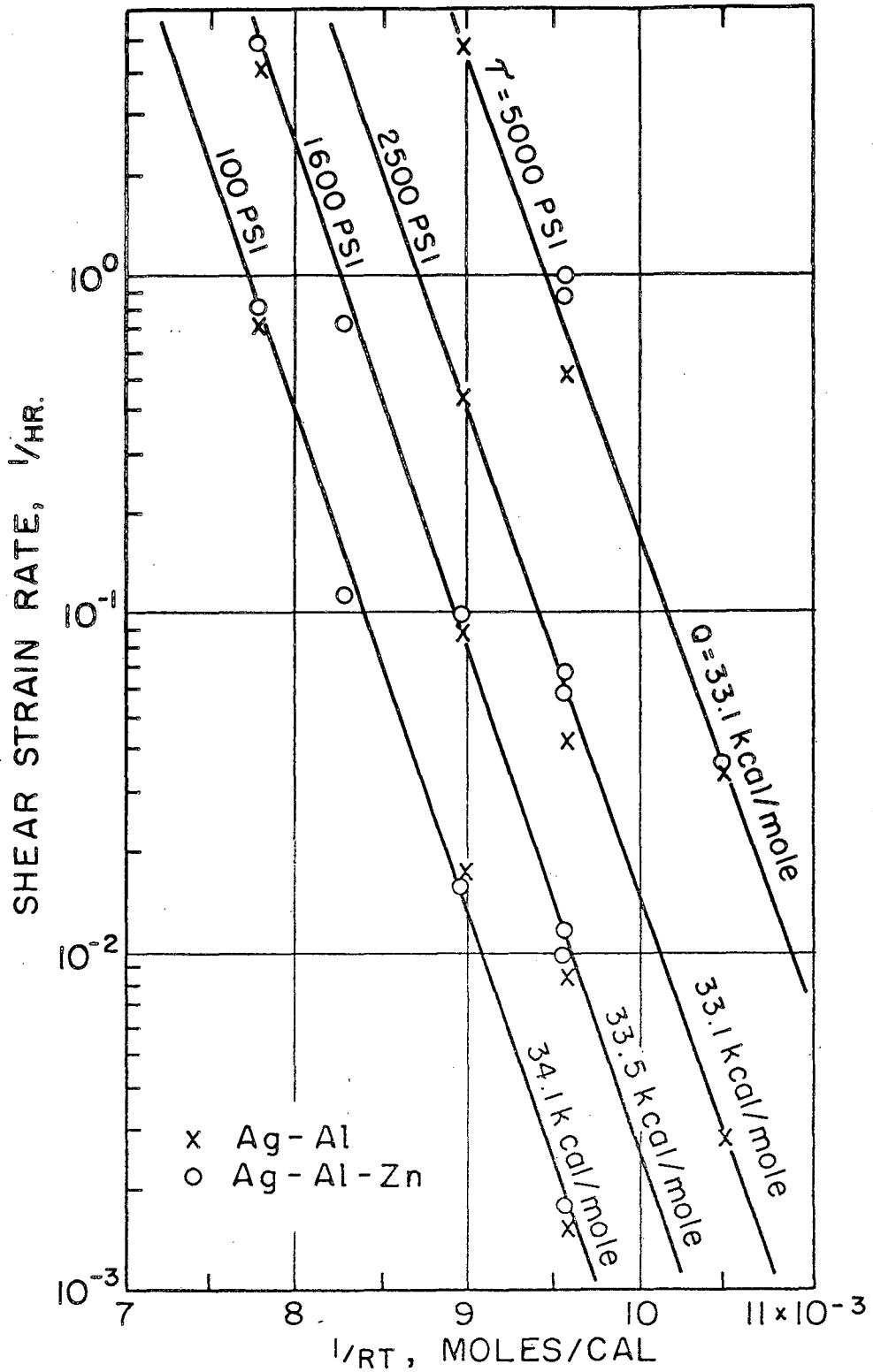


FIG. 7. DETERMINATION OF THE APPARENT ACTIVATION ENERGY Q FOR PRISMATIC SLIP OF Ag-Al AND Ag-Al-Zn SINGLE CRYSTALS.

of the stress. The slope of the line in Fig. 7 gives an average activation energy of $3.3 \times 10^4 \pm 0.1 \times 10^4$ cal./mole. There is no observable difference between the activation energies for the two alloys.

The steady state creep rates for the two alloys are as follows:

$$\dot{\gamma} = (1.4 \pm 0.3) \tau^{3.6 \pm 0.1} e^{-\frac{Q}{RT}} \text{ per sec.} \quad (1a)$$

for Ag-Al and

$$\dot{\gamma} = (0.08 \pm 0.02) \tau^{4.0 \pm 0.5} e^{-\frac{Q}{RT}} \text{ per sec.} \quad (1b)$$

for Ag-Al-Zn.

Where creep is due to a single thermally activated process, the strain rate may frequently be described by an equation of the form

$$\dot{\gamma} = K e^{-\frac{Q}{RT}} \quad (2)$$

where K may be a function of stress and structure. Therefore an alternate technique may be used to determine the activation energy. Since the substructure just preceding and immediately following rapid changes in temperature is essentially the same,

$$R \ln \frac{\dot{\gamma}_1}{\dot{\gamma}_2} = Q \left(\frac{1}{T_2} - \frac{1}{T_1} \right) \quad (3)$$

where $\dot{\gamma}_1$ and $\dot{\gamma}_2$ refer to the creep rates at T_1 and T_2 , respectively, just before and immediately following a rapid change in temperature.

A typical creep curve for the Ag-Al-Zn alloy, employing the change in temperature technique, is shown in Fig. 8. The effect of strain on the activation energy is shown in Fig. 9, and the activation energy as a function of temperature is shown in Fig. 10. The value of the activation

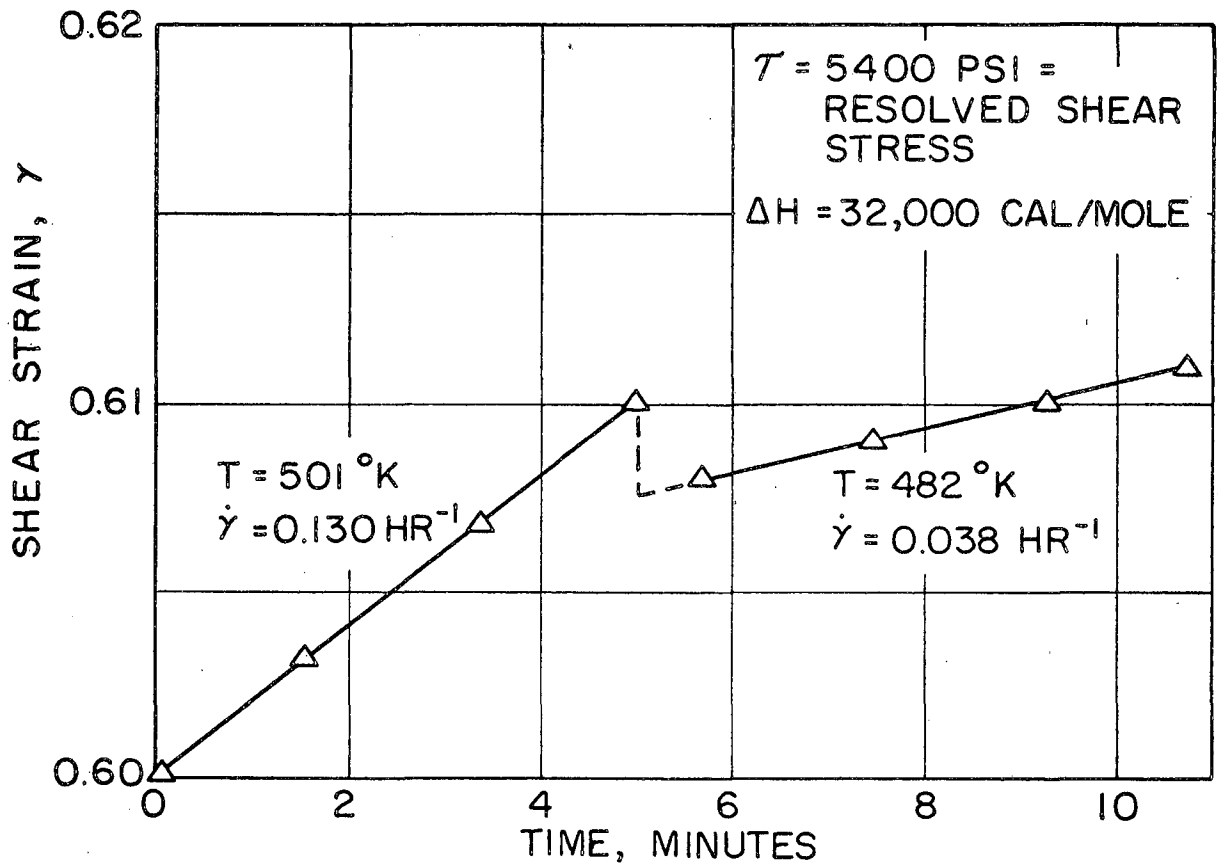


FIG. 8. THE EFFECT OF AN ABRUPT TEMPERATURE CHANGE ON THE TYPICAL CREEP CURVE OF Ag-Al-Zn SINGLE CRYSTAL ORIENTED FOR PRISMATIC SLIP.

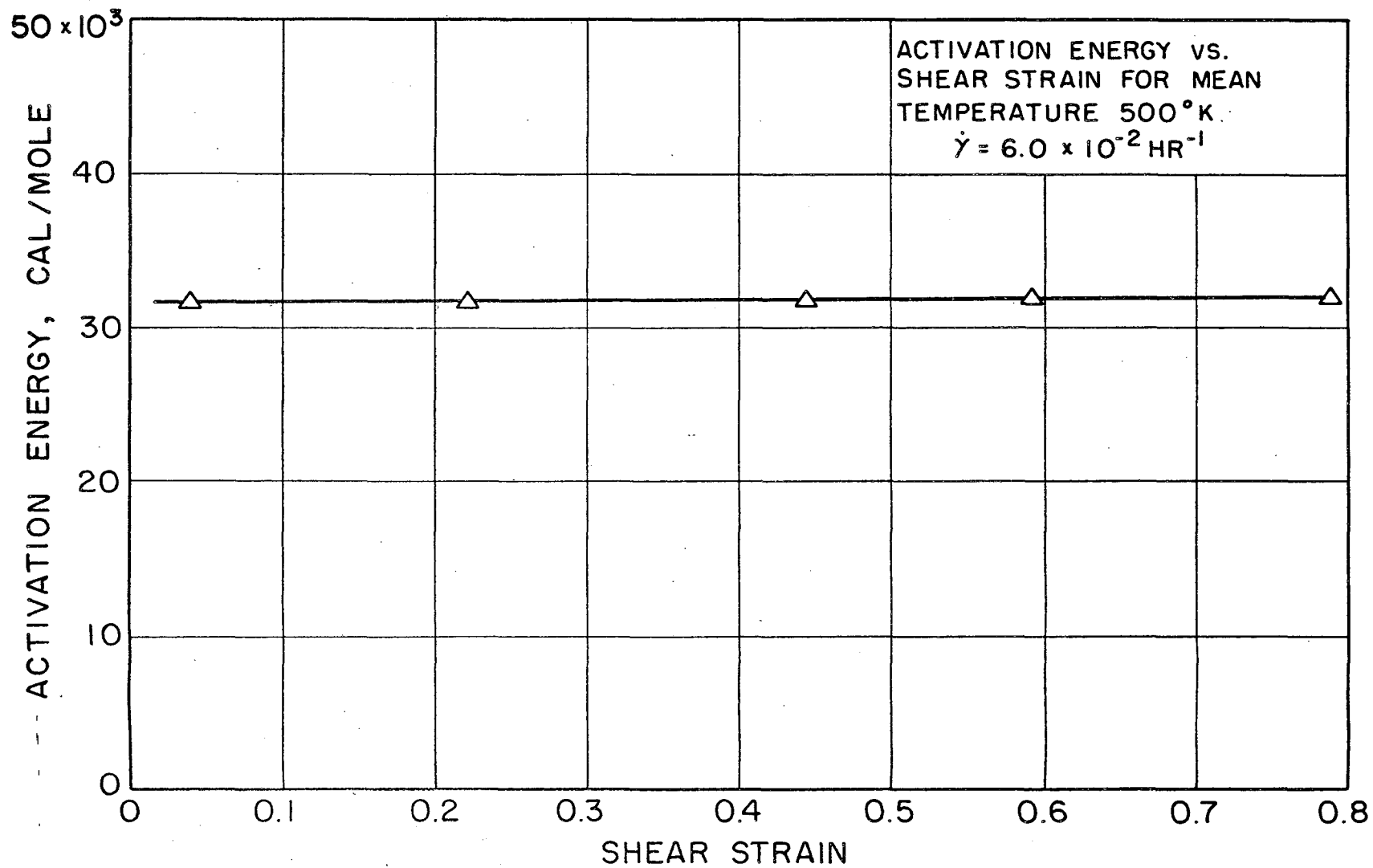


FIG. 9. EFFECT OF SHEAR STRAIN ON THE AVERAGE ACTIVATION ENERGY FOR PRISMATIC SLIP.

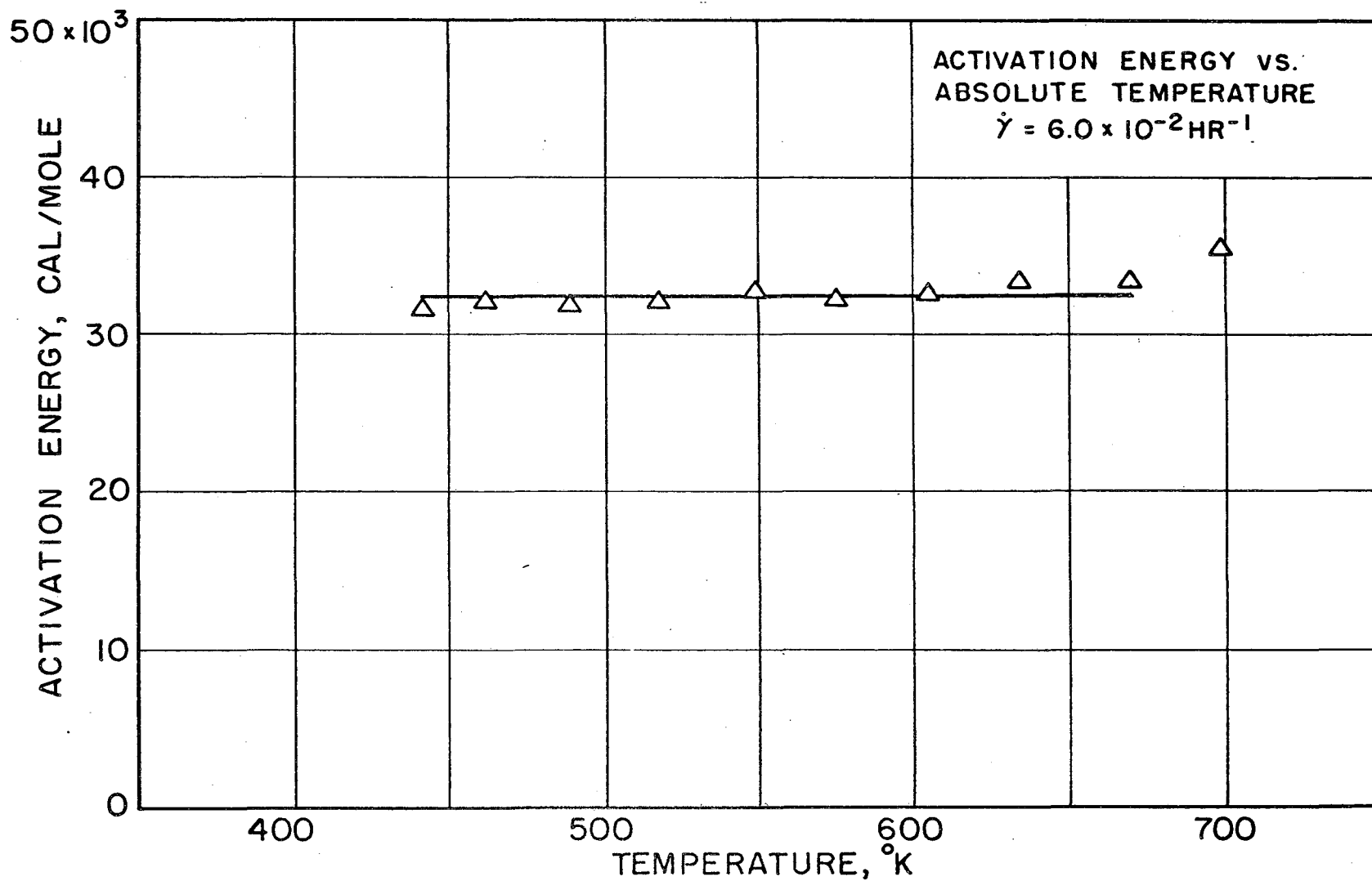


FIG. 10. EFFECT OF TEMPERATURE ON THE AVERAGE ACTIVATION ENERGY FOR PRISMATIC SLIP.

energy, as determined by this technique, is essentially the same as that determined from Fig. 6.

Although Eqn. 1a was deduced from experimental results over the temperature range $475^{\circ} - 650^{\circ}\text{K.}$, it gives the correct stress-temperature dependence at a constant strain rate over the entire temperature range of $475^{\circ} - 800^{\circ}\text{K.}$, as illustrated in Fig. 11. This reveals that the same deformation mechanism is operative over Regions III and IV.

DISCUSSION

In many respects the creep of the hexagonal Ag-Al phase described here is quite simple. First, the usual primary stage of decreasing creep rates, indicative of the formation of more creep resistant dislocation substructural modifications, is absent. In lieu of the usual primary stage of creep, an inverted transient of short duration is noted, over which the creep rate accelerates slightly to its steady state value. When the stress is reduced, a reverse inverted transient is observed; the creep rate at the reduced stress decreases with time until its steady state value for this stress is reached. The duration of these transients becomes greater at the lower test temperatures. Although such transients exist, the steady state creep rate is a constant, independent of the duration of the test or the preceding strain history. It must therefore be concluded that such modifications of dislocation arrangements, which might attend creep in this phase, have no effect on the creep rate; or, alternately, only minor substructural changes take place during creep.

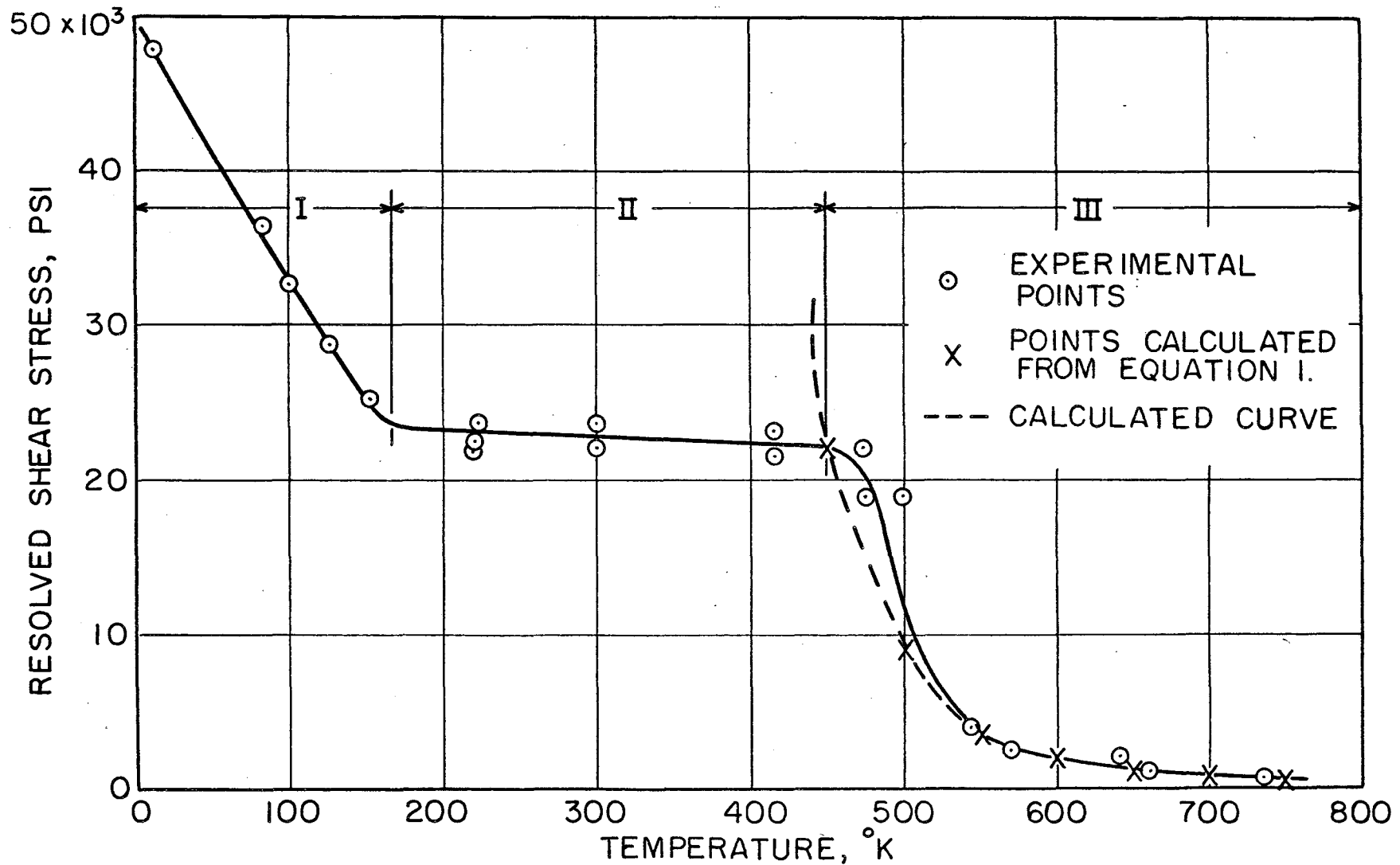


FIG. II. PRISMATIC YIELD STRESS AS A FUNCTION OF TEMPERATURE FOR SINGLE CRYSTALS OF THE ALLOY $33\frac{1}{3}$ ATOMIC PERCENT ALUMINUM IN SILVER, AFTER MOTE, TANAKA AND DORN².

The activation energy for prismatic creep of the Al-Ag intermediate phase is insensitive to the stress over wide ranges of the creep rate; it has the constant value of about 33,000 cal. /mole, which is estimated to approximate that for diffusion. Although no data have been found in the literature on the activation energy for diffusion of either Ag or Al in this phase, the empirical method of estimating this quantity, developed by Sherby and Simnad, ⁽²⁾ gives 36,900* cal. /mole. The fair correlation between the activation energy for creep and the estimated activation energy for diffusion suggests that the creep under consideration here is determined by some diffusion-controlled dislocation mechanism. Whereas the preexponential term of Eqn. 1, namely $A \hat{\tau}^n$, accurately describes the simple effect of stress on the creep rate, the term A may yet depend mildly on the temperature, a possibility which is not easily established experimentally because of the overwhelming effect of the temperature in the exponential term of the creep-rate equation.

Several alternate mechanisms might be responsible for the observed creep behavior of the Al-Ag intermediate phase under prismatic slip: These are (1) climb of dislocations, (2) viscous drag of dislocations, (3) fluctuations in short-range order, (4) stress-induced diffusion-controlled disordering. Other commonly observed mechanisms

*From the Sherby and Simnad relation

$$Q_{S.D.} = RT_m (K_o + V)$$

where T_m = the melting temperature

K_o = a constant = 17 for HCP metals

and V = the valence

one obtains

$$Q_{S.D.} \approx 36,900$$

An average valence $V = 1.67$ for the compound was obtained by assuming 5 conduction electrons per unit cell and 2 Ag + 1 Al atom per unit cell.

such as the Peierls mechanism, intersection of dislocations, cross slip, etc., can be summarily discarded because they exhibit activation energies that are sensitive to the applied stress.

1. Dislocation Climb: Although, as shown in several papers by Weertman,⁽³⁾ the formal dependency of the strain rate on stress and temperature, as given by Eqn. 1, is in good agreement with the expected trends for operation of a dislocation climb mechanism, several other factors serve to disqualify the identification of the creep mechanism observed here with that of climb. In the primary stage of creep by climb, edge dislocations become arrested and thereby induce back stress fields on other moving dislocations. Thus the climb mechanism is characterized by a primary stage of creep over which the creep rate continues to decrease. Since the creep rate in the present case increased over the initial period, the operative mechanism cannot be that of climb. Furthermore the independence of the prismatic flow stress of this alloy over the range from 190^o to 475^oK., as shown in Fig. 1, reveals that an athermal dislocation mechanism controls the deformation over this range. Such a behavior could be due to long-range back stresses,⁽⁴⁾ Suzuki locking,⁽⁵⁾ long-range ordering,^(6, 7) or short-range ordering.⁽⁸⁾ In the absence of deformation, long-range back stresses could not be great enough to account for the very high deformation stress that was observed; no Suzuki locking can occur because the dislocations on the prismatic planes are undissociated; and detailed studies⁽⁹⁾ have shown that this alloy does not exhibit long-range ordering. Consequently the intermediate-temperature prismatic slip of this alloy has been ascribed to short-range ordering. Only on the basis of this hypothesis has it been possible to account for the high deformation stress

that is observed for this alloy. At higher temperatures, somewhat above 475°K., however, the flow stress decreased with increasing temperature, revealing, in this way, the introduction of a thermally activated dislocation mechanism. But the actual flow stress that was obtained in this higher temperature region was much smaller than that required to disorder the alloy across the slip plane. Consequently if the equilibrium degree of order for this temperature range had been maintained, no creep (by climb or any other thermally activated process) could have taken place at the stress levels that were obtained for flow. Consequently the climb mechanism cannot be responsible for the observed creep by prismatic slip in the Ag-Al intermediate phase. Similar arguments disqualify the motion of jogged screw dislocations.

2. Viscous Drag of Dislocations: Several details of the experimental observations are in good qualitative agreement with the hypothesis that the creep rate for the Ag-Al intermediate phase under discussion here arises from the viscous drag of a Cottrell atmosphere⁽¹⁰⁾ on the moving dislocation. Not only is the activation energy correct, but the stress law, as deduced by Weertman⁽¹¹⁾ is also about right. Furthermore, the initial transients in the creep rate due to changes in stress that were observed can be accounted for, at least qualitatively, by this mechanism. Two significant factors, however, demand that the observed creep cannot be ascribed to this mechanism. First, the atomic radii of Al and Ag are almost identical and the Cottrell interaction therefore must be extremely small. Second, the same arguments based on short-range order strengthening, that were used to disqualify the climb mechanism, equally serve to disqualify the viscous drag process. Consequently all acceptable mechanisms must involve, in one way or another, thermally

activated processes that result in reductions of short-range order.

Two such processes can be conceived, namely, fluctuations in composition and short-range order, and alternately, stress induced disordering. We now discuss each of these in sequence.

3. Fluctuations in Short-Range Order: The stress, τ_0 , required to move a dislocation on the prismatic plane has been shown to be

$$\tau_0 = \frac{8 m_A m_B}{1.61 a_0^3} \alpha_0 \epsilon \quad (4)$$

where a_0 = the lattice constant

m_A = mole fraction of A atoms

m_B = mole fraction of B atoms

α_0 = Cowley's degree of nearest neighbor order

ϵ = the ordering energy

At temperatures below about one-half of the melting temperature where diffusion is extremely slow, α_0 refers to the frozen-in degree of order. In this range therefore, τ_0 is insensitive to the temperature. Above about one-half of the melting temperature, where diffusion is reasonably rapid, α_0 refers to the equilibrium degree of order. Here α_0 decreases rather gradually with an increase in temperature. In this region, however, flow can occur as a result of fluctuations in short range order and composition even when the applied stress, τ , is less than τ_0 . For example, when the degree of order over a region across the slip plane just in advance of a dislocation fluctuates from the average equilibrium value of α_0 to a value less than α , where α is the degree of order that will just permit slip for an applied stress, τ , the dislocation will move into the region of the fluctuation. An accurate analysis of this

problem has not yet been accomplished, but preliminary estimates suggest that the observed prismatic creep of the Al-Ag intermetallic does not occur by this mechanism. Only a qualitative description of some significant issues need be given here. First, the creep rate by this model would be proportional to the probability, p , for such a fluctuation times a frequency factor, $\nu e^{-\frac{U_D}{kT}}$, namely

$$\dot{\gamma} \propto p \nu e^{-\frac{U_D}{kT}}$$

where U_D is the activation energy for diffusion. Since the probability, p , is also very sensitive to the temperature, the apparent activation energy for creep by this model is complicated, being much greater than that for diffusion, and decreasing rather rapidly with an increase in temperature. Furthermore, as the applied stress is decreased, larger fluctuations with much lower probabilities would be required, suggesting a much more rapid decrease in creep rate with decrease in stress than was observed experimentally. It therefore appears to the authors that fluctuations in short-range order contribute only infinitesimally to the observed creep rate.

4. Stress Induced Disordering: The theory for stress induced disordering, as this term will be used here, is also very difficult to establish in quantitative form. But certain general aspects of the subject can be set forth at present, in sufficient semi-quantitative detail to permit a preliminary understanding of the major issues.

The details of short-range order strengthening at high temperatures, where diffusion occurs reasonably rapidly, is somewhat different from those that pertain to low temperatures where the short-range order is frozen in the lattice. When a dislocation is moved through the lattice at low temperatures, it destroys the frozen-in local order across the

slip plane in good agreement with the assumptions made in Fisher's⁽⁶⁾ model of the process. But due to diffusion at high temperatures, the local degree of order in the vicinity of the core of a dislocation must differ somewhat from that which exists in the ideal regions of the crystal. This deduction is obvious for the case where Cottrell atmospheres are formed about the dislocations. But even in such cases as the present one, where the dislocation-atom interactions are so small that the influence of the Cottrell type of segregation about the core of a dislocation is negligible, the degree of order, defined in the usual way, must in general differ near the core of a dislocation from its value some distance from the dislocation. Three factors contribute to this difference: First, the coordination number at the core differs from that in the lattice; second, the bond energies for each type of bond can change somewhat differently with bond distance; and third, in cases of complete or partial covalent bonding, the bond energy will depend on the bond angles.

Neglecting for the present the added influence of stress-induced disordering, we arrive at the following qualitative description of this yet incomplete model. When a dislocation is moved so rapidly through the crystal that reordering at the dislocation core as a result diffusion cannot keep pace, the Fisher type of analysis applies, the phenomenon of plastic deformation per se being athermal. At very slow rates of dislocation motion, almost complete equilibrium will be maintained at the moving dislocation core. Under these conditions a slightly greater increase in free energy attends the motion of the dislocation and the flow stress will therefore be slightly greater than that given by the Fisher analysis. But, for this model, which neglects the effects of stress-induced disordering, the plastic deformation per se remains

athermal. It should, however, exhibit a yield stress and perhaps the Portevin-Le Chatelier effect under appropriate ranges of temperature and strain rate.

When we impose upon this model the concept of stress induced disordering, the plastic behavior acquires the characteristics appropriate for a thermally activated mechanism. The association of disordering across the slip plane with a stress high enough to move a dislocation several or more Burgers vectors has already been emphasized. Consider now applying a stress τ less than τ_0 , as defined by Eqn. 2. Under these conditions, the core of the dislocation will initially move less than a Burgers vector. Bond lengths and angles will change, particularly for atoms near the core of the dislocation. Under this condition, at high enough temperatures, atoms will so diffuse at the core of the dislocation and in the surrounding region to reduce the strain energy. Consequently, the core of the dislocation will then be displaced farther in the direction of motion of the dislocation. This process will continue, especially when the Peierls stress is very low and when the temperature is high. Finally a local order about the core of the dislocation will be reached, which will permit the completion of the motion of the core by one Burgers vector. The original displacement of the core of the dislocation will increase with increased stressing, whereas the displacement at the break away point will decrease with increasing values of the applied stress. Therefore the distance over which the core must move from its original displacement to its break away position will be quite sensitive to the applied stress. Consequently the rate of displacement of the core, which is diffusion controlled, is given by

$$\frac{dx}{dt} = f\{x\} e^{-\frac{U_D}{kT}}$$

and consequently the wait time, t , for break away is determined from

$$\int_{x_i}^{x_b} \frac{dx}{f\{\tau\}} = \frac{1}{\psi\{\tau\}} = e^{-\frac{u_D}{kT}} t$$

where x_i and x_b , the initial and break away displacements of the dislocation, are functions of the applied stress, τ . According to this formulation, the mean dislocation velocity is $v = b/t$ and the creep rate becomes

$$\dot{\gamma} = \rho b v = \rho b^2 \psi\{\tau\} e^{-\frac{u_D}{kT}} \quad (5)$$

which is at least in qualitative agreement with the experimentally determined creep law if $\psi\{\tau\}$ is associated with $A/\rho b^2 \tau^n$. The observed inverted transients obtained upon change in stress are also consistent with this model.

Recently, Lawley, Cole and Cahn⁽¹²⁾ have shown that the creep of the short-ranged ordered alloy of Fe containing 22.0 to 25.5 at. % Al obeyed the relationship

$$\dot{\gamma} = A e^{-\frac{Q}{RT}} \tau^n \quad (6)$$

where $4.6 \leq n \leq 6.0$

which, excepting for the stress power, is quite analogous to that observed here for the prismatic slip of the Ag-Al intermediate phase; these investigators also observed initial transient creep rates in their Fe-Al alloy which were completely analogous to those observed in the Ag-Al system. They attributed their transients to a Snoeck type of stress induced ordering of the entire crystal. It appears that this explanation cannot apply to the Ag-Al alloys since the atomic radii of Ag and Al are almost identical and such an effect should indeed be negligibly small. They attributed the observed creep law to a viscous drag on the dislocations but, in view of the similarity of the atomic radii of Ag and Al, this

effect should also be small in the Ag-Al alloy. The major issue concerns the fact that the creep of Ag-Al takes place at stresses below those capable of disordering the alloy across the slip plane. For these reasons the model based on local diffusion-controlled disordering in the vicinity of a dislocation appears more appropriate than the general stress-induced ordering and viscous-drag model. The observed transients are a direct consequence of the model and therefore do not require special consideration.

CONCLUSIONS

1. Above about 475°K. the steady state creep rates for prismatic slip of the hexagonal Ag-33 atomic percent Al and Ag-33 atomic percent Al-1 atomic percent Zn intermediate phases are given by

$$\dot{\gamma} = (1.4 \pm 0.3) \tau^{3.6 \pm 0.1} e^{-\frac{33,000}{RT}}$$

and

$$\dot{\gamma} = (0.08 \pm 0.02) \tau^{4.0 \pm 0.5} e^{-\frac{33,000}{RT}}$$

respectively.

2. The creep rate cannot be attributed to the motion of jogged screw dislocations, the climb of edge dislocations or viscous drag.

3. The creep rate was attributed to localized stress-induced disordering in the immediate vicinity of a dislocation.

4. The qualitative model reveals that the activation energy should be independent of the stress and slightly less than that for self-diffusion which is in agreement with the experimental facts.

5. The experimentally observed initial inverse transient creep rates are immediate consequences of the model.

ACKNOWLEDGEMENTS

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