THE RATE-CONTROLLING DISLOCATION MECHANISM FOR PLASTIC FLOW IN POLYCRYSTALLINE MAGNESIUM

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ABSTRACT

The deformation behaviour of polycrystalline magnesium is studied with the help of differential-stress creep tests and constant-rate tensile tests. The activation parameters characterizing the rate-controlling barrier to dislocation motion have been calculated. The obstacle energy is found to be 0·5 eV. The results indicate that the same dislocation intersection mechanism is rate-controlling in polycrystals of magnesium as in single crystals.

INTRODUCTION

THE plastic deformation of metals at low temperatures is now known to be facilitated by thermal fluctuations and an understanding of the mechanical behaviour of metals is incomplete without identifying the dislocation mechanisms responsible for the flow stress. Previous investigations[1-5] on f.c.c. metals showed that the thermally activated deformation mechanism is that of intersection of glide and forest dislocations. Comparatively less information is available on the operative dislocation mechanisms in c.p.h. metals. However, intersection mechanism is reported to be rate-controlling for low temperature deformation of magnesium single crystals[6] and polycrystalline cadmium.[7]

Bishop and Hill[8] attempted to predict the plastic behaviour of polycrystalline aggregates from single crystal data by assuming that, among others, the same systems of slip and mechanism of deformation apply to the polycrystal as to the single crystal. Mitra and Dorn[3] concluded from studies on aluminium that the same dislocation mechanism is operative in both single and the polycrystals. This is probably true for the case of f.c.c. metals. Extension of this to c.p.h. metals may not seem permissible because of the limited number of slip systems. In a polycrystalline material, at least five independent deformation systems are required in each grain. Inasmuch as there is only one basal plane per grain in hexagonal metals, the other systems are supplied by glide on prismatic or pyramidal planes or by twinning. The present investigation was initiated to ascertain whether or not the same deformation mechanism as is rate-controlling in magnesium single crystals applies to polycrystalline magnesium. With this end in view, the deformation behaviour of magnesium polycrystals is studied employing differential-stress creep experiments at 87° K. and 197° K. and constant-rate tensile tests in the temperature range 87°–700° K.

EXPERIMENTAL

Polycrystal samples of magnesium (Zn < 0·01%, Al 0·007%, Si 0·005%, Pb + Cu + Ca + Sn + Ni + Fe + Mn < 0·006%, rest Mg) were obtained for creep and tensile tests from rolled sheet with their axes in the rolling direction. All specimens were annealed at 673° K. for 45 minutes. Creep tests were performed on samples with average grain diameters of 0·03 mm. and 0·003 mm., obtained by controlling the mechanical treatment. Specimens employed for tensile tests had an average grain diameter of 0·003 mm. Each specimen was etched before testing.

Creep testing was done with a constant-stress tensile creep unit. The creep strain was measured using an LVDT transducer, the amplified output of which is fed to a strip-chart millivolt recorder. The accuracy of the strain measurement was 10^{-5}. Tests were conducted at 87° K. and 197° K. by surrounding the sample, fixed in a cage, with liquid oxygen and solid CO_2-alcohol mixture respectively. The specimen was initially loaded in steps of about 80 kg./cm.^{2} until creep was observed on the recorder chart. Creep curves were then recorded at stress intervals of about 80 kg./cm.^{2} At each stress level, a small load (~ 8 kg./cm.^{2}) was added to record the incremental creep curve. After a certain time, the incremental load was removed to get the decremental creep curve. This procedure was repeated until the creep rate became very low. Then a major load was added and the process of stress addition and removal repeated.

Tensile tests were performed in a modified Hounsfield tensometer on 2·5 cm. gauge length specimens. Three cross-head speeds corresponding to tensile strain rates of $7 \cdot 8 \times 10^{-4}$, $1 \cdot 8 \times 10^{-4}$ and $2 \cdot 1 \times 10^{-5}$ sec^{-1} were employed in the temperature range 87° to 700° K. For tests above room temperature, thermostatically controlled (± 2° C.) silicone oil bath was used. The load on the sample was recorded by measuring the deflection of the spring beam with the help of an LVDT transducer. The stress values reported are accurate to within 0·1 kg./mm.^{2}

Results

If a single thermally activated mechanism is rate controlling, the shear strain rate, $\dot{\gamma}$, is given by :

$$\dot{\gamma} = NAb\,\nu \exp.\left(\frac{-U}{kT}\right) \qquad (1)$$

where N = the number of points of contact per unit volume between glide dislocations and obstacles, A = the area slipped per successful activation event b = the Burgers vector, ν = the frequency of vibration of the dislocation segment involved in thermal activation, U = the energy to be supplied through thermal fluctuations for the dislocation to overcome the rate-controlling obstacle, k = the Boltzmann's constant and T = the absolute temperature. It is customary to accept Seeger's approximation[9] that the thermal activation energy varies with the thermal component of the flow stress, τ^*, acting on a dislocation according as :

$$U = U_0 - v\,\tau^* \qquad (2)$$

where U_0 is the total energy required for the activation process and 'v' is commonly referred to as the activation volume. Using Equations (1) and (2), the following expressions, which assist in the identification of the deformation mechanism, can be obtained :

$$v = kT\left[\frac{\delta\, ln\, \dot{\gamma}}{\delta \tau^*}\right]_T \simeq kT\left[\frac{\Delta\, ln\, \dot{\gamma}}{\Delta \tau^*}\right]_T = BkT \qquad (3)$$

$$U = -k\left[\frac{\delta\, ln\, \dot{\gamma}}{\delta\,(1/T)}\right]_{\tau^*} \simeq -k\left[\frac{\Delta\, ln\, \dot{\gamma}}{\Delta\,(1/T)}\right]_{\tau^*} \qquad (4)$$

and

$$U = kT\, ln\left(NAb\,\frac{\nu}{\dot{\gamma}}\right). \qquad (5)$$

The results of the tensile tests on magnesium polycrystals are summarised in Fig. 1, in which is shown the variation of tensile flow stress, σ, with temperature for the three strain rates employed. The flow stress values for all specimens are read at 0·1% strain offset from the modulus line. The stresses are corrected for the temperature variation of shear modulus, taking the shear modulus values for magnesium from Slutsky and Garland.[10]

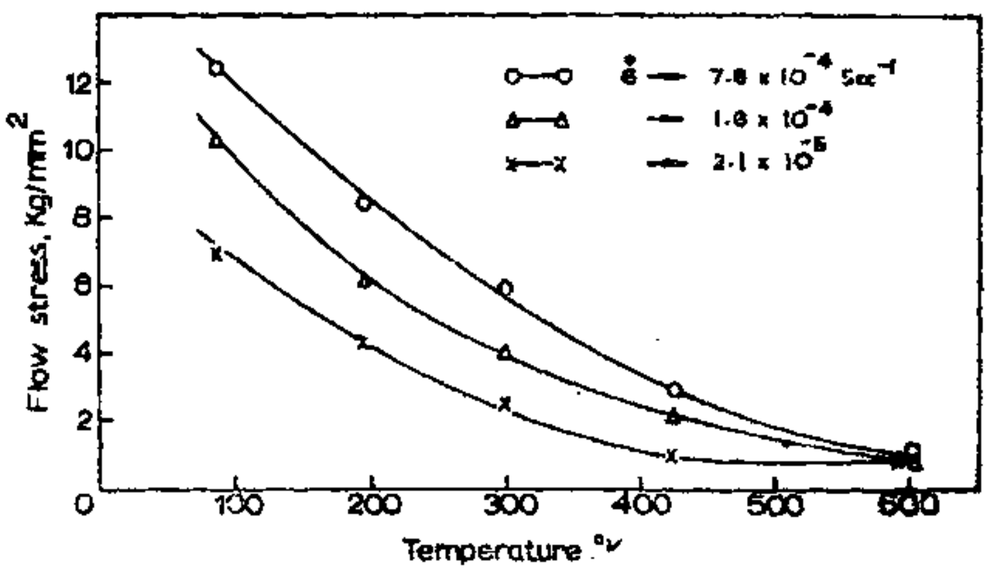

FIG. 1. Effect of temperature and strain rate on the flow stress of magnesium.

Results of the creep tests are utilised to calculate the flow parameter, B. Logarithmic creep is observed over the entire stress range (8–22 kg./mm.^{2} at 87° K. and 4–20 kg./mm.^{2} at 197° K.) studied. A typical plot of the creep rate, $\dot{\epsilon}$, obtained by graphical differentiation of the creep strain, ϵ, *vs.* log (time) plot, is shown in Fig. 2 as a function of strain. B is calculated from the change in creep rate at the instant the increment or decrement is made. The shear stress, τ, and the shear strain, γ, are calculated from the tensile counterparts by assuming that $\tau = \frac{1}{2}\sigma$ and $\gamma = 2\epsilon$. The dependence of activation volume obtained from Equation (3), on the effective stress, τ^*, is presented in Fig. 3. Here, the effective stress

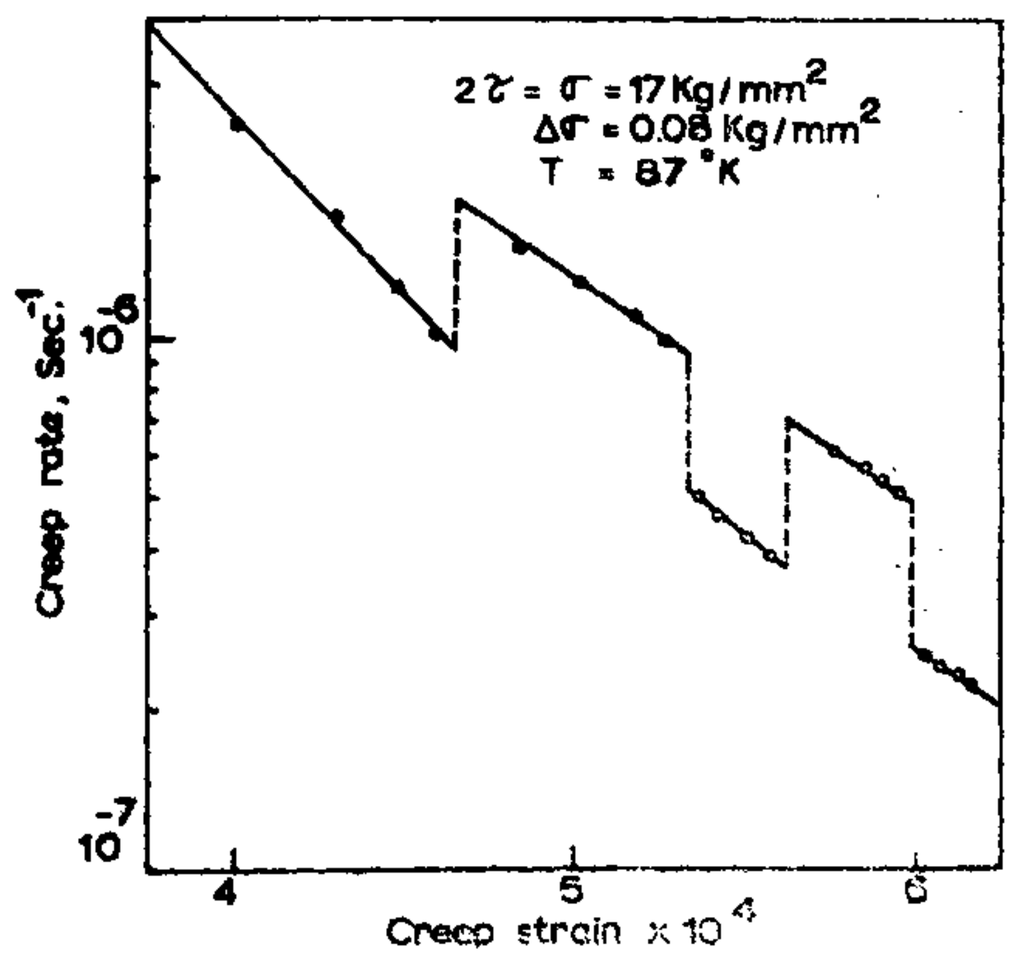

FIG. 2. Effect of stress increments and decrements on creep rate.

is taken to be the difference between the applied stress and the temperature-independent part of the flow stress. The activation volume decreases with increasing stress reaching a near constant value of 50 b^3 at 87° K. and 100 b^3 at 197° K. The Burgers vector for undissociated basal dislocations is used in this calculation. Furthermore, it can be seen from the data at 87° K. that fine-grained magnesium has a smaller activation volume at a given effective stress and the difference becomes insignificant at high stresses.

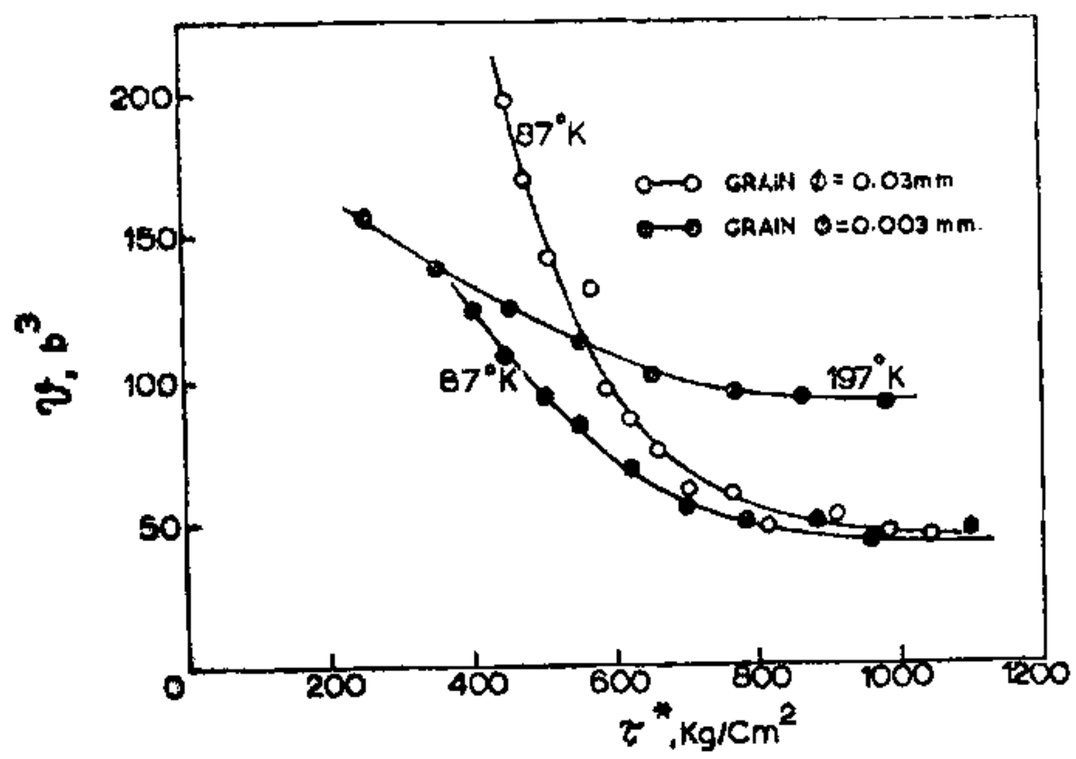

FIG. 3. Activation volume as a function of effective stress.

Discussion

Conrad *et al.*[6] have demonstrated that the rate-controlling mechanism for plastic flow of magnesium single crystals at low temperatures is the intersection of dislocations. In polycrystalline aggregates, the number of operative slip systems and arrangement of dislocations are different from those in single crystals. Inasmuch as the stress to induce cross slip in single crystal is much lower than the applied stress, some cross slip may take place in favourably oriented grains. Cross slip can be thermally activated and hence the rate-controlling mechanism for the deformation of a magnesium polycrystal as a whole may be that of cross slip or mutual intersection. However, the possibility that a third dislocation mechanism could be operative should also be considered.

The rate-controlling dislocation mechanisms suggested for the plastic flow of c.p.h. metals at low temperature are: (*i*) intersection of dislocations, (*ii*) overcoming the Peierls-Nabarro (P-N) stress, (*iii*) cross slip and (*iv*) overcoming the impurity atoms. The mechanism of glide against P-N stress and cross slip require that the activation volume be small, about 10 b^3, and it be independent of temperature at a given effective stress for the P-N mechanism.[11] A value of 50 b^3 at 87° K. and 100 b^3 at 197° K. is obtained in the present investigation even at high effective stresses (Fig. 3). Furthermore, the dependency of activation volume on grain size disqualifies the P-N stress as well as the impurity atoms to be obstacles to dislocation motion in polycrystalline magnesium. In these two cases, the average distance between the obstacles, on which depends the activation volume, should not change with grain size. On the other hand, the measured activation volume is of the magnitude to be expected for the thermally activated intersection mechanism.[12]

The thermal activation energy, U, is calculated, in view of Equation (4) as function of stress from the slope of $ln\ \gamma$ *vs.* 1/T plots for a given stress from the data of Fig. 1. The assumption underlying this procedure, *viz.*, the structure remains constant for the different specimens used, is not unreasonable since the stress values are measured at constant strain and U is calculated at a given stress. Figure 4 indicates that U is sensitive function of stress as required for the intersection mechanism. By obtaining the corresponding temperature for the stress in Fig. 4, the variation of U

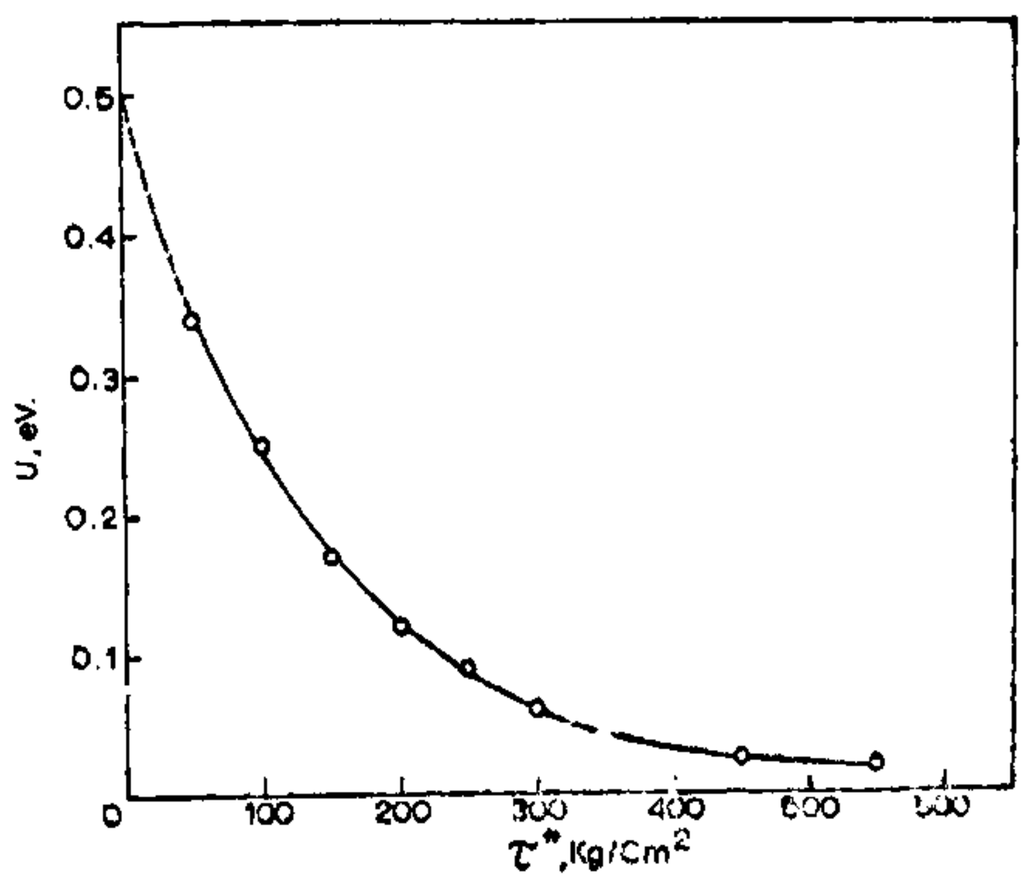

FIG. 4. Activation energy as a function of effective stress.

with temperature is found to be linear (Fig. 5) as dictated by Equation (5). Figure 5 also suggests that the pre-exponential factor ($P = NAb\nu$) is fairly independent of temperature and hence of stress. This again seems to favour the intersection mechanism over others.

The dependency of activation volume on grain size gives further support to the intersection mechanism. For intersection, the acti-

vation volume is given by : $v = bld$, where 'l' is the length of the dislocation segment involved in thermal activation and 'd' is the activation distance. The increase in activation volume with increasing grain size is then a reflection of the increase in the mean free slip length, l. This is plausible inasmuch as there is a tendency in c.p.h. crystals for increasing non-basal slip with decreasing grain size. In fact, Risebrough and Teghtsoonian[13] reported decreasing amount of $\{11\bar{2}2\} < 11\bar{2}3 >$ slip in cadmium as the grain size increases. The grain size effect is not observed beyond a stress of ~ 900 kg./cm.^{2} since at such stresses a given slip system can be expected to operate with equal ease in fine-grained as well as coarse-grained specimen.

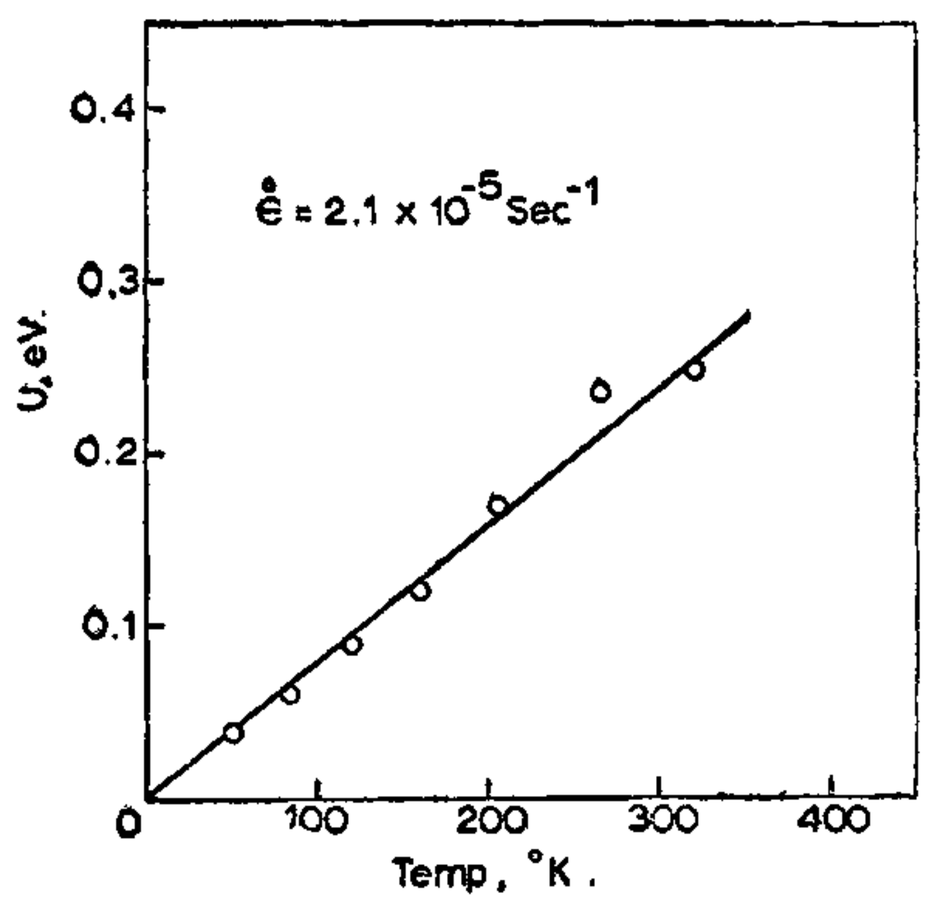

FIG. 5. Activation energy as a function of temperature.

The total intersection energy, U_0, is obtained as 0·5 eV by extrapolation of the $U - \tau^*$ curve (Fig. 4) to $\tau^* = 0$. This is only slightly higher than the theoretical estimate[14] of the jog energy in magnesium and points to a small separation of partial dislocations on basal plane. Dorn and Mitchell[15] have deduced the stacking fault width in magnesium to be ~ 2 b. An estimate of forest spacing is possible since 'l' $= v_0/2\,b^2$, where v_0 is the activation volume at zero effective stress. This yields a forest spacing of about 7×10^{-6} cm. and a dislocation density of nearly 2×10^{10} cm.$^{-2}$ which is of reasonable magnitude. Knowing 'l', the frequency factor, P, to be expected for intersection mechanism can be calculated. Taking $N = l^{-3}$, $A = l^2$ and $\nu = b\nu_0/2\,l$ where ν_0 is the Debye frequency, P calculates to be 2×10^8 sec.$^{-1}$ which is not far from the value of 10^9 sec.$^{-1}$ obtained from the slope of Fig. 5.

The results of the current research thus show that plastic flow at low temperatures is controlled by the same thermally activated deformation mechanism in magnesium single and polycrystals. The obvious inference is that some cross slip may occur in polycrystalline magnesium but subsequent to cross slip, the dislocations move at a rate dictated by intersection.

ACKNOWLEDGEMENTS

The authors are thankful to Dr. S. Dhawan, Director, Indian Institute of Science and Prof. A. A. Krishnan for their interest in this work and to other members of the Materials Research Group for their co-operation.

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