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Inorganic Materials Research Division, Lawrence Radiation Laboratory, and Department of Mineral Technology, College of Engineering, University of California, Berkeley, California

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ABSTRACT

The effect of strain rate and temperature on the tensile flow stress of a polycrystalline b.c.c. alloy of Mg containing 14 wt.% Li and 1.5 wt.% Al was investigated for strain rates of 3.13×10^{-5} to 3.13×10^{-3} per second over the range from 20° to 300°K.

From about 180° to 300°K the alloy exhibited an athermal deformation behavior where the flow stress was independent of strain rate and increased only slightly with decreasing temperature. At lower temperatures the flow stress was strongly strain rate and temperature dependent, characteristic of deformations controlled by thermally activated mechanisms. The activation volume for thermally activated plastic deformation, was between 5 and 30 cubic Burgers vectors, independent of plastic strain. This low temperature thermally activated deformation behavior was found to be in satisfactory agreement with the theoretical dictates of the Dorn-Rajnak[1] formulation of the Peierls mechanism where deformation is controlled by the rate of nucleation of pairs of dislocation kinks over the Peierls energy barriers.

I. INTRODUCTION

Several experimental studies of the low temperature thermally activated deformation of b.c.c. metals and alloys (Mo,[1] Ta,[1] Fe - 2% Mn,[2] Fe - 11% Mo[3] and AgMg[4]) have revealed that the strain rate is controlled by the activation of dislocations over the Peierls-Nabarro energy hills. Although there is some uncertainty as to the nature and effect of solute atom-dislocation interactions during low temperature deformation of b.c.c. metals, it has been concluded by Dorn and Rajnak,[1] Conrad,[5] and Christian and Masters[6] among others that 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 probable mechanism controlling low temperature deformation.

The purpose of this research was to investigate the low-temperature plastic deformation of the b.c.c. alloy Mg - 14 wt.% Li - 1.5 wt.% Al to determine if the behavior of this alkalai metal alloy might be analogous to that for other b.c.c. metals. This alloy was selected because of its availability and its current industrial importance as a lightweight material for aircraft and aerospace applications.

II. EXPERIMENTAL PROCEDURE

Polycrystalline tensile specimens having cylindrical gage sections 2 in. long by 0.2 in. in diameter were machined from as received alloy sheet stock of Mg - 14 wt.% Li - 1.5 wt.% Al. Specimens were annealed in an argon atmosphere at 423°K for 4 hours and maintained in a kerosene bath together with the sheet stock to prevent corrosion. The resulting specimen microstructure consisted of a coarse uniform dispersion of

incoherent precipitate $MgLi_2Al$ particles[7] in a b.c.c. beta phase matrix having an average grain size of 150 microns. Prior to testing the specimens were chemically polished in dilute hydrochloric acid. Comparison of tensile properties and microstructures of specimens cut from center and edge sections of the sheet stock revealed no effects of inhomogeneities in the sheet material.

Tensile tests were performed on an Instron machine at cross head speeds corresponding to tensile strain rates of 1.56×10^{-5} and 1.56×10^{-3} per second. Stresses were determined to $\pm 2 \times 10^6$ dynes/cm^2 and strains to within ± 0.0001. Average values of shear stress τ and shear strain γ reported were taken as on half the tensile stress and 3/2 the tensile strain respectively. Flow stresses were taken at 0.05% strain offset. Test temperatures down to 77°K were obtained by immersing the specimens in constant temperature baths. Lower temperature tests were performed in a liquid helium cryostat to within ±2° of the reported values. Prior to testing at the various temperatures and strain-rates all specimens were prestrained at 235°K to a standard state stress level to obtain a uniform initial state. Additional tests were made to determine the effect of changes in strain rate and strain on the flow stress by rapidly changing the cross-head motion during testing.

Shear moduli of elasticity needed for analyses of the data, were obtained at several temperatures by a common technique of determining the resonant frequencies of vibrations of rectangular test specimens.

III. EXPERIMENTAL RESULTS

Figure 1 shows the experimentally determined flow stress versus temperature for two strain rates. Two distinct regions of behavior are evident. Below about 180°K the strong increase in flow stress with increased strain rate and decreasing temperature indicates that deformation is controlled by a thermally activated dislocation mechanism. At higher temperatures an athermal region is evident where the flow stress is independent of strain rate and only slightly dependent on temperature.

The applied stress τ to cause plastic flow was separated into two components

$$\tau = \tau^* + \tau_A \tag{1}$$

where τ^* is the effective stress required to aid thermal activation of dislocations over barriers and where τ_A is the athermal component. The latter decreases only modestly with increasing temperature and as shown in Fig. 1 it parallels the temperature dependence of the shear modulus of elasticity. The thermally activated component of the flow stress was determined from the relationship

$$\tau^* = \tau_T - \tau_{A(235)} \frac{G_T}{G_{(235)}} \tag{2}$$

where τ_T is the applied stress for plastic flow at temperature T, G_T and $G_{(235)}$ are the shear moduli of elasticity at temperature T and 235°K respectively, and the quantity $\tau_{A(235)}\, G_T/G_{(235)}$ is the athermal stress component at temperature T corrected for the change in shear modulus from the reference state at 235°K. The effective flow stress

as deduced by Eq. (2) is shown in Fig. 2.

The apparent activation volume, defined by

$$V_a = \left(\frac{\partial \ln \dot{\gamma}}{\partial \tau^*}\right)_T \simeq \left(\frac{\ln \dot{\gamma}_2 - \ln \dot{\gamma}_1}{\tau_2^* - \tau_1^*}\right)_T \tag{3}$$

was deduced directly from the experimental data recorded in Fig. 2. As shown in Fig. 3, the activation volume decreases with increasing values of τ^* from about 30 b^3 to 6 b^3, where b is the Burgers vector. Furthermore, as shown by the typical cyclic strain rate test data given in Fig. 4, the activation volume is independent of the strain.

IV. DISCUSSION

The fact that the activation volume is independent of the strain-hardened state disqualifies all such mechanisms, e.g. intersection, in which the density of barriers to thermally activated slip changes with deformation. The relatively small activation volumes, their decrease with increasing values of the effective stress, and the τ^* - T relationship appear to conform with the dictates of the Peierls mechanism. To examine this possibility in detail, the results will be compared with predictions based on the Dorn-Rajnak model for the Peierls mechanism.

As shown by Guyot and Dorn[8] the plastic strain rate for the Peierls mechanism is closely approximated by

$$\dot{\gamma} = \frac{\rho L b^3 \nu \tau_p}{\pi^2 \Gamma} e^{-U_n\{\tau^*\}/kT} \tag{4}$$

where ρ is the length of mobile dislocations per unit volume; L, the dislocation length swept out by a pair of kinks per nucleation; ν,

the Debye frequency; τ_p the Peierls stress; Γ the line energy; kT the Boltzmann constant times the absolute temperature and U_n is the activation energy required to nucleate a pair of kinks. At a critical temperature T_c where $\tau^* = 0$, the activation energy is $2U_k$, i.e. twice the kink energy. At the absolute zero where $U_n = 0$, τ_o^* equals the Peierls stress at the absolute zero. The functional dependence of $U_n/2U_k$ on τ^*/τ_p deduced for sinusoidal-like Peierls hills by Dorn and Rajnak are shown by the curves in Fig. 5. For tests at the same $\dot{\gamma}$ and ρ, then

$$\frac{U_n\{T\}}{kT} = \frac{2U_k\{T_c\}}{kT_c} \text{ or } \frac{U_{no}}{2U_{ko}} = \frac{T}{T_c}\frac{G\{T_c\}}{G\{T\}} \tag{5}$$

Assuming that the Peierls stress, τ_p also varies with temperature as does the shear modulus of elasticity the data given in Fig. 2 were plotted in Fig. 5 to compare with the theory. Although the experimental data fall slightly below the theoretical curves, the agreement nevertheless is deemed to be quite good. The small variation between theory and experiment could easily arise from deviation of the actual Peierls hills from the shapes, analyzed by Dorn and Rajnak.

The kink energy, at the absolute zero, U_{ko}, can be deduced directly from the experimentally determined values of T_{c1} and T_{c2} given in Fig. 2 for the two strain rates, $\dot{\gamma}_1$ and $\dot{\gamma}_2$. It follows directly from Eq. (4) and the fact that $U_n = 2U_k$ at T_c that

$$\frac{\dot{\gamma}_1}{\dot{\gamma}_2} = \frac{e^{-\frac{2U_{ko}}{kT_{c1}}\frac{G\{T_{c1}\}}{G_o}}}{e^{-\frac{2U_{ko}}{kT_{c2}}\frac{G\{T_{c2}\}}{G_o}}} \tag{6}$$

Introducing the experimentally determined values of $T_{c1} = 117°K$ and $T_{c2} = 180°K$ for $\dot{\gamma}_1 = 3.13 \times 10^{-5}$ and $\dot{\gamma}_2 = 3.13 \times 10^{-3}$ per second, and the values for G shown in Fig. 1 gives the not unreasonable value of $2U_{ko}$ of 2×10^{-13} ergs or 0.13 eV. As shown by Mitchell and Dorn,[9] the kink energy and the Peierls stress are related to the line energy of the dislocation by

$$\frac{2\pi U_{ko}}{a\Gamma_o} = 3.89 \left(\frac{\tau p_o ab}{\pi \Gamma_o}\right)^{1/2} \tag{7}$$

where a is the periodicity of the Peierls hills on the slip plane. Using the previously determined values of τ_{po} and U_{ko} suggest that $\Gamma_o \approx 2.65\ G_o b^2$. The fact that this is about 5 times greater than the theoretical estimate by Nabarro, namely $\Gamma_o \approx 0.5\ G_o b^2$, can arise from a number of factors. For example, the theoretical estimate is a gross approximation based on the isotropic theory of elasticity and the Dorn-Rajnak theory[1] is a simple line-energy model for the Peierls mechanism, which contains a number of simplifying assumptions. As reported by Guyot and Dorn,[8] the line energies deduced from the Peierls stress and the kink energy are usually several times greater than those given by the Nabarro estimate. Consequently the correlation obtained here supports the thesis that the Peierls mechanism is operative.

Assuming that ρL of Eq. (4) is independent of τ^* it is seen that the apparent activation volume (vide Eq. (3) and Fig. 3) equals the theoretical activation volume defined by

$$V^* = -\left(\frac{\partial U_n}{\partial \tau^*}\right)_T = -\frac{\partial U_K}{\tau_p}\left(\frac{\partial U_n/2U_K}{\partial \tau^*/\tau_p}\right) \tag{8}$$

The curves of Fig. 6 give the theoretical expectations based on the Dorn-Rajnak model.[1] The datum points were obtained by employing the apparent volume and the experimentally deduced values for τ_p and U_k. This rather good agreement between experiment and theory constitutes strong support for the contention that the Peierls mechanism is operative.

All material sensitive factors of Eq. (4) have been deduced from the experimental data excepting the product ρL. When the Debye frequency is taken to have the reasonable frequency of about 10^{12} per second, ρL has the low but nevertheless acceptable value of about 5 per cm.

In conclusion, the evidence favoring the operation of the Peierls mechanism for the low temperature deformation in the b.c.c. Mg alloy containing 14 wt.% Li and 1.5 wt.% Al is indeed very substantial in every respect. Whether the mechanism is a true Peierls mechanism or a pseudo-Peierls mechanism[10,11] arising from a tendency of the a/2[111] dislocations to dissociate asymmetrically cannot be established from these polycrystalline data which average the asymmetries of slip and yielding on the $\{\bar{1}01\}$ and $\{\bar{2}11\}$ planes.*

V. CONCLUSIONS

1. For strain rates of about 10^{-4} per second the deformation of the polycrystalline alloy of Mg containing 14 wt.% Li and 1.5 wt.% Al is controlled by a thermally activated mechanism up to about 180°K; from about 180°K to the highest temperature investigated, 300°K, an athermal mechanism prevails.

*Note added in final proof: Recently Bernstein and Li[12] have reported that single crystals of the similar alkalai metal K deform by the Peierls mechanism at low temperatures.

2. All features of the thermally activated mechanism are in good agreement with the requirements of a Peierls process. These include the following:

(a) flow stress versus temperature;

(b) effect of strain rate on flow stress;

(c) the small activation volumes, their insensitivity to strain and their rate of decrease with increasing stress;

(d) fair estimate of the line energy of a dislocation from the experimentally determined Peierls stress and kink energy.

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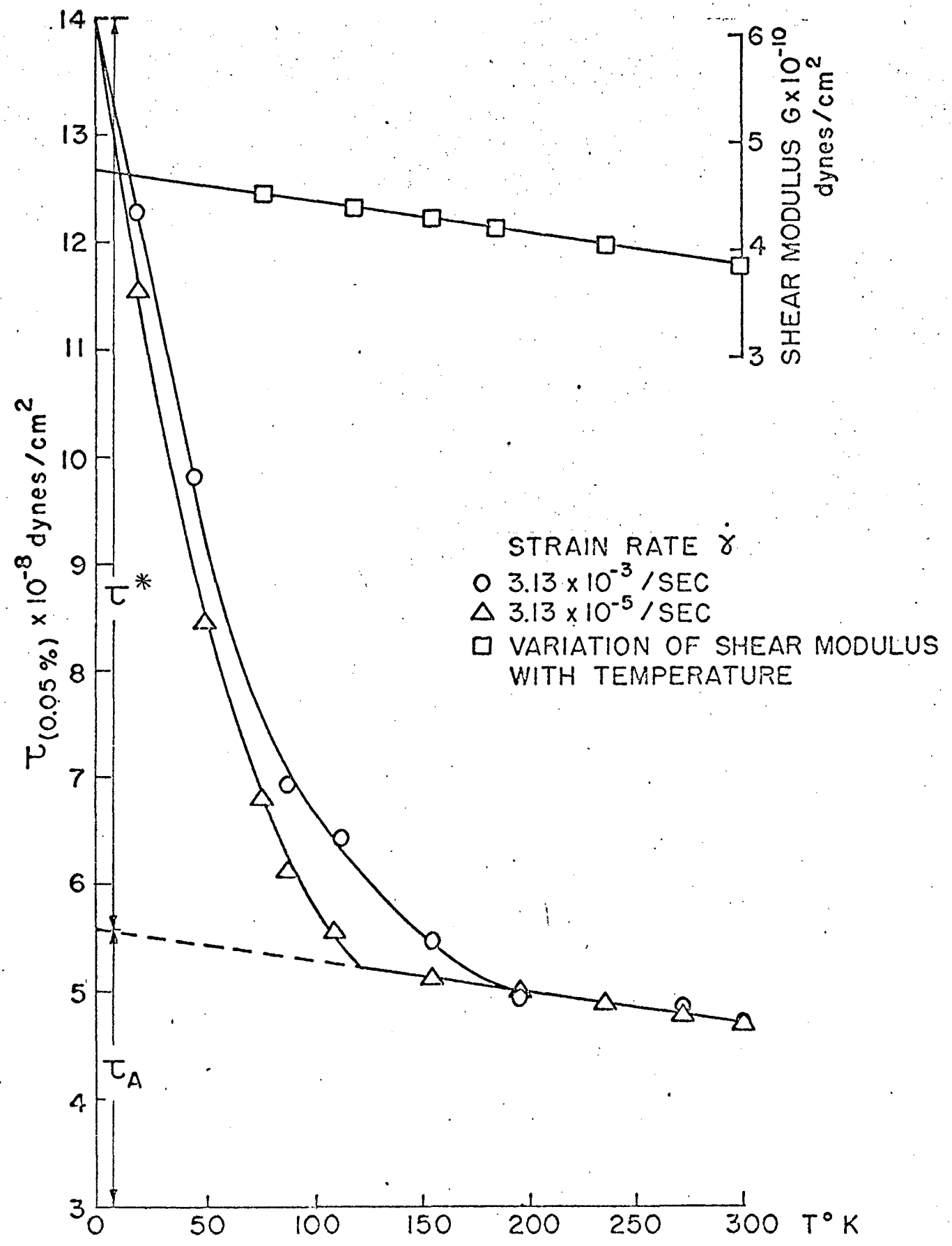

FIG 1 0.05 % FLOW STRESS VS. TEMPERATURE

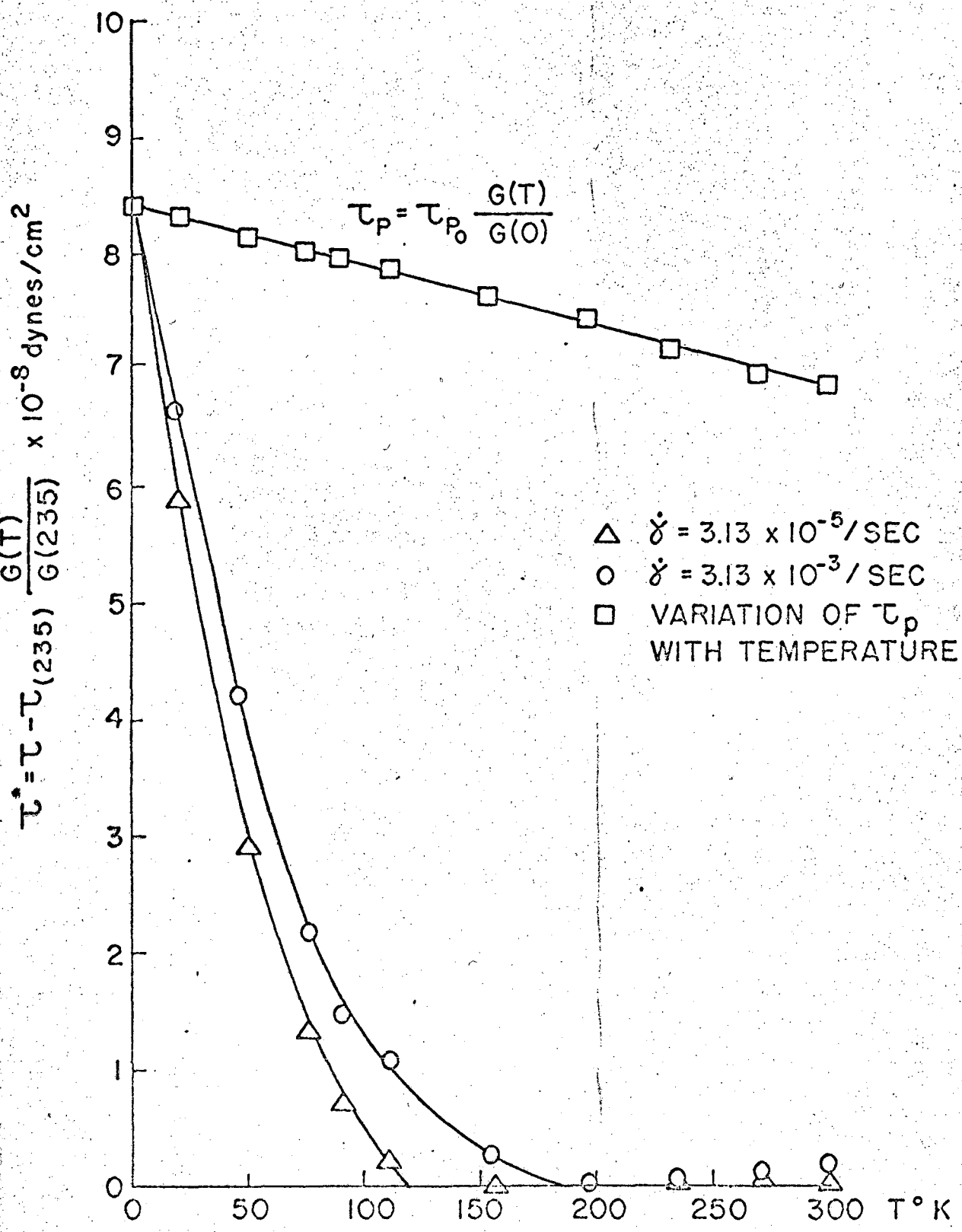

FIG 2 THERMALLY ACTIVATED 0.05 % FLOW STRESS VS. TEMPERATURE

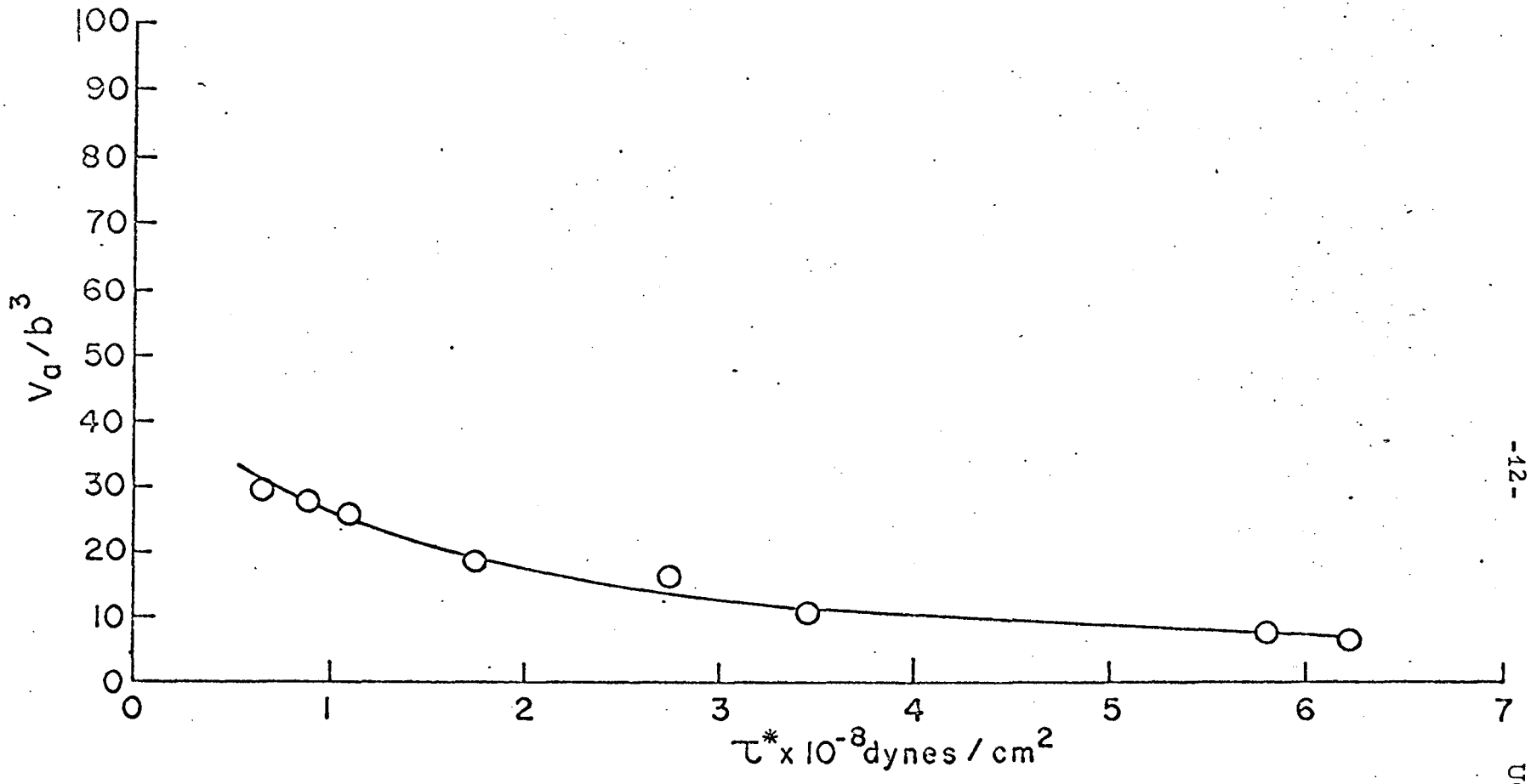

FIG 3 THE THERMALLY ACTIVATED COMPONENT OF THE FLOW STRESS VS. THE ACTIVATION VOLUME IN UNITS OF b^3

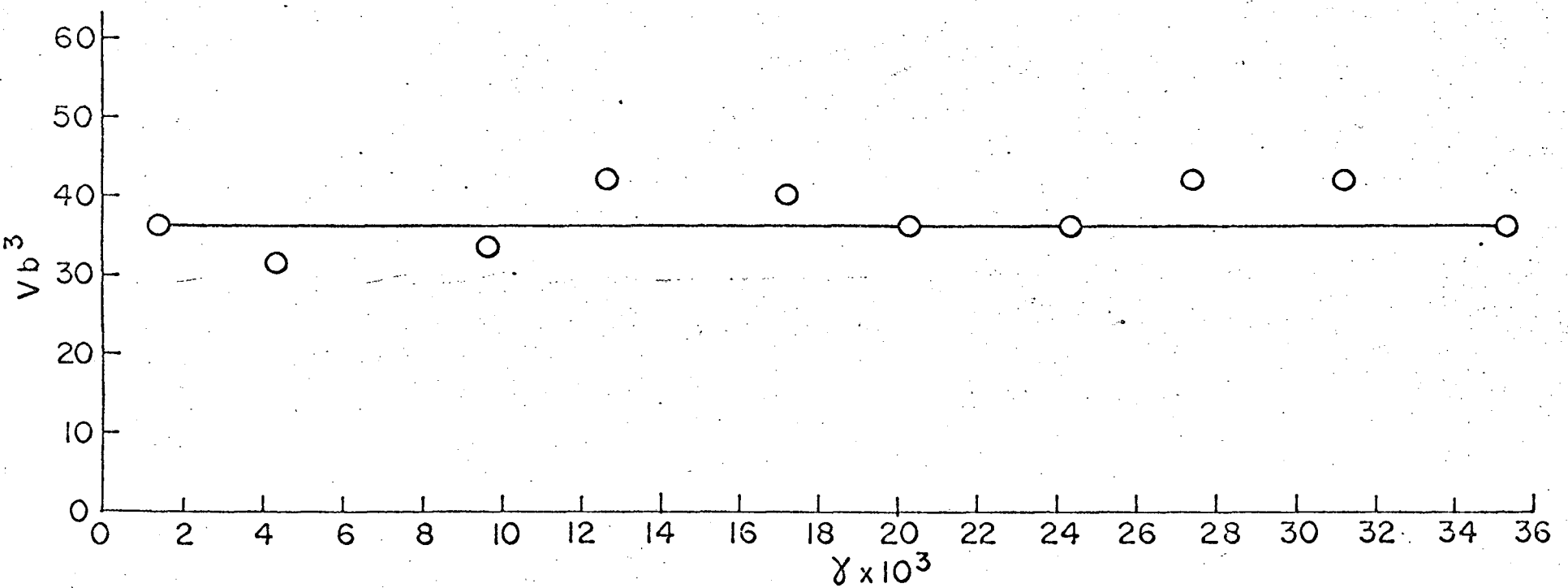

FIG 4 APPARENT ACTIVATION VOLUME AT 77° K VS. SHEAR STRAIN

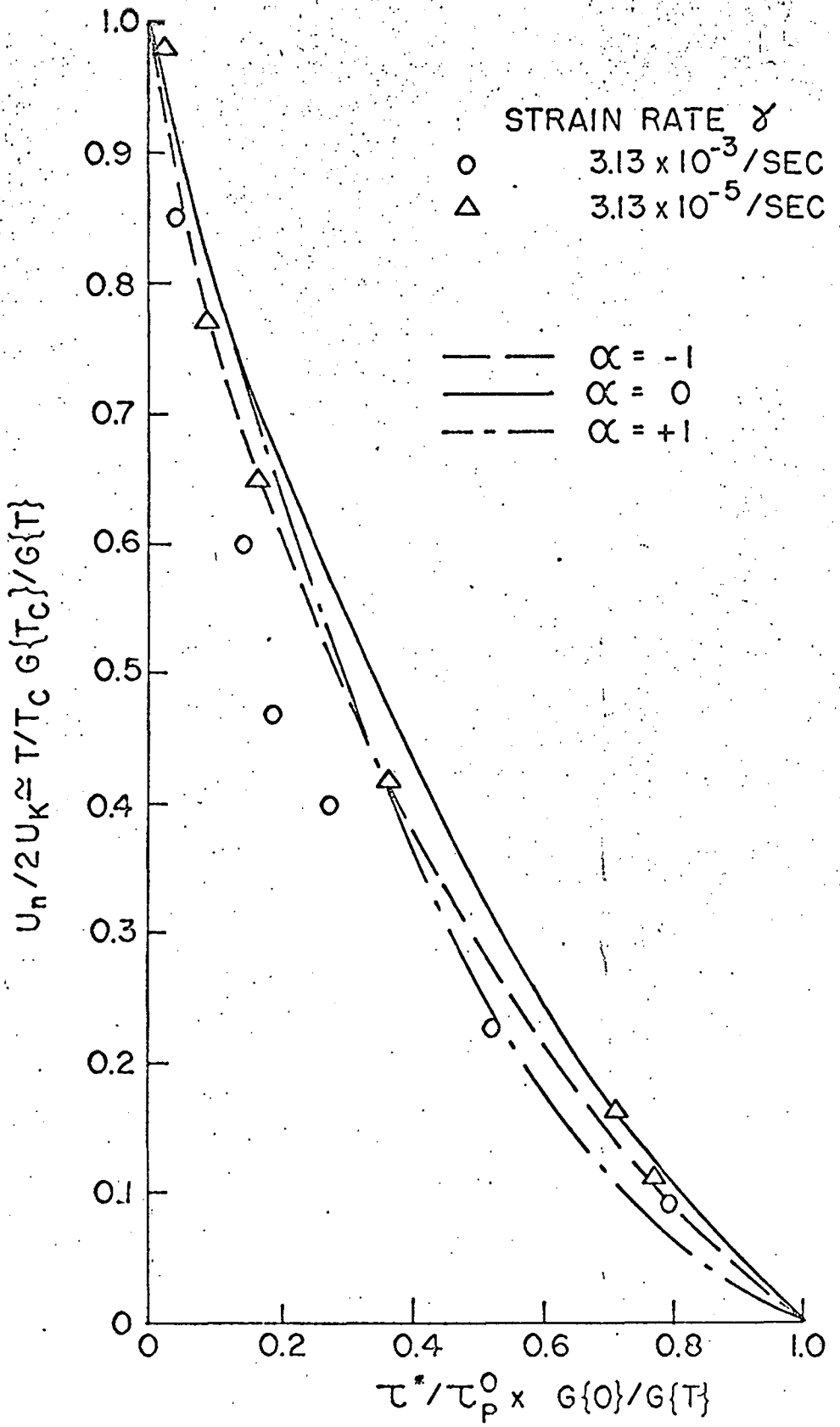

FIG 5 THE THERMALLY ACTIVATED COMPONENT OF THE FLOW STRESS VS TEMPERATURE IN DIMENSIONLESS UNITS.

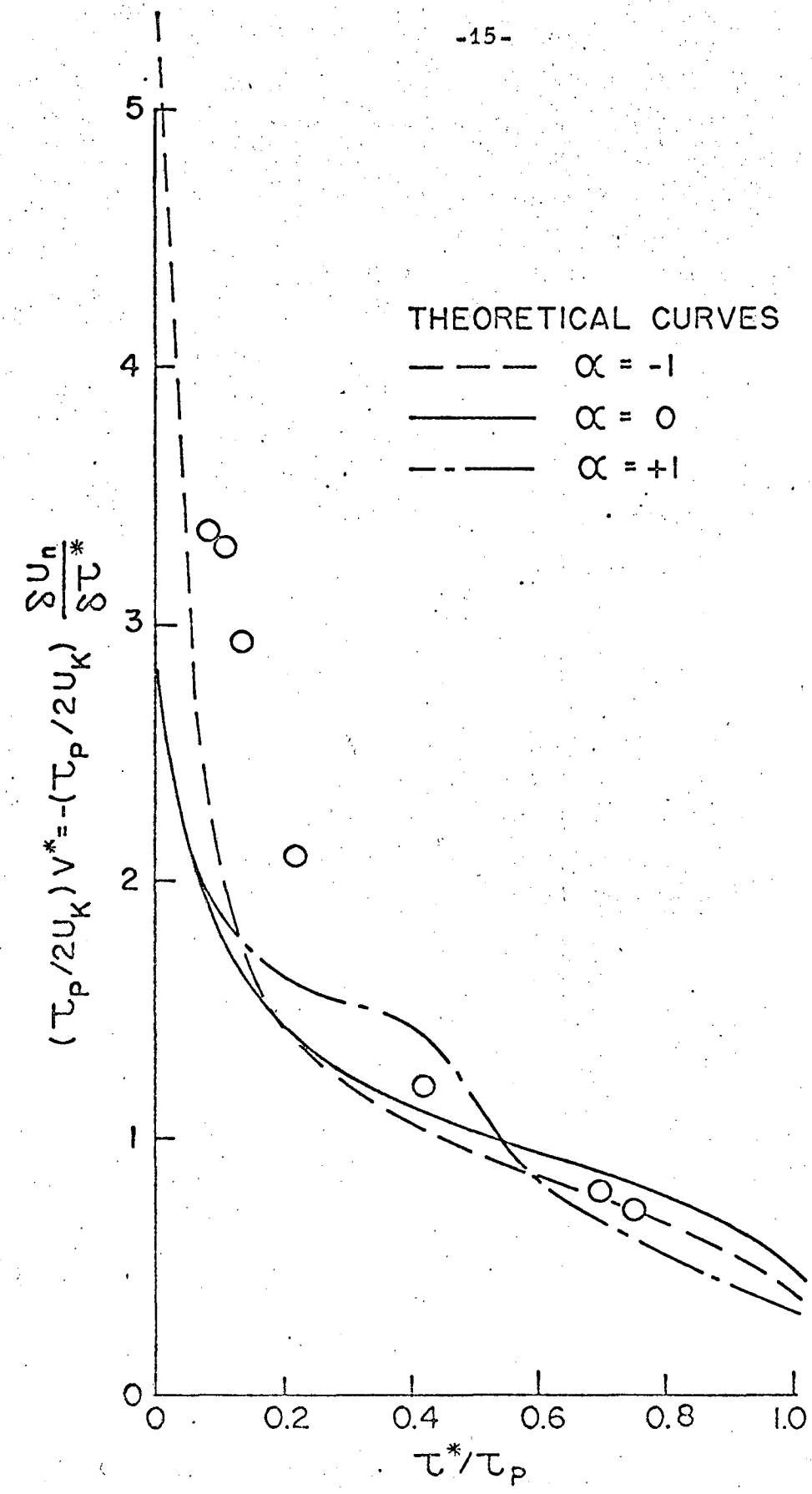

FIG 6 THE THERMALLY ACTIVATED COMPONENT OF FLOW STRESS VS. ACTIVATION VOLUM IN DIMENSIONLESS UNITS

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