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Mohamed Osama Abo-el-Fotch

(M. S. thesis)

March 1967

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Contents

THE LOW TEMPERATURE THERMALLY ACTIVATED DEFORMATION MECHANISMS FOR BCC MAGNESIUM-LITHIUM-ALUMINUM ALLOY

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March, 1967

ABSTRACT

The effect of strain rate and temperature on the flow stress was
investigated in a polycrystalline magnesium-14 wt.% lithium -1.5 wt.%
aluminum alloy. Whereas the flow stress increased only slightly as the
temperature was decreased from room temperature to about 115°K, a rapid
increase in the flow stress was obtained with yet greater decreases
in temperature from about 115° to 20°K. The strong temperature and
strain rate dependence of the flow stress below 115°K was interpreted
in terms of the Dorn-Rajnak theory of the Peierls mechanism when the
deformation is controlled by the rate of nucleation of pairs of kinks.
Above 115°K, an athermal mechanism is operative.

I. INTRODUCTION

This investigation was undertaken for the purpose of elucidating the rate-controlling mechanism for slip in polycrystalline aggregates of b.c.c. magnesium, 14 wt.% Li, 1.0 - 1.5 wt.% Al alloy as part of a more general program of study on the plastic behavior of b.c.c. metals, alloys, and intermetallic compounds. The flow stress of this alloy depends strongly on temperature and strain rate over the low temperature region (below about 115°K). A basic understanding of its mechanical behavior necessitates a thorough investigation of the rate controlling mechanisms of the mobile dislocations. The following thermally-activated dislocation mechanisms have been proposed:

(a) Interaction of dislocations with interstitial impurity atoms or with solute atoms in general,

(b) Overcoming the Peierls-Nabarro stress,

(c) Resistance to the motion of dislocations due to jogs on screw dislocations,

(d) Overcoming interstitial precipitates,

(e) Cross-slip.

It was concluded by Dorn and Rajnak,[1] Conrad,[2] and Christian and Masters,[3] that in b.c.c. metals the overcoming of the Peierls-Nabarro stress which arises from the variations in bond energies of atoms in the dislocation core as it is displaced, is the most probable rate controlling mechanism. Several models of the Peierls mechanism have been formulated. Experimental results obtained from the deformation of Mo, Nb, Ta, AgMg, Ag_2Al and Fe-2 % Mn alloy have been successfully explained by the Dorn-Rajnak model of nucleation of kink pairs.

The purpose of this work was to investigate the mechanical behavior
of polycrystalline aggregates of magnesium, 14 wt.% Li, and 1.0 - 1.5% Al
alloy at low temperatures and to correlate the experimental results with
the Peierls mechanism using the Dorn and Rajnak model.

It will be shown that the strong dependence of temperature and strain
rate can be satisfactorily explained by the rate of nucleation of pairs
of kinks on dislocations involved in the Peierls mechanism for plastic
deformation.

II.   DORN AND RAJNAK'S MODEL OF PEIERLS' MECHANISM

A straight dislocation line has its lowest energy when it lies in a potential valley parallel to lines of closest packing of atoms on the slip plane. When such a straight dislocation line moves from one valley toward the next, the atoms in the vicinity of the core of the dislocation change their positions and bond angles, causing the energy of the dislocation to increase. The core energy of the dislocation is assumed to reach a maximum value midway between the two adjacent valleys. Any additional small displacement will cause the dislocation to fall down the hill into the next valley which is another minimum energy position for the dislocation. The shear stress necessary to promote such forward motion of the dislocation at the absolute zero is known as the Peierls stress τ_p.

A forward motion of such kind can be achieved by the nucleation of a pair of kinks under the influence of an applied stress and a thermal fluctuation. When a stress τ^* less than τ_p is applied to the slip plane in the direction of the Burger's vector, the dislocation will move as shown in Fig. 1, from its original position $A_oB_oC_o$ in the valley to a parallel position ABC part way up the Peierls hill. No further motion will occur at the absolute zero. At higher temperatures, thermal fluctuations cause the dislocation to vibrate about its mean position. When a local thermal fluctuation is sufficiently energetic, a dislocation loop AB'C of a critical size is produced which no longer returns to its original position. For all configurations exceeding the critical one, the two kink segments AB' and B'C will move apart under the action of the applied stress, resulting in a forward motion of the dislocation by a displacement,

a, equal to the periodicity of the rows of closely spaced atoms on the

slip plane.

Dorn and Rajnak applied the suggestion of Friedel that the major

factor involved in kink nucleation is the additional energy due to the

increase in the length of the dislocation line. Γ_y was taken to be the

energy per unit length of a dislocation where the y direction was

defined as shown in Fig. 1. Therefore, the line energy was assumed to

be a periodic function of y with a period a. The minor variations of

the line energy with curvature and proportions of edge and screw dislocations

were neglected. Furthermore, although the exact shape of the Peierls hill

was not known, it was assumed by Dorn and Rajnak to be approximated

by:

$$\Gamma_{(y)} = \frac{\Gamma_c + \Gamma_o}{2} + \frac{\Gamma_c - \Gamma_o}{2} \left(\frac{\alpha}{2} + \cos \frac{2\pi y}{a} - \frac{\alpha}{4} \cos \frac{4\pi y}{a} \right) \qquad (1)$$

where Γ_c and Γ_o are the energies per unit length of a dislocation lying

at the top and bottom of the Peierls hill, respectively, and α is a hill

shape factor that was assumed to vary between -1 and +1.

Under a stress τ^* the stable equilibrium position of an infinitely

long dislocation is $y = y_o$. The difference in energy of a displaced

dislocation line (AB'C, Fig. 1) and that of the corresponding straight

dislocation line lying along $y = y_o$ is given by

$$U_n = \int_{-\infty}^{\infty} \left\{ \Gamma_{(y)} \sqrt{1 + \left(\frac{dy}{dx}\right)^2} - \Gamma_{(y_o)} - \tau^* b (y - y_o) \right\} dx \qquad (2)$$

where the first two terms of the integrand are the line energies of the

dislocation in the two positions, and the third term gives the extra work

done by the applied stress τ^* in displacing the dislocation from y_o to y.

The critical energy for nucleating one pair of kinks was calculated by

using Euler's conditions for minimizing the energy. Upon numerical

integration of the above equations, Dorn and Rajnak were able to obtain a

universal relationship between $U_n/2U_K$ and τ^*/τ_p (U_K is the kink energy),

from which relationship could be derived between the activation volume

and the applied stress, and between the velocity of the dislocation and

the applied stress.

III.   EXPERIMENTAL PROCEDURE

The material used in this investigation consisted of a b.c.c.
magnesium, 14 wt.% Li 1.0 - 1.5 wt.% Al alloy. Tensile specimens 0.2"
in dia having 2" long gauge section were machined from the as received
1 in x 6 in x 12 in sheets, annealed under argon at 300°F ($\approx$150°C)
for 4 hours to remove all the deformation put into the specimen during
the machining and finally etched in dilute hydrochloric acid solution to
remove the thin oxide films.

All tensile tests were performed at strain rates of 3.13×10^{-3} sec^{-1}
and 3.13×10^{-5} sec^{-1} on an Instron testing machine in controlled temperature
baths. The temperature variation was controlled within $\pm$2°K of the
recorded value. For testing temperatures below 77°C, a specially designed
Cryostat was used.

IV.   EXPERIMENTAL RESULTS AND DISCUSSION

The dependence of flow stress on temperature and strain rate:
The applied shear stress that is required to cause plastic flow is
given by:

$$\tau = \tau^* + \tau_A \tag{3}$$

where τ is the applied shear stress, and τ^* is the stress required to aid
the thermal activation of the rate controlling mechanism and therefore
decreases precipitously as T increases. τ_A is the stress necessary to
overcome any athermal barriers and therefore it decreases only modestly
as the temperature increases, usually parallel to the shear modulus of
elasticity.

The primary interests of this work lie in the temperature and strain
rate dependence of the dependence of the thermally activated component of
the stress.   The results are shown in Fig. 2 and Fig. 3.

The increasing flow stress with increasing strain rate as well as
decreasing temperature attests to the fact that the operative deformation
mechanism is thermally activated.   Tests below 20°K were not performed
due to the difficulties in controlling the stability of temperature.
The curves in Figs. 2 and 3 were extrapolated to 0°K. Over the lower
ranges of test temperatures, where the flow stress decreases rapidly from
20°K to 115°K for both strain rates, the thermally activated component
τ^* of the stress was calculated from the relationship:

$$\tau^* = \tau_T - \tau_{235} \frac{G(T)}{G(235)} \tag{4}$$

where τ_T is the total resolved shear stress for flow at temperature $T^\circ K$,
and the term $\tau_{235} \frac{G(T)}{G(235)}$ is the total back stress, corrected for the
change in shear modulus with temperature. The variation of shear modulus
with temperature was calculated from data obtained by J. Trivisonno and
S. Smith, Fig. 4. Values of τ^*, which are now corrected for specimen
variation in τ_A, are shown in Fig. 3 for two strain rates.

For a thermally activated mechanism, the plastic strain rate is
given by (see Appendix 1):

$$\dot{\gamma} = \frac{\rho L a b^2 \nu}{2w^2} e^{-\frac{U_n(\tau^*)}{kT}} \qquad (5)$$

where ρ = density of mobile dislocations

a = the distance between Peierls' valleys

b = Burger's vector

ν = the Debye frequency

L = the mean length swept out by a pair of kinks once nucleation

occurs in that length

w = width of a pair of kinks at the saddle point free energy

configuration

$U_n(\tau^*)$ = Saddle point free energy for nucleation of pairs of kinks

K = Boltzmann constant

T = absolute temperature

When the testing temperature reaches a critical temperature, T_c, for a
given strain rate, the thermally activated component of the stress τ^*
becomes zero and the thermal energy that needs to be supplied to nucleate

a pair of kinks is just $2U_K$, where U_K is the energy of a single kink.
Therefore, at $T = T_c$

$$\dot{\gamma} = \frac{\rho L a b^2 \nu}{2 w_c^2} e^{-\frac{2U_K(T_c)}{kT_c}} \qquad (6)$$

The theory of Dorn and Rajnak predicts a universal relationship between
$U_n/2U_K$, where U_n is the saddle-point free energy for nucleation of a pair
of kinks and U_K is the kink energy, and τ^*/τ_p. But as shown by Dorn and
Rajnak, to a very good approximation $w \simeq w_c$, where w_c is the critical
width of a pair of kinks and therefore, as shown by Eqs. (5) and (6),

$$\frac{U_n(T)}{2U_K(T)} = \frac{U_n(T)}{2U_K(T_c)\frac{G(T)}{G(T_c)}} \simeq \frac{T}{T_c}\frac{G(T_c)}{G(T)} \qquad (7)$$

It is the purpose of this work to correlate the relationship predicted
by the theory with that obtained from the experimental results. The
expected theoretical trends as shown by the solid curves in Fig. 5 are
in excellent agreement with the experimental results at low temperatures
(from 20°K to 115°K for $\dot{\gamma} = 3.13 \times 10^{-5}$ sec^{-1} and from 20°K to 180°K for
$\dot{\gamma}$ 3.13 x 10^{-3} sec^{-1}) and it appears to agree best with the curve representing
$\alpha = \pm 1$. At higher temperatures an athermal mechanism is operative.
Experimental results are presented in Table 1.

Table 1. Strain rate, temperature dependence of the thermally
          activated component of the flow stress.

T°K	$\tau^* \times 10^{-8}$ dynes/cm^2	
	$\dot{\gamma} = 3.13 \times 10^{-5}$ sec^{-1}	$\dot{\gamma} = 3.13 \times 10^{-3}$ sec^{-1}
19		6.71
20	5.97	
45		4.32
50	2.99	
77	1.43	2.55
90	0.78	1.56
113	0.29	1.15
156	0	0.32
196	0	0
235	0	0.05
273	0.09	0.12
300	0	0.21

τ_p extrapolated = 8.40 $\times 10^8$ dynes/cm^2

The critical temperature at $\dot{\gamma}_1 = 3.13 \times 10^{-5}$/sec. $T_{c_1} = 115$°K

The critical temperature at $\dot{\gamma}_2 = 3.13 \times 10^{-3}$/sec. $T_{c_2} = 180$°K

V. ACTIVATION VOLUME

The activation volume is defined by:

$$V = -\left(\frac{\partial U}{\partial \tau^*}\right)_T = KT \left(\frac{\partial \ln\gamma}{\partial \tau^*}\right)_T \simeq KT \left(\frac{\Delta \ln\gamma}{\Delta \tau^*}\right)_T \qquad (8)$$

The activation volume approximates the product of the Burger's vector and the area swept out during the nucleation of the critical loop (AB'C, Fig. 1). The Peierls mechanism has a unique low value of the activation volume (usually ranging from $5b^3$ to $50b^3$). The activation volume remains constant and is unaffected by increasing the strain. These properties of the activation volume are the most reliable verification of the Peierls mechanism. The properties of the activation volume of the other suggested mechanisms are shown in Table 2. The experimental activation volumes are obtained by the effect of small changes in strain rate on the flow stress. A quantity β is defined as:

$$\beta = \left(\frac{\partial \ln\gamma}{\partial \tau^*}\right)_T \qquad (9)$$

βkT is defined as the apparent activation volume:

$$V_a = \beta kT = kT \left(\frac{\partial \ln\gamma}{\partial \tau^*}\right)_T \qquad (10)$$

For the Peierls mechanism (Eq. (5)), this becomes

$$V_a = kT \left(\frac{\partial \ln\rho}{\partial \tau^*}\right)_T - 2kTw \left(\frac{\partial \ln w}{\partial \tau^*}\right) - \frac{\partial Un}{\partial \tau^*} \qquad (11)$$

Table 2. The properties of the activation volume
of other suggested mechanisms.

Mechanism	Properties
Interaction of dislocations with interstitial impurity atoms or with solute atoms in general.[10-13]	1. V_a depends on interstitial content. Higher values of V_a (probably up to $1000b^3$).
Resistance to the motion of dislocations due to jogs on screw dislocations.[14-16]	1. V_a depends on the structure. 2. Athermal at low temperatures thermally activated at relatively high temperature. 3. $V_a = \ell_j b^2$ (ℓ_j = length of jog).
Overcoming interstitial precipitates.[17]	1. $V_a = \ell_i b^2$ (ℓ_i = length of the precipitates). 2. V_a varies with the impurity content.
Cross-slip.[18]	1. High activation volume (of the order of $700b^3 - 800b^3$).

The negative of the last term of Eq. (11) is the theoretical activation volume V^*. The theoretical activation volume V^* can be rewritten as:

$$V^* = -\frac{\partial Un}{\partial \tau^*} = -\frac{2U_K(0)}{\tau_p^o}\frac{\partial\left(\frac{Un}{2U_K}\right)}{\partial\left(\frac{\tau^*}{\tau_p}\right)}$$ (12)

whereas the term containing w in Eq. (11) is always negligibly small, the apparent activation volume can be slightly larger than the theoretical one, v^*, as a result of the possible increase in the dislocation density, ρ, as the stress is increased. Consequently, when the Peierls mechanism is operative, V_a closely follows the trends of v^*. Fig. 6 shows the apparent activation volume as a function of stress, the low values of V_a in terms of b^3 agree very well with the low activation volume as predicted by the Peierls mechanism. Figure 7 also shows the relatively constant value of the apparent activation volume with increasing strain. The theoretical plot for the activation volume as a function of $\frac{\tau^*}{\tau_p}$ for different values of α is shown in Fig. 8. The experimental values of $-\left(\frac{\tau_p}{2U_K}\right)\frac{\partial U}{\partial \tau^*}$ for different values of $\frac{\tau^*}{\tau_p^o}$ (corrected for change in shear modulus with temperature) are also plotted on the same figure, indicating the right order of magnitude of the experimental results.

VI. ACTIVATION ENERGY OF PAIRS OF KINKS

The apparent activation energy for nucleation of pairs of kinks is defined as:

$$q_{\tau^*} = \frac{\partial \ln \dot{\gamma}}{\partial \left(-\frac{1}{kT}\right)_{\tau^*}} \tag{13}$$

The apparent activation energy for nucleation of pairs of kinks at $0°K$ can be determined by the change in strain rate due to the change in temperature at constant stress (τ^*).

$$\dot{\gamma} = \dot{\gamma}_0 \, e^{-\frac{Un}{kT}}$$

where:

$$\dot{\gamma}_0 = \rho ab \, \frac{L}{2w^2} \nu$$

at the critical temperature (where $\tau^* = 0$).

$$\dot{\gamma}_1 = \dot{\gamma}_0 \, e^{-\frac{2U_K(T_{c1})}{kT_{c1}}}$$

$$\dot{\gamma}_2 = \dot{\gamma}_0 \, e^{-\frac{2U_K(T_{c2})}{kT_{c2}}}$$

since $\dot{\gamma}_0$ remains constant for both cases,

$$\frac{\dot{\gamma}_2}{\dot{\gamma}_1} = \frac{\exp - \left[\frac{2U_K(T_{c2})}{kT_{c2}}\right]}{\exp - \left[\frac{2U_K(T_{c1})}{kT_{c1}}\right]} = \frac{\exp - \left[\frac{2U_K(0)}{kT_{c2}} \cdot \frac{G(T_{c2})}{G(0)}\right]}{\exp - \left[\frac{2U_K(0)}{kT_{c1}} \cdot \frac{G(T_{c1})}{G(0)}\right]}$$

or

$$2U_K(0) = \frac{kG(0)}{\dfrac{G(T_{c1})}{T_{c1}} - \dfrac{G(T_{c2})}{T_{c2}}} \ln \frac{\dot{\gamma}_2}{\dot{\gamma}_1} \tag{14}$$

Substituting:

$$k = 1.38 \times 10^{-16} \text{ ergs/}^\circ K$$

$$G(0) = 1.74 \times 10^{10} \text{ dynes/cm}^2$$

$$G(T_{c1}) = 1.63 \times 10^{10} \text{ dynes/cm}^2$$

$$G(T_{c2}) = 1.56 \times 10^{10} \text{ dynes/cm}^2$$

$$T_{c1} = 115^\circ K$$

$$T_{c2} = 180^\circ K$$

$$\ln \frac{\dot{\gamma}_2}{\dot{\gamma}_1} = 4.602$$

$$2U_K(0) = 0.2 \times 10^{-12}$$

$$= 2 \times 10^{-13} \text{ ergs} = 0.13 \text{ eV}$$

This value is of the right order of magnitude for activation energy, when the rate controlling mechanism is by nucleation of pairs of kinks.

VII.   LINE TENSION OF DISLOCATION SITUATED AT THE PEIERLS VALLEY

Dorn and Rajnak gave the theoretical values of $\frac{\tau_p ab}{\pi \Gamma o}$ and $\frac{2U_K \pi}{a \Gamma o}$

for different values of α and R-1, where $R = \frac{\Gamma c}{\Gamma o}$ the ratio of the energy

at the top and at the bottom of a Peierls hill. The theory demands that:

$$\frac{2\pi U_K(0)}{a \Gamma o} = 6.49 \sqrt{R - 1}$$

and

$$\frac{\tau_p ab}{\pi \Gamma o} = 1.3 \, (R - 1)$$

or

$$\frac{2\pi U_K(0)}{a \Gamma o} = 5.75 \left[\frac{\tau_p^* ab}{\pi \Gamma o} \right]^{1/2} \tag{15}$$

for $\alpha = +1$, where

$$a \simeq b = 3.04 \times 10^{-8} \text{ cm}$$

$$2U_K(0) = 0.2 \times 10^{-12} \text{ ergs}$$

τ^* (extrapolated) $= 8.40 \times 10^8$ dynes/cm^2.

Solving for Γo; $\Gamma o = 0.54 \times 10^{-4}$ dynes. The line tension of the dislocation

is related to the shear modulus G and the Burger's vector b by the equation:

$$\Gamma o = \delta G(0) b^2 \tag{16}$$

Hence,

$$\delta \simeq 3$$

This value compared with Nabarro's estimation of the line tension $\frac{Gb^2}{2}$

proves to be of the right order of magnitude.

VIII.   EXPERIMENTAL VALUE OF THE NUMBER OF DISLOCATIONS PER UNIT LENGTH (ρL)

For a given strain rate and temperature, the value of ρL can be determined from the equation.

$$\dot{\gamma} = \frac{\rho L a b^2 \nu}{2w^2} \, e^{-\frac{Un}{kT}} \tag{17}$$

The Debye frequency was estimated to be 5×10^{12} per sec from the Debye characteristic temperature of magnesium-lithium alloy. Estimating w_c to be $60b$ from Fig. 6 and substituting the values of a, b, $2U_K$ and T_c, it was found that $\rho L = 0.3$ per cm. This value of ρL seems to be somewhat low when compared with that of AgMg[9] (227 per cm) and Ta[6] (104 per cm).

There are some possibilities that might account for this low value of the number of mobile dislocations per unit length (ρL). An obvious possibility is that the preexponential expression of Eq. (17) is somewhat - in doubt since w is not well-defined. This arises because the kinks of the critical pair are not abrupt. Another possibility is that the precipitates and impurities could pin-down the dislocation line, modifying the expansion of the kinks by affecting L and w, as qualitatively considered by Kossowsky and Brown. The theory neglects the effects that arise in dislocation segments of finite length which may be restrained at their terminal points on the slip plane. Nevertheless, it is a possible number in terms of the possibility of a low value of L.

IX. CONCLUSIONS

(1) The strong temperature dependence of the flow stress below 115°K can be explained by the Dorn and Rajnak theory of the Peierls mechanism of plastic deformation. Above about 115°K an athermal mechanism is operative.

(2) The shape of the Peierls hills for magnesium, 14 wt.% Li, 1.5 wt.% Al alloy seems to approach the theoretically predicted curve with $\alpha = +1$.

(3) The activation energy of the process of the nucleation of pairs of kinks is estimated to be 0.13 eV., which is of the right order of magnitude for activation energy, when the rate-controlling mechanism is by nucleation of pairs of kinks.

(4) The experimentally deduced values of the apparent activation volume and the line tension are in agreement with the theory.

APPENDIX I

Various authors, Celli et al.,[19] Friedel,[20] Jøssang et al.,[21]
Seeger et al.,[22] have described the formulation of the forward velocity
of dislocations and the strain rate resulting from the nucleation of
pairs of kinks (see Brailsford (1961) regarding the redistribution of
existing kinks along a dislocation due to the action of a stress). Only a
first order approximation was attempted by P. Guyot and J. Dorn (1966).
L (Fig. 1) was assumed to be the average length of a dislocation that
might be swept out by a pair of kinks following their nucleation. L was
also assumed to be much larger than w, the width of the critical pair of
kinks, and end effects were neglected. One possible formulation was
based on the fact that there are L/b (b is the Burger's vector) points
along the length L at which a pair of kinks might be produced and
consequently,

$$\nu_n \simeq \nu_E \frac{L}{2b} e^{-\frac{Un}{kT}} \qquad (A.I.1)$$

where ν_E is the Einstein frequency. This would apply for cases where
the fluctuation might be localized. On the other hand when the thermal
fluctuation is spread over the critical width, w, of the pair of kinks,

$$\nu_n' \simeq \frac{\nu b}{w} \frac{L}{2w} e^{-\frac{Un}{kT}} \qquad (A.I.2)$$

where $\nu b/w$ is the frequency of vibration of the dislocation, ν is the
Debye frequency, and $\frac{L}{2w}$ is approximately the number of wave lengths along
the dislocation line at which nucleation might occur. These expressions

differ somewhat from the original suggestion of Dorn and Rajnak (1964)
(which was based partly on both concepts) that

$$\nu_n \simeq \frac{b}{w} \frac{L}{b} e^{-\frac{Un}{kT}} \qquad (A.I.3)$$

An exact analysis for ν_n is quite complicated. Inasmuch as the vibrations
are coupled, it appears that Eq. (A.I.2) might prove to be the more
satisfactory approximation in most cases. These equations apply only
when the velocity of the kinks is so great relative to their nucleation
rate that not more than one pair of kinks exist in length L at any one
time. Dorn and Rajnak (1964) have also described cases where the kink
velocity might be so slow relative to the nucleation rate that several
pairs of kinks will be moving along a single dislocation segment at one
time. Thus far, however, there have been no experimental confirmations of
this possibility.

The average velocity of a dislocation moving as a result of nucleation
of pairs of kinks is

$$v = \nu_n a = \frac{\nu abL}{2w^2} e^{-\frac{Un}{kT}} \qquad (A.I.4)$$

and this gives a shear strain rate of

$$\dot{\gamma} = \rho bv = \frac{\rho Lab^2}{2w^2} \nu e^{-\frac{Un}{kT}} \qquad (A.I.5)$$

where ρ is the total length of all thermally activatable dislocation
segments per unit volume of the crystal.

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FIGURE CAPTIONS

Figure 1. Schematic diagram showing the nucleation of a pair of kinks.

a is the spacing between parallel rows of closely spaced atoms

of the slip plane.

Figure 2. 0.05% flow stress, τ, vs. temperature.

Figure 3. The thermally activated 0.05% flow stress, τ^*, vs. temperature.

Figure 4. Variation of shear modulus G with temperature.

Figure 5. The thermally activated component of the flow stress vs.
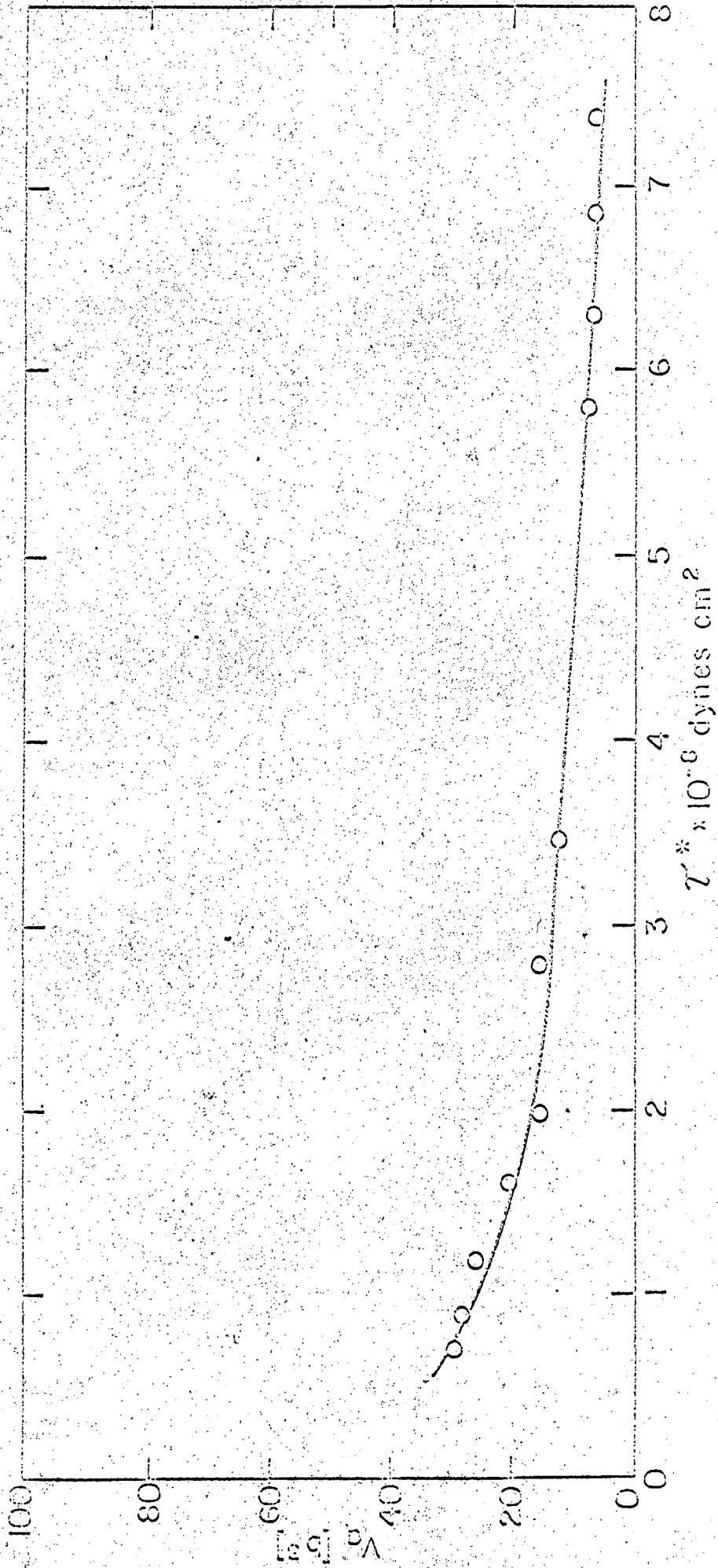
temperature in dimensionless units.

Figure 6. The thermally activated component of the flow stress vs. the

activation volume in units of b^3.

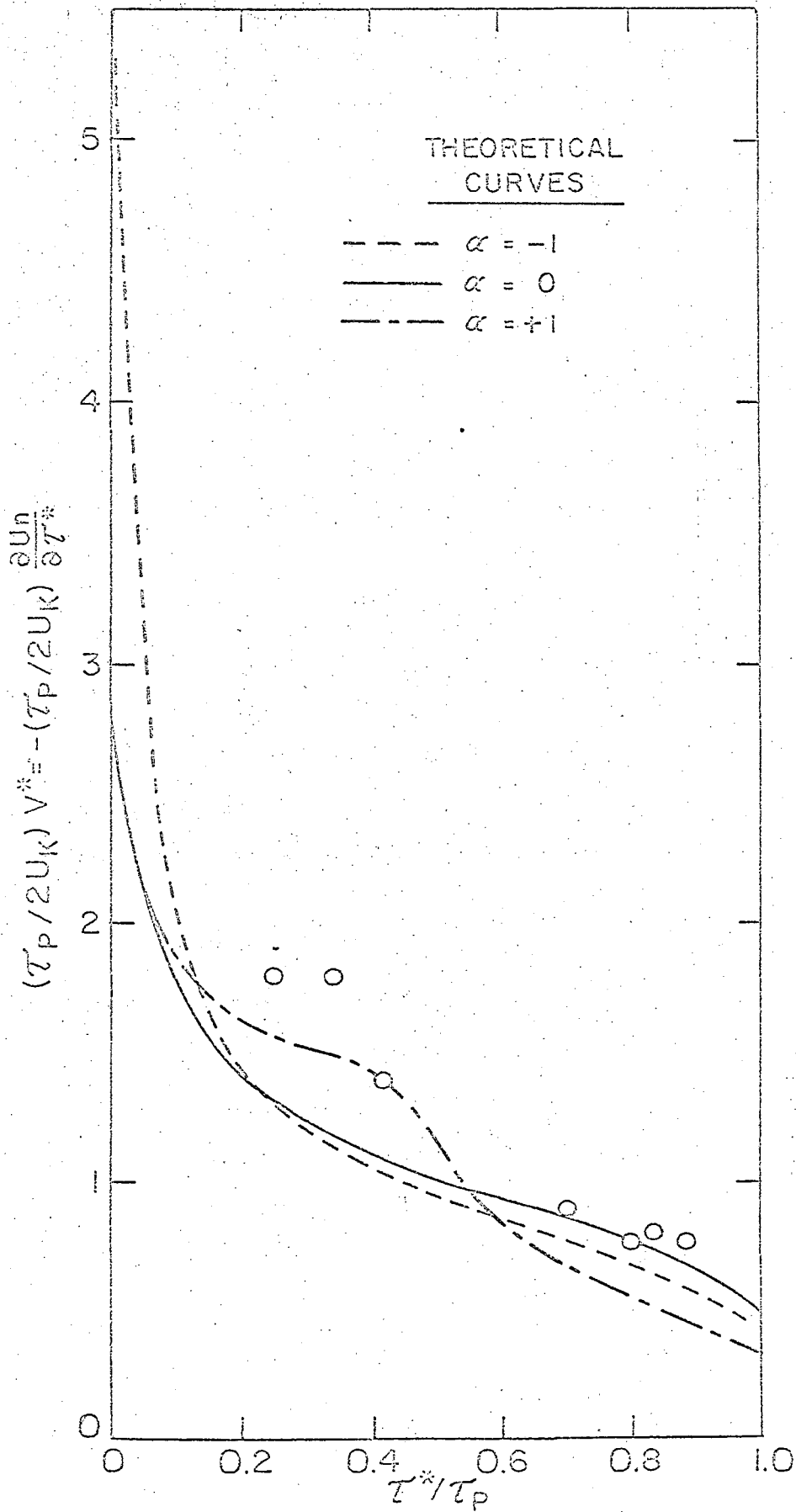Figure 7. Apparent activation volume at 77°K vs. strain.

Figure 8. The thermally activated component of the flow stress vs.

activation volume in dimensionless units.
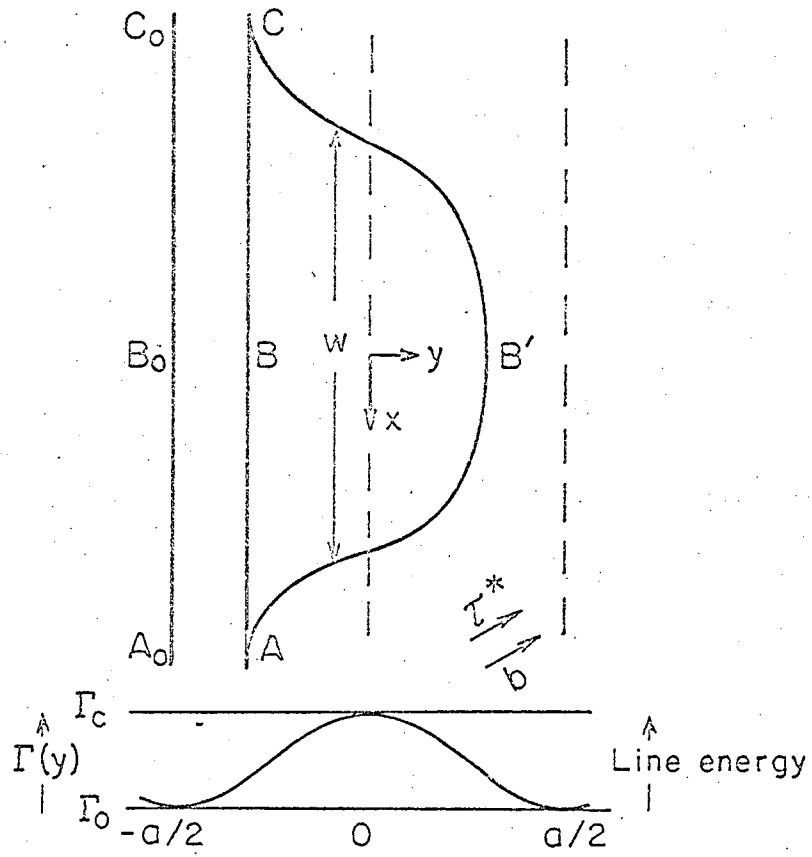
FIG. I NUCLEATION OF A
PAIR OF KINKS.
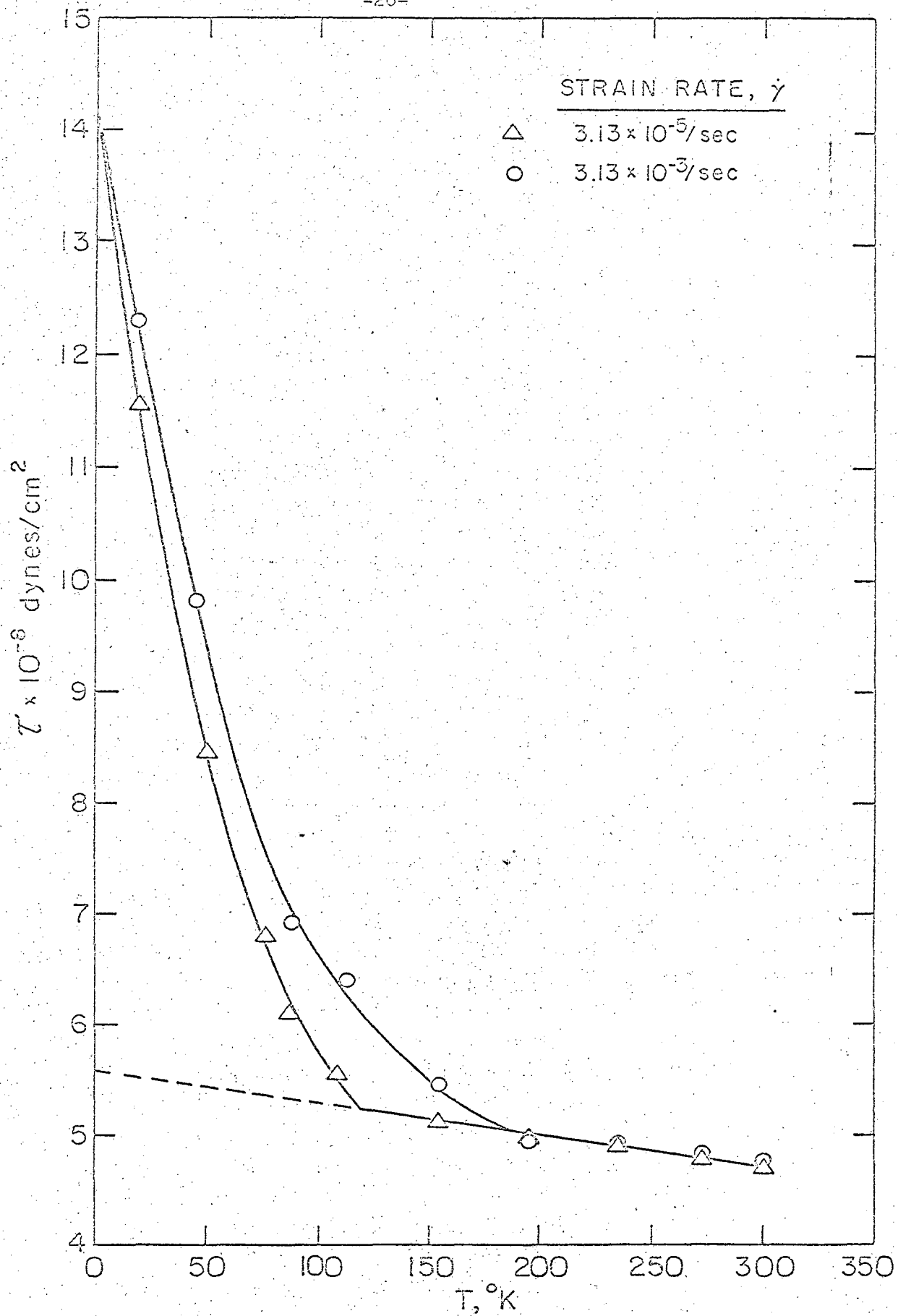
FIG. 2   0.05 % FLOW STRESS vs TEMPERATURE.
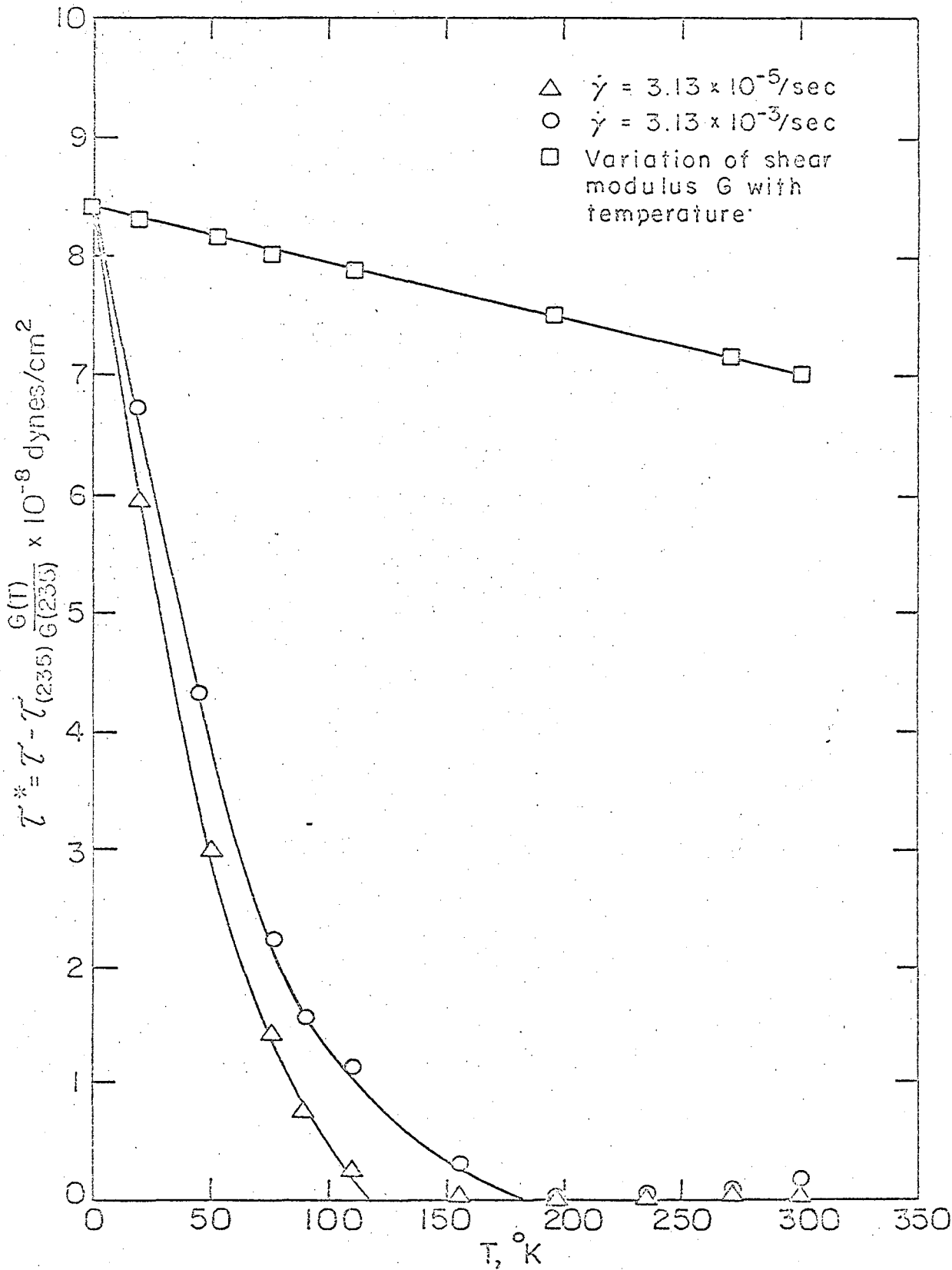
FIG. 3 THERMALLY ACTIVATED 0.05% FLOW STRESS
vs TEMPERATURE.
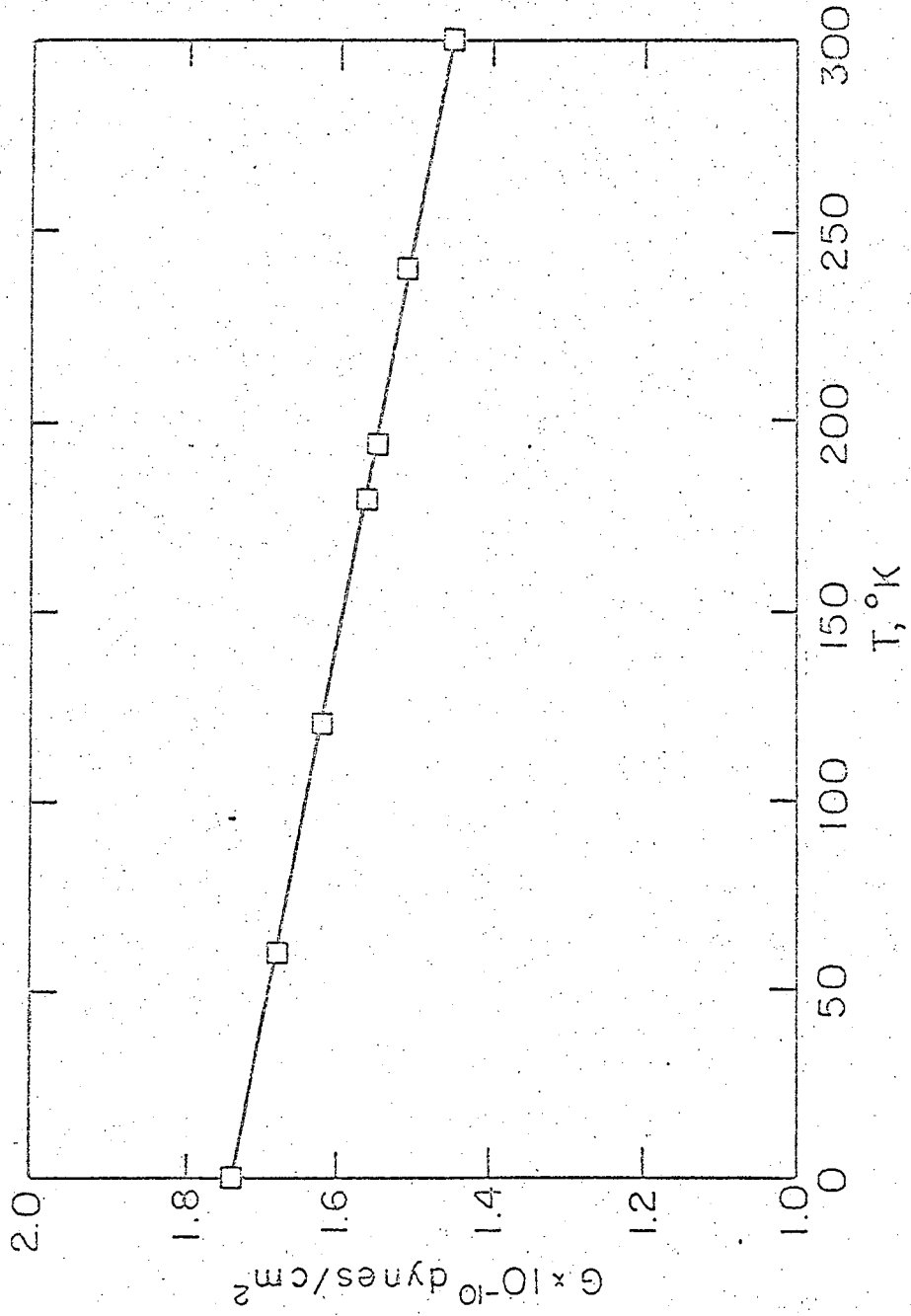
FIG. 4 VARIATION OF SHEAR MODULUS G WITH TEMPERATURE.
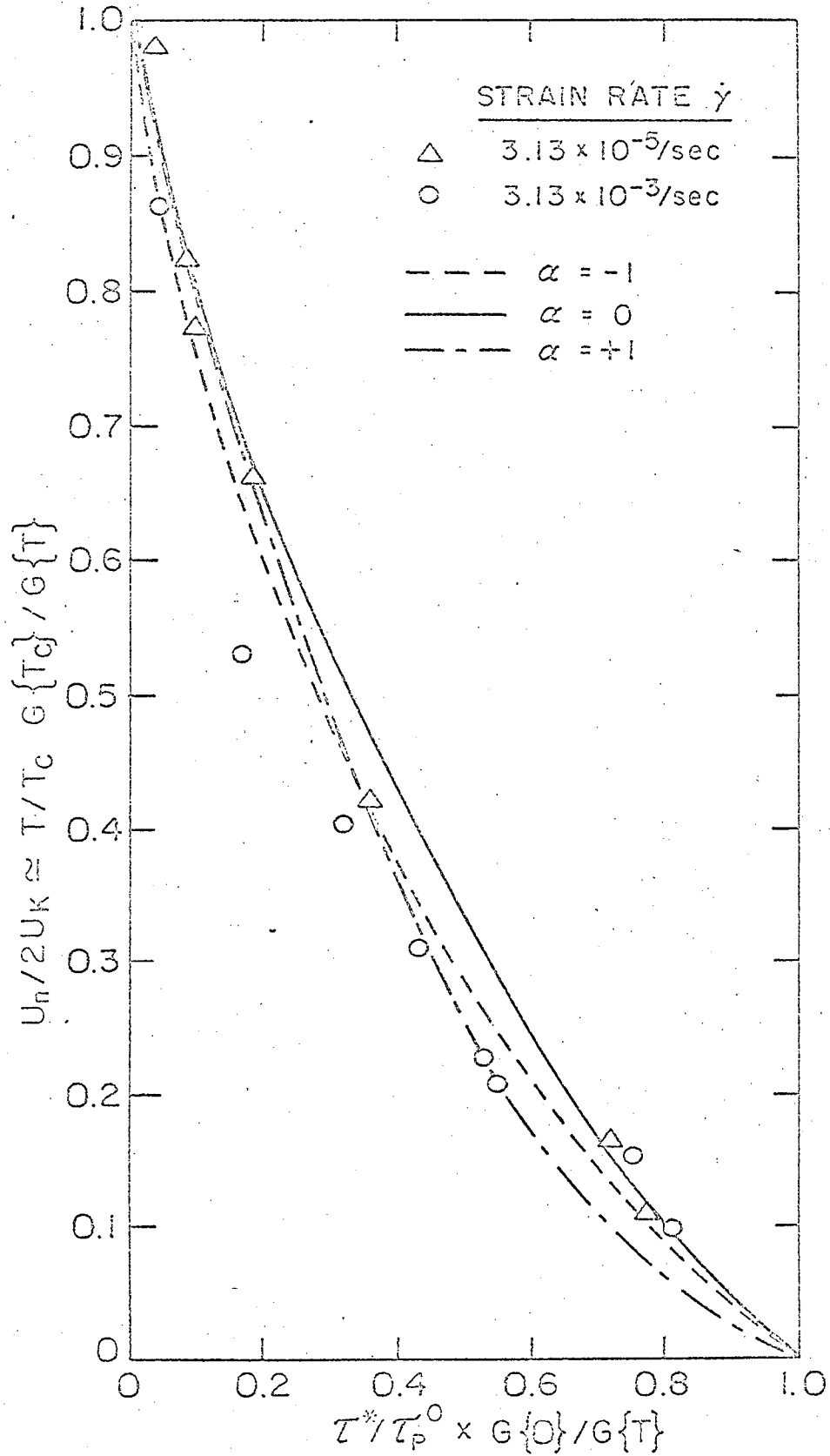
FIG. 5 THE THERMALLY ACTIVATED COMPONENT
OF THE FLOW STRESS vs TEMPERATURE IN
DIMENSIONLESS UNITS.

FIG. 6. THE THERMALLY ACTIVATED COMPONENT OF THE FLOW STRESS
vs THE ACTIVATION VOLUME IN UNITS OF b^3.
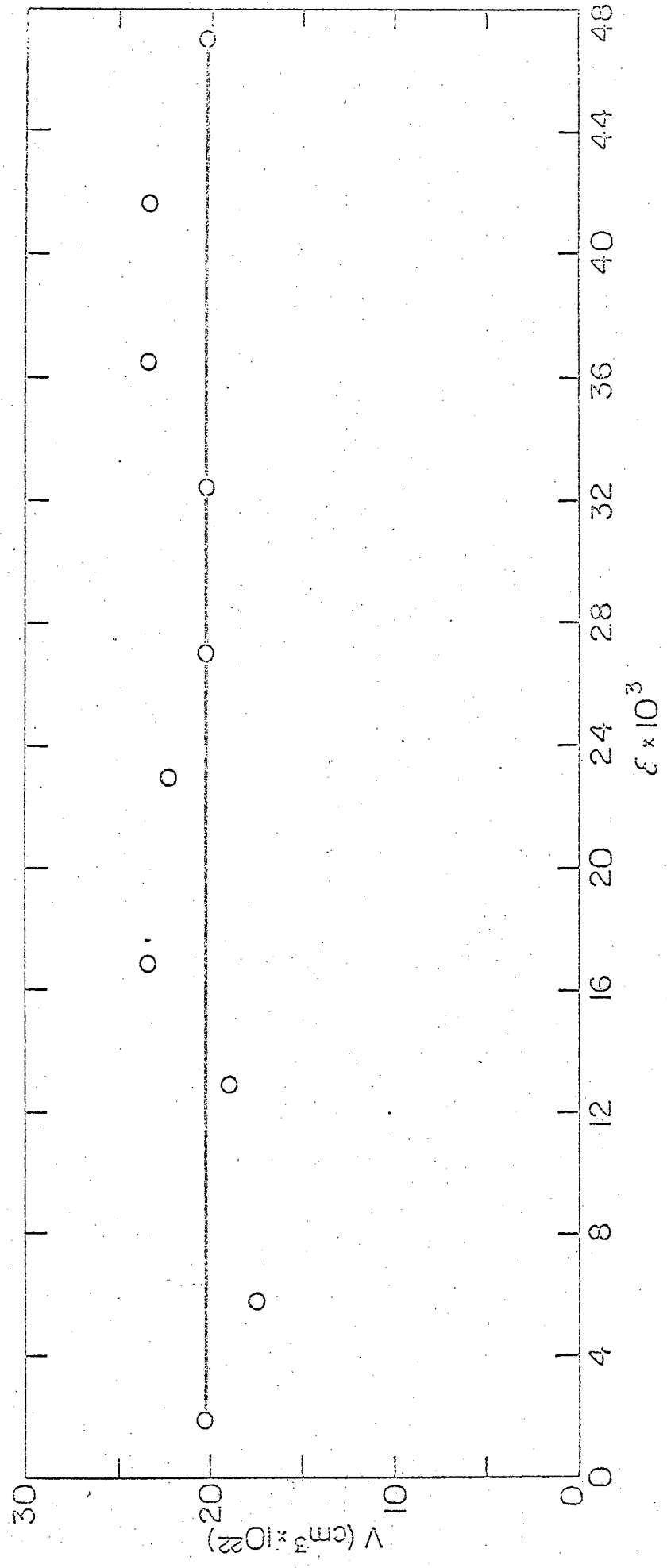
FIG. 7 APPARENT ACTIVATION VOLUME AT 77°K vs STRAIN.

FIG. 8 THE THERMALLY ACTIVATED COMPONENT OF
FLOW STRESS vs ACTIVATION VOLUME IN DIMENSIONLESS
UNITS.