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Berkeley, California

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ABSTRACT

A detailed investigation of the dislocation mechanisms controlling prismatic slip in alpha solid solutions of Mg containing up to 15.9 at.% Li revealed that low temperature prismatic slip is controlled by the Peierls mechanism for all alloy compositions. At higher temperatures slip in the 7.9 at.% Li alloy was found to follow the dictates of the cross-slip mechanism from basal to prism planes, while the higher Li alloys exhibited an athermal deformation process which might be due to short-range ordering.

The major effect of Li additions on the low temperature shear stress for prismatic slip appears to be due to its effect in lowering the Peierls stress whereas an increase in the shear stress with increased Li content at higher temperatures might be due to short-range order strengthening. Additions up to 7.9 at.% Li appear to have only minor effects on the stacking fault energy.

I. INTRODUCTION

Coarse-grained polycrystalline aggregates of distilled Mg exhibit very limited ductility at low temperatures.[1] The ductility of polycrystalline Mg, however, is much improved by solid solution additions in excess of about 8 atomic percent Li.[2] Although the relative amounts of basal slip and twinning are not materially altered by such alloying, a substantial change takes place in the relative amounts of prismatic slip. Whereas prismatic slip is negligible and limited to regions of high stress concentrations in the vicinity of the grain boundaries in pure Mg, extensive prismatic slip over entire grains occurs in the higher Li content alloys. Furthermore as the Li content is increased above 8 atomic percent, the flow stress for polycrystalline aggregates decreases.

Comparison of the single crystal data on pure Mg for basal slip by Sheely and Nash[3] and those for prismatic slip by Flynn, Mote and Dorn[4] indicates that the critical resolved shear stress for prismatic slip is much above that for basal slip. Moreover, the data of Quimby, Mote and Dorn[5] demonstrate that the critical resolved shear stress for basal slip increases with increasing lithium content. These observations suggest that the increased ductility and decreased flow stress of polycrystalline alloys containing more than 8 at.% Li results from a substantial decrease in the critical resolved shear stress for prismatic slip as the Li content is increased.

This investigation was initiated to study the effect of Li addition on prismatic slip in single crystals of Mg in an attempt to identify the cause for the decrease in the critical resolved shear stress with increasing

Li content. Toward this end tests were conducted to uncover the dislocation mechanisms that might be responsible for prismatic slip in the Li alloys of Mg.

II. EXPERIMENTAL TECHNIQUE

Oriented single crystal specimens 3" long and 3/8" diam. were prepared from the master alloys, whose compositions are given in Table I using a modified Bridgman technique fully described in a previous paper by Quimby, Mote and Dorn.[5]

Typical variations in the composition along the specimen length are shown in Table II.

The crystals were tested in tension with an Instron TTCL machine. Stresses were measured to within $\pm 2 \times 10^6$ dynes/cm^2 and strains to within ±0.0001. Specimen temperatures were obtained over the temperature range of 4.2°K to about 400°K by immersing them in constant temperature baths which were controlled to within ±2° of the reported values. Strain rates were determined from the rate of cross head motion of the tensile machine. Two different types of tests were employed to obtain the temperature and strain rate dependence of the critical resolve shear stress.

One series of specimens were yielded at various temperatures at constant strain rates of 1.66×10^{-4} and 4.16×10^{-6} per sec as shown in Figs. 1a to 1c. Additional tests were made over the same temperature range in which the strain rate was rapidly changed by a factor of ten. The average strain rate for these tests was about 1.31×10^{-5}/sec.

TABLE I

Chemical Composition in Weight Percent

Alloy No.	At.% Li	Al	Ca	Cu	Fe	Un	Ni	Si	Zn	Mg	Li
1	18.30	<0.03	0.147	<0.001	0.007	<0.01	<0.001	<0.01	<0.02	Bal.	6.01
2	15.62	"	"	"	"	"	"	"	"	Bal.	5.02
3	9.79	"	"	"	"	"	"	"	"	Bal.	3.01

TABLE II

Composition Gradients in Single Crystals

Alloy No.	Crystal No.	At. % Li in gage section Lower End	Upper End
1	1	15.93	15.93
	2	15.65	15.85
	3	15.87	16.36
		Average: 15.9	
2	1	12.91	13.73
	2	12,15	12.79
		Average: 12.9	
3	1	7.15	7.72
	2	8.00	8.77
		Average: 7.9	

III. EXPERIMENTAL RESULTS

The critical resolved shear stress (CRSS) for yielding by prismatic slip vs the temperature is shown in Figs. 1a through 1c for each of two strain rates. Pure Mg crystals fractured in a brittle manner following minor amounts of prismatic slip below about 250°K and the 7.9 at.% Li alloy exhibited the same ductile to brittle transition at about 200°K. In contrast the 12.9 and 15.9 at.% Li alloys underwent extensive straining even at 4.2°K and fractured on the prismatic plane upon incidence of duplex prismatic slip. Whereas those alloys that showed brittle behavior exhibited minor amounts of twinning in the fractured region, no twinning and exclusively prismatic slip was observed in all other cases.

The data recorded in Figs. 1 clearly reveal that Li additions result in solid solution softening of the low temperature prismatic slip in Mg. Whereas, the CRSS for prismatic slip in pure Mg and the 7.9 at.% Li alloy continues to decrease with increasing temperature (Fig. 1a), revealing that prismatic slip is thermally activated in these materials over the entire range of temperatures that were investigated, the CRSS for prismatic slip in the 12.9 and 15.9 at.% Li alloys, following an initial rapid decrease, continues (for temperatures above about 280° and 250°K respectively) to decrease only modestly and almost linearly as the temperature is further increased (Fig. 1b and 1c). Special tests (not recorded in detail here) employing dynamic oscillations revealed that Young's modulus and the shear modulus of elasticity for the Li alloys was practically the same as that for pure Mg and that the temperature variation of these moduli also followed closely that for pure Mg as reported by Roberts.[6] The slope of

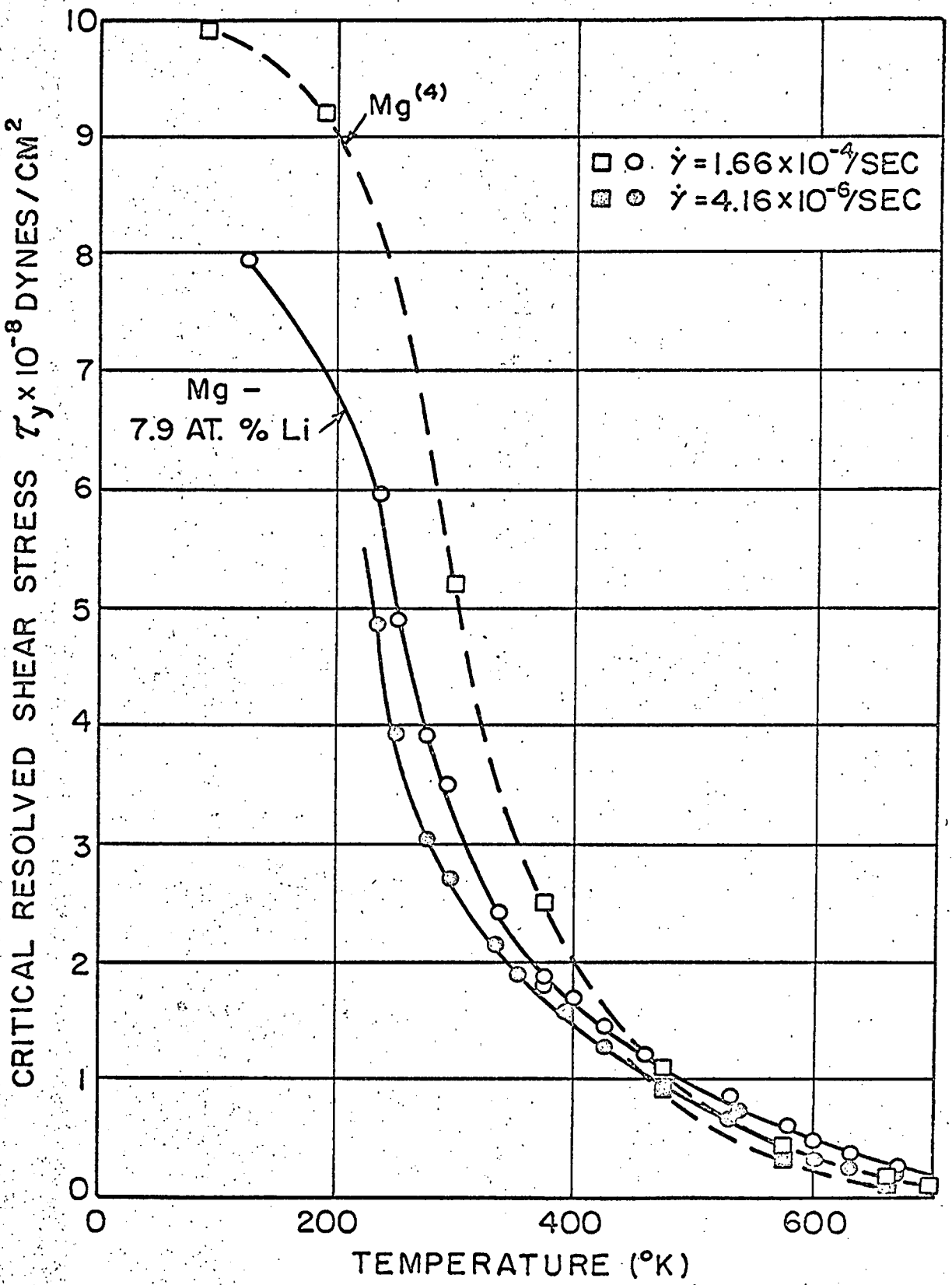

FIG. 1a CRITICAL RESOLVED SHEAR STRESS vs. TEMPERATURE FOR PRISMATIC SLIP IN PURE Mg AND Mg-7.9 AT. % Li ALLOY.

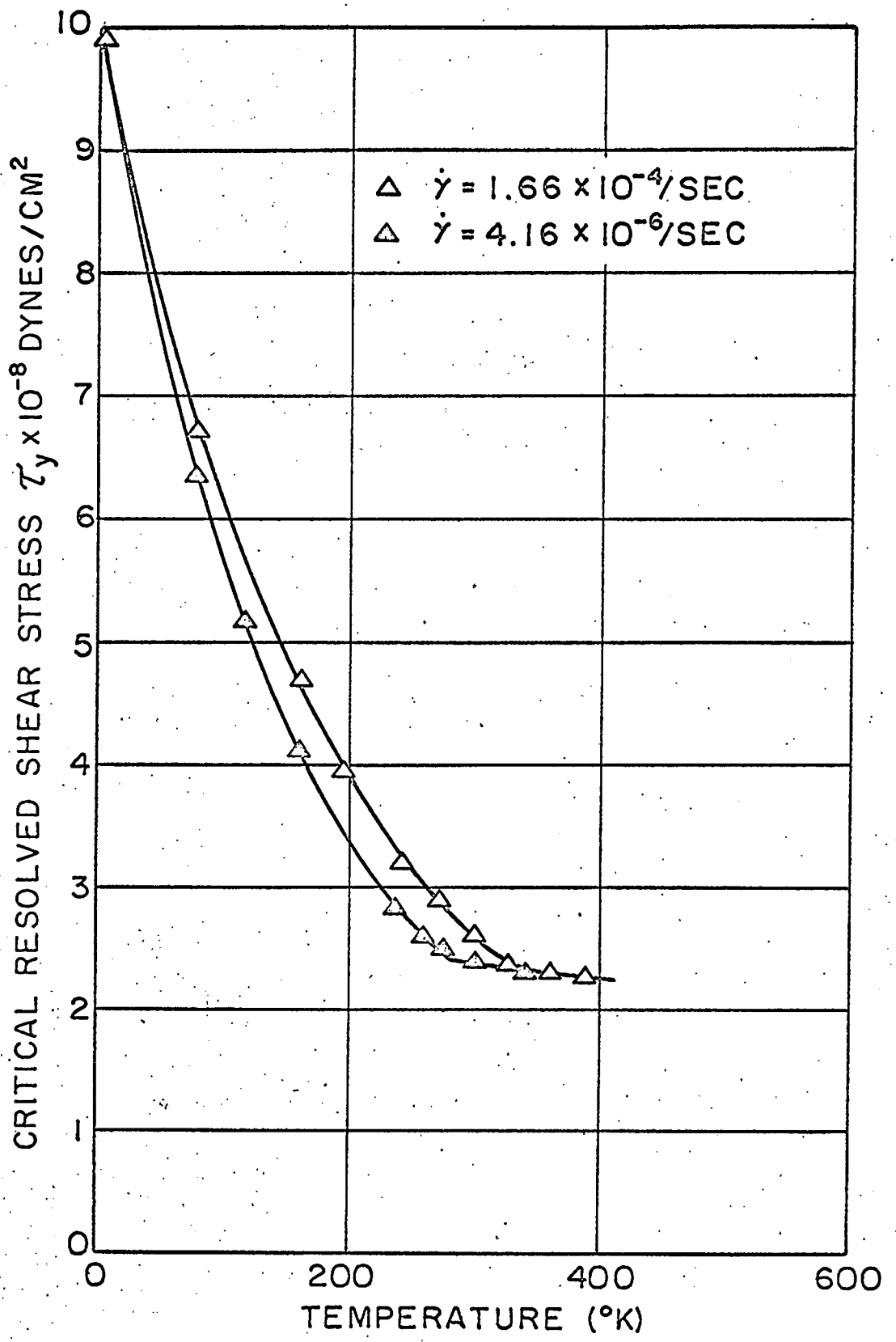

FIG. 1b CRITICAL RESOLVED SHEAR STRESS vs. TEMPERATURE FOR PRISMATIC SLIP IN Mg – 12.9 AT. % Li ALLOY.

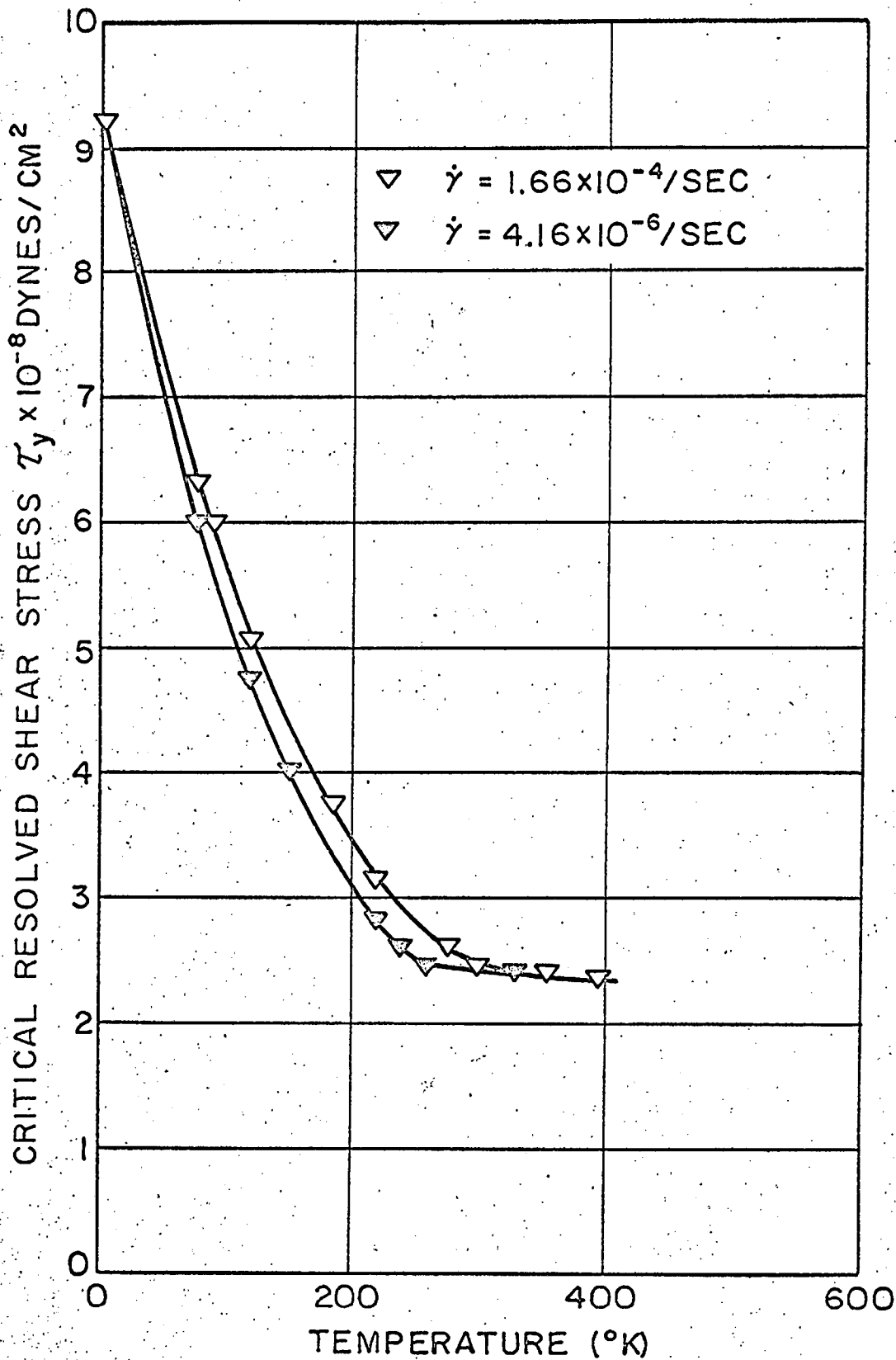

FIG. 1c CRITICAL RESOLVED SHEAR STRESS vs. TEMPERATURE FOR PRISMATIC SLIP IN Mg - 15.9 AT. % Li ALLOY.

the shear modulus vs temperature curve was found to agree reasonably well with the slope of the CRSS vs temperature data recorded in Figs. 1b and 1c above 280° and 250°K for the 12.9 and 15.9 at.% Li alloys indicating that athermal dislocation mechanisms account for the higher temperature prismatic slip in these alloys. Whereas the CRSS for the 12.9 at.% Li alloy was above that for the 15.9 at.% Li alloy at 4°K that for athermal slip above 280°K was slightly higher for the 15.9 at.% Li alloy. Investigations are currently in progress to ascertain whether this athermal behavior is due to short-range ordering in accord with the Fisher[7] mechanism.

Additional data were obtained in terms of

$$\beta = \frac{\partial \ln \dot{\gamma}}{\partial \tau} \simeq \frac{\ln \dot{\gamma}_1 / \dot{\gamma}_2}{\tau_2 - \tau_1} \qquad (1)$$

where $\dot{\gamma}$ is the resolved shear strain rate and τ is the resolved shear stress for slip. For thermally activated mechanisms

$$\dot{\gamma} = A\, e^{-\frac{U}{kT}} \qquad (2)$$

where A is dependent on the dislocation substructure and either independent or only mildly dependent on τ, U is the activation energy, which depends on τ and occasionally also on the dislocation substructure and kT has its usual meaning of the Boltzmann constant times the absolute temperature. Consequently, the apparent activation volume, v, is defined by

$$v = \beta kT \simeq -\frac{\partial U}{\partial \tau} \qquad (3)$$

Such apparent activation volumes are rather sensitive guides for ascertaining the operative dislocation mechanisms. For example for the intersection mechanism, the activation volumes vary from about 500 to 100 b^3 (b is the Burger's vector) and, being dependent on the dislocation forest spacing,

decrease with increasing strain.[8] Activation volumes for the cross-slip mechanism are of the same order as those for intersection.[4] Activation volumes for the Peierls process are lower varying from about 80 b^3 to about 5 b^3, independent of the dislocation density, and decrease with increasing values of the thermal stress $\tau^* = \tau - \tau_A$ where τ_A is the athermal component of the stress.[17] For athermal mechanisms, τ does not change with modest changes in $\dot{\gamma}$ and v approaches infinity.

Apparent activation volumes were obtained from the data for the two strain rates recorded in Figs. 1 and additional tests involving the effect of rapid changes in strain rate on the stress during the course of straining. The latter tests gave apparent activation volumes that were in agreement with those calculated from Figs. 1 and were essentially independent of the strain for all ranges of tests. The apparent activation volumes that were obtained are recorded as a function of test temperature in Fig. 2. The 15.9, 12.9 and 7.9 at.% Li alloys exhibited quite small apparent activation volumes below about 200°K. The extremely high activation volumes that were obtained between about 250°K and 350°K for the 15.9 and 12.9 at.% Li alloys are further reflections of the athermal prismatic slip that these alloys undertake in this range. The decrease in the apparent activation volume for these two alloys as the temperature is further increased reveals the introduction of an additional thermally activated mechanism of prismatic slip at yet higher temperatures. Slightly above 250°K, the apparent activation volume for the 7.9 at.% Li alloy began to increase rapidly but then it exhibited an inversion following which the apparent activation volume again increased ever more rapidly with temperature.

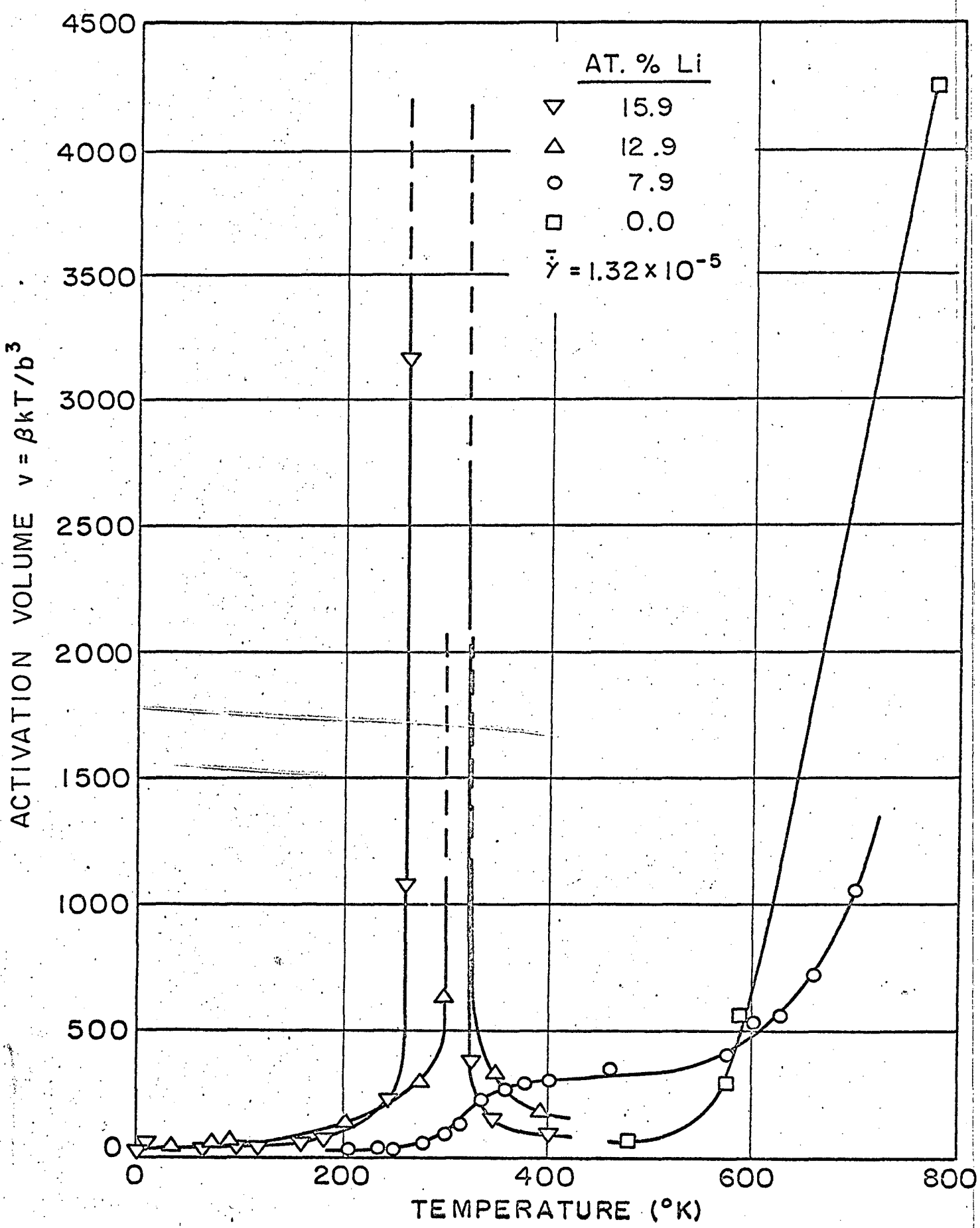

FIG. 2 APPARENT ACTIVATION VOLUME vs. TEMPERATURE FOR PURE Mg AND Mg-Li ALLOYS.

The trend in the latter stage of this behavior is somewhat parallel to that recorded for pure Mg.[4] At the lower temperatures, pure Mg fractured following very small amounts of prismatic slip.

IV. DISCUSSION

The experimental data documented in the preceding section clearly reveal the operation of several different dislocation mechanisms for prismatic slip in Mg over the range of temperatures and Li contents that were investigated. The behavior of the 15.9 and 12.9 at.% Li alloys are quite similar, which suggests that they undertake the same kind of mechanisms. It will be demonstrated that they follow closely the dictates of the Peierls mechanism from 0° to 250° or 280°K respectively. Their athermal behavior above these upper temperatures is probably ascribable to short-range ordering, an issue for future investigation. A complete analysis of the low-temperature plastic behavior of the 7.9 at.% Li alloy cannot be made since its yield stress exceeds its fracture strength below about 200°K. But as will be shown, the τ - T trend and the activation volumes from 200°K to slightly above 280°K are consistent with the dictates of the Peierls process for an alloy that has a very high Peierls stress. The CRSS for slip and the apparent activation volumes for the 7.9 at.% Li alloy above about 450°K parallel analogous data for high temperature prismatic slip in pure Mg. Flynn, Mote and Dorn[4] have demonstrated that the Friedel cross-slip mechanism[9] applies here. It will be demonstrated that this is also true for the higher temperature prismatic slip in the 7.9 at.% Li alloy.

The low-temperature prismatic slip in the 15.9 and 12.9 at.% Li alloys cannot be ascribed to either the intersection or the cross-slip mechanism because of the extremely small apparent activation volumes that were obtained. Furthermore, their mechanism of slip cannot be due to inter-

actions of stress fields due to solute atoms with dislocations because the thermally variable stress τ^* decreases with increasing Li content. Consequently, the only known mechanism that cannot be disqualified on these bases is the Peierls mechanism.

The analyses of these data will be based on the formulation of the Peierls mechanism, as presented by Dorn and Rajnak,[10] since it has now been verified by agreement with the low temperature plastic behavior of three different alloys, namely $\{10\bar{1}0\}\langle\bar{1}210\rangle$ slip in Ag_2Al,[11] $\{123\}\langle 11\bar{1}\rangle$ slip in AgMg[12] and for low temperature deformation in a polycrystalline Fe containing 2 at.% Mn.[13] This theory gives A of Eq.2 as

$$A = \rho a b \nu \frac{L}{w} \tag{4}$$

where ρ is the density of mobile dislocations, a is the distance between the Peierls valleys on the slip plane, b is the Burger's vector, ν is about the Debye frequency, L is the distance swept out by a pair of kinks, and w is approximately the separation of a pair of kinks at the saddle-point free energy for nucleation of kinks. As discussed by Dorn and Rajnak,[10] the only term in A that depends upon the stress is w and its variation with stress is so minor in contrast to the major effect of the stress on the activation energy that w might be assigned some average constant value. The value of τ^*/τ_p is given by the theory as a universal function of $U/2U_k$ as shown in Fig. 3, where U is the saddle-point free energy for the nucleation of a pair of kinks under stress τ^*, U_k is the energy of a stable kink ($\tau^* = 0$), τ^* is the thermally affected stress, τ_A is the athermal stress which is not surmountable by thermal fluctuations and τ_p is the Peierls stress. The values of α (vide Dorn and Rajnak)[10] given in

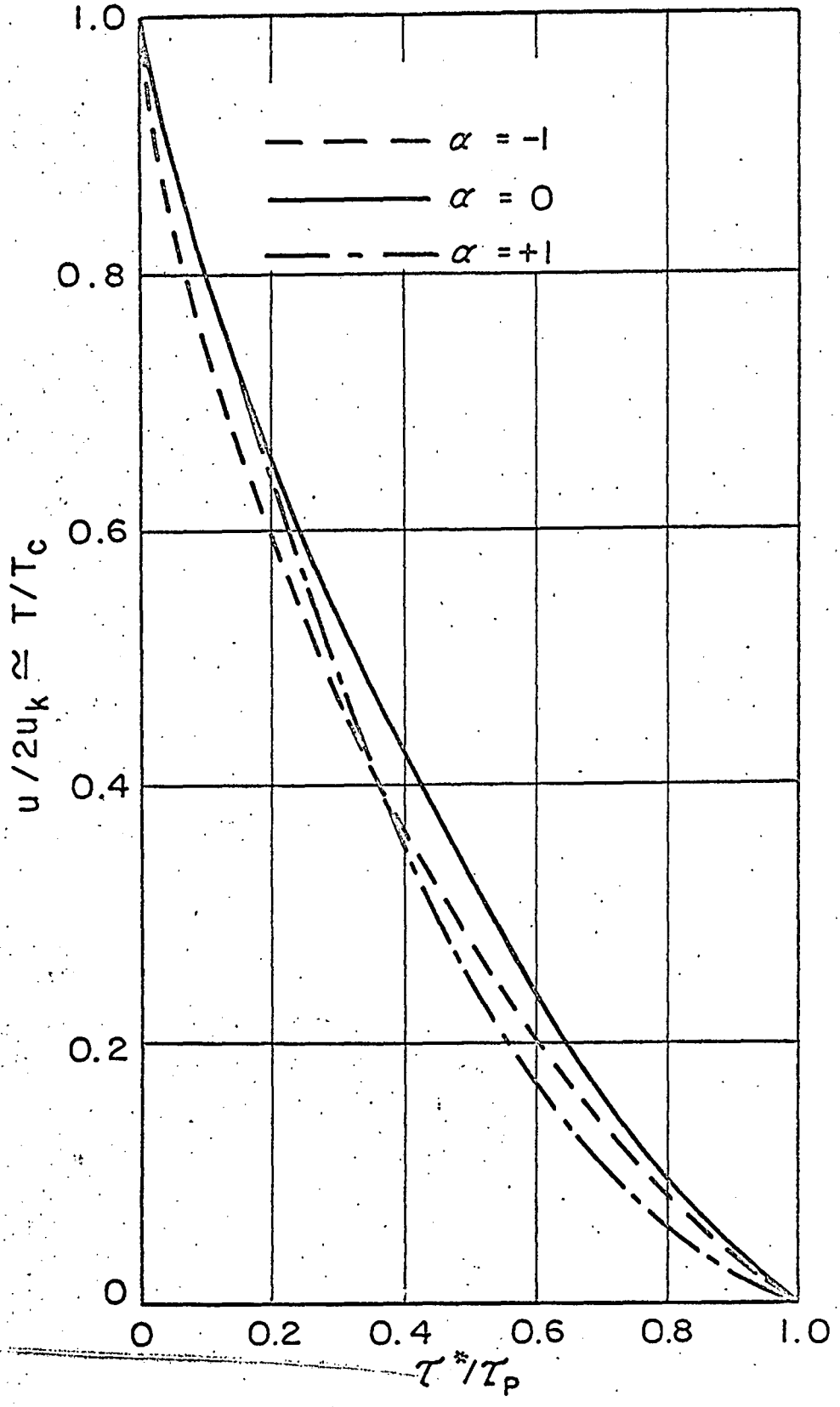

FIG. 3 EFFECT OF STRESS ON THE ACTIVATION ENERGY FOR KINK NUCLEATION.

Fig. 3 refer to deviations of the Peierls potential hill from purely sinusoidal ($\alpha = 0$) shape.

It is assumed that

$$\tau_A = \tau_A^o \, G/G^o \quad \text{and} \quad \tau_p = \tau_p^o \, G/G^o \tag{5}$$

as discussed by Friedel,[14] where G is the shear modulus of elasticity and the superscript zero refers to the values of these quantities at the absolute zero. The variation of G/G^o with temperature used here was calculated from the data given by Roberts.[6] A typical example of the data when the stress τ_A is subtracted from the total stress τ to give τ^* is illustrated in Fig. 4 for the 12.9 at.% Li alloy. Whereas, in accord with theory τ_p^o is τ^* at $T = 0$, τ_p decreases with increasing temperature as shown by the broken line in Fig. 4. The values of τ_p^o obtained in this way are given in Table III.

The activation energy and the kink energy depend on the temperature according to

$$U = U^o \, G/G^o \text{ and } U_k = U_k^o \, G/G^o \tag{6}$$

Thus the strain rate (Eq. 2) becomes

$$\dot{\gamma} = \rho ab \frac{L}{w} \nu \, e^{-\frac{U^o G/G^o}{kT}} \tag{7}$$

At $T = T_c$ (vide Fig. 4) $\tau^* = 0$ and $U = 2U_k$, whence

$$\dot{\gamma} = \rho ab \frac{L}{w} \nu \, e^{-\frac{2U_k^o \, G^{T_c}/G^o}{kT_c}} \tag{8}$$

For the same strain rate, and density of dislocations, therefore, as shown by Eqs. 7 and 8

$$\frac{U^o}{2U_k^o} = \frac{T}{T_c} \, \frac{G^{T_c}}{G} \tag{9}$$

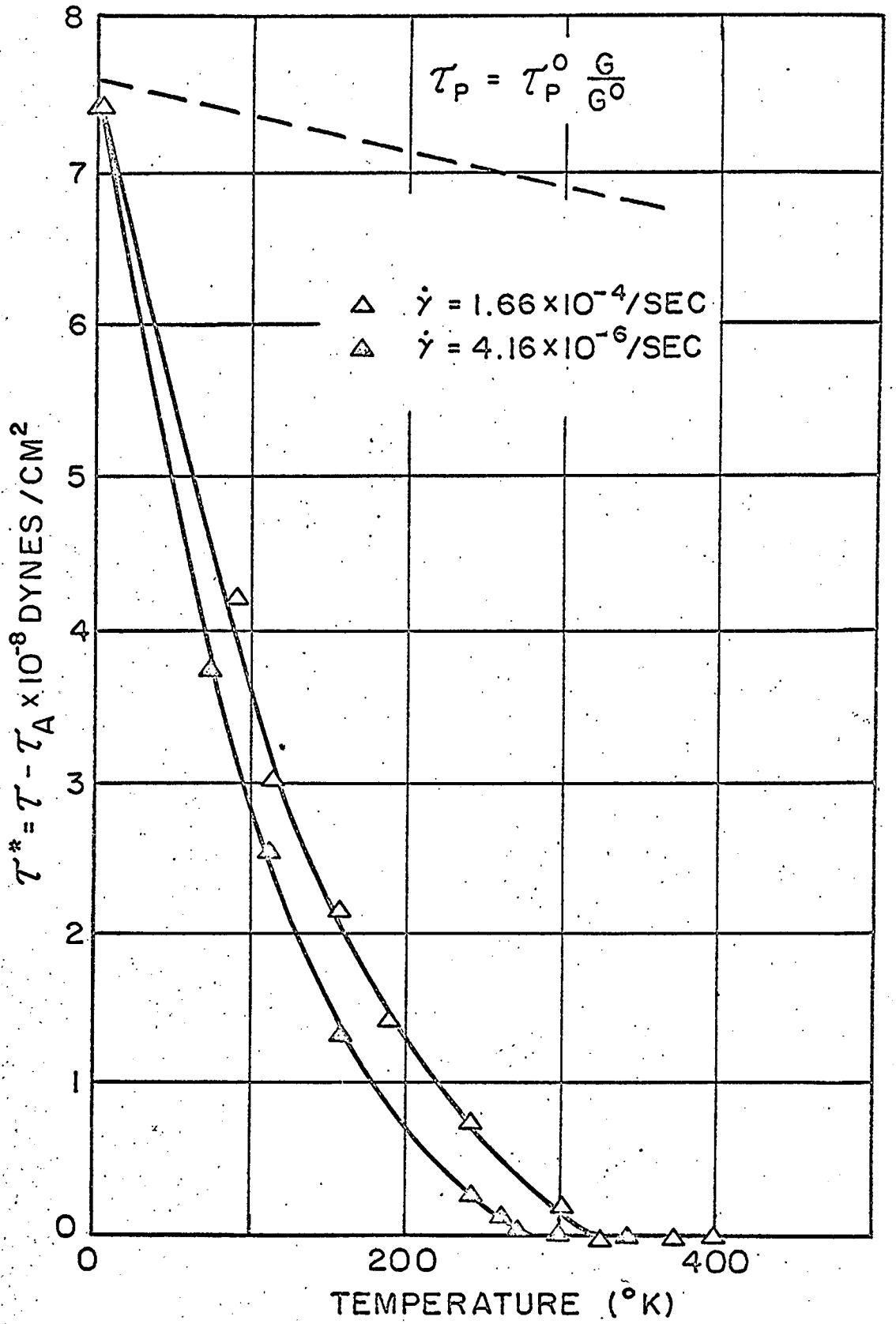

FIG. 4 THERMALLY ACTIVATED COMPONENT OF THE FLOW STRESS vs. TEMPERATURE FOR 12.9 AT. % Li-Mg ALLOY.

TABLE III

Kink Energies and Line Energies Deduced from the Peierls Analysis

Alloy at.% Li	τ_p^o dynes/cm^2	U_k^o ergs	Γ_o^o ergs/cm	$G^o b^2$ ergs/cm
15.9	6.8×10^8	0.41×10^{-12}	4.74×10^{-4}	1.85×10^{-4}
12.9	7.6×10^8	0.43×10^{-12}	4.67×10^{-4}	1.85×10^{-4}

Consequently, the theoretical relationship

$$\frac{\tau^*}{\tau_p} = f\left(\frac{U}{2U_k}\right) = f\left(\frac{U^o}{2U_k^o}\right) \tag{10}$$

given in Fig. 3 requires that

$$\frac{\tau^*}{\tau_p} = f\left(\frac{TG^{T_c}}{T_c G}\right) \tag{11}$$

The experimental data points for both strain rates for the 12.9 at.% Li alloy and the corresponding data points for the 15.9 at.% Li alloy replotted in accord with Eq. 11, are shown in Fig. 5. The curves in this figure refer to the theoretical curves of Fig. 3. As noted the data are in good agreement with the theory for $\alpha \simeq -1$.

The kink energy was determined from the two values of T_c for the two different strain rates that were employed as demanded by Eq. 8 which gives

$$\frac{\dot{\gamma}_1}{\dot{\gamma}_2} = \frac{e^{-\frac{2U_k^o G^{T_{c_1}}}{kT_{c_1} G^o}}}{e^{-\frac{2U_k^o G^{T_{c_2}}}{kT_{c_2} G^o}}} \tag{12}$$

Since the values of T_{c_1} and T_{c_2} were not very different, only approximate values of the kink energies, given in Table III, could be obtained. As shown, the Peierls stress appears to decrease with increasing Li content.

The most discriminating correlation for validating the Peierls mechanism is the dependence of the activation volume on the stress. The theoretical deductions given by Dorn and Rajnak[10] reveal that the activation volume v is related to the stress as shown by the curves of Fig. 6.

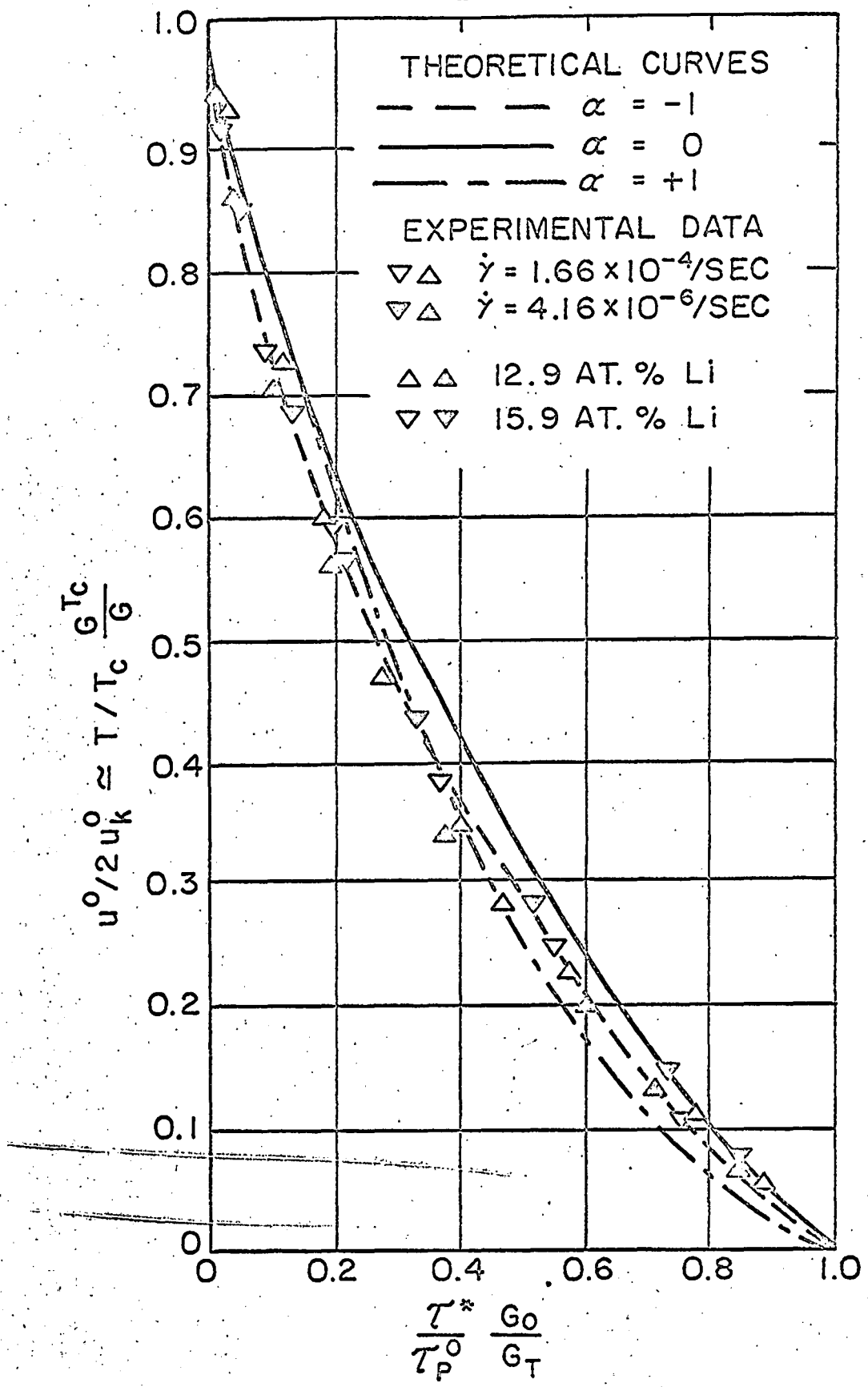

FIG. 5 TEMPERATURE vs. THERMALLY ACTIVATED FLOW STRESS IN DIMENSIONLESS UNITS FOR 12.9 AND 15.9 AT. % Li ALLOYS.

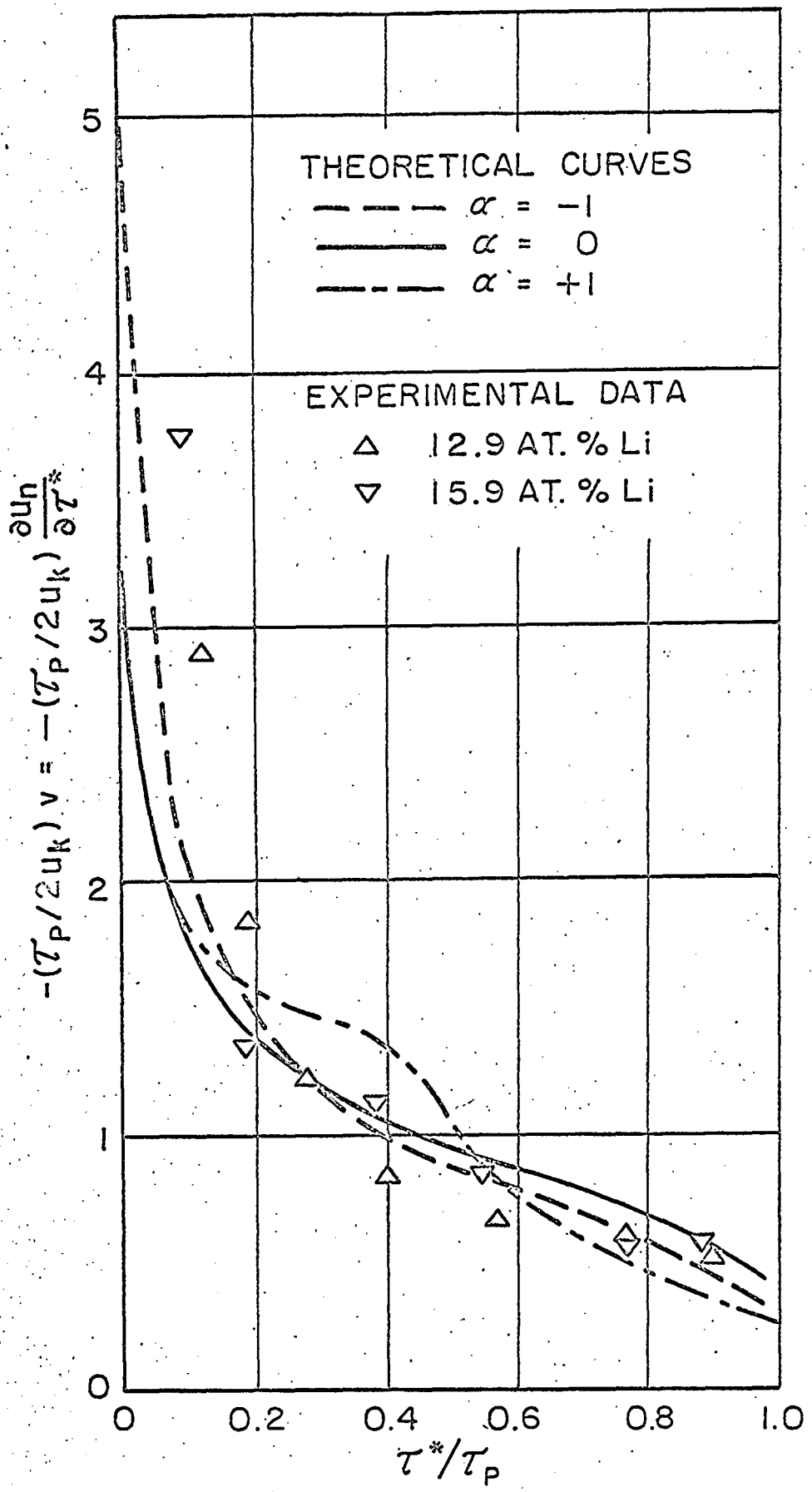

FIG. 6 ACTIVATION VOLUME vs. THERMALLY ACTIVATED COMPONENT OF FLOW STRESS IN DIMENSIONLESS UNITS FOR 12.9 AND 15.9 AT. % Li

The experimental points for the 12.9 and 15.9 at.% Li alloys were deduced from the activation volumes given in Fig. 2 using the values of τ_p^o and U_k^o recorded in Table III.

As shown by Dorn and Mitchell[15]

$$\frac{2\pi U_k^o}{a\Gamma_o^o} = 3.8\left(\frac{\tau_p^o\, ab}{\pi\Gamma_o^o}\right)^{1/2} \quad \text{for} \quad \alpha = -1 \tag{13}$$

where Γ_o^o is the line energy, namely

$$\Gamma_o^o \simeq G^o b^2 \tag{14}$$

as indicated by Friedel.[16]

Since all of these quantities are now known excepting Γ_o^o, it is possible to estimate the line energy. Although the values of Γ_o^o deduced in this way are somewhat greater than about Gb^2 (vide Table III), the theoretically accepted order of magnitude,[16] the correlation is nevertheless deemed to be satisfactory. Such differences might have arisen from approximations[16] in using the isotropic theory of elasticity for estimating theoretically the line energy to be Gb^2.

Thus, the correlation between theory for the Peierls mechanism and the experimental results for τ vs T, the activation volume, and the dislocation line energy all substantiate the thesis that the low temperature prismatic slip in the 15.9 and 12.9 at.% Li alloys obeys the Peierls process.

Obviously the same type of analysis cannot be made on the low temperature data for the 7.9 at.% Li alloy because it exhibited brittle fracturing below 200°K. But the τ - T relationship from 200° to 280°K (Fig. 1a) and

the activation volumes over the same range (Fig. 2) are consistent with the expectations of the Peierls mechanism for an alloy that has a very high Peierls stress. As shown by the τ - T curves of Fig. 1a and the activation volume data of Fig. 2, the 7.9 at.% Li alloy does not exhibit athermal behavior in the vicinity of 300°K as did both the 12.9 and 15.9 at.% Li alloys. Rather above 300°K, the trends in the activation volume data indicate transitions in mechanisms. As will be shown, above about 450°K both the 7.9 at.% Li alloy and pure Mg appear to follow the cross-slip mechanism.

Dorn and Mitchell[15] have demonstrated that Friedel's theory for cross slip[9] upon minor revision is given by

$$\dot{\gamma} = \frac{NL_sAb^4\nu\tau^{*2}}{8\Gamma R} e^{-\left\{\frac{U_c}{kT} + \frac{2^{3/2}(\Gamma R^3)^{1/2}}{\tau^* b\ kT}\right\}} \tag{15}$$

where N is the number of screw segments of dislocations per unit volume dissociated on the basal plane, each having length L_s, A is the area swept out per activation, ν is the Debye frequency, R is the recombination energy per unit length, U_c is the constriction energy and the remaining symbols have the previously quoted meanings. Rearranging Eq. 15 gives

$$\frac{1}{\tau^* T} = \frac{kb}{2(2\Gamma R^3)^{1/2}} \ln \frac{NL_sAb^4\nu}{8\dot{\gamma}\Gamma R} + \frac{kb}{(2\Gamma R^3)^{1/2}} \ln\tau^* - \frac{bU_c}{2(2\Gamma R^3)^{1/2}T} \tag{16}$$

which reveals that over the range of conditions where the deformation is controlled by the cross-slip mechanism $1/\tau^*T$ varies linearly with $\frac{1}{T}$ with a slope $-S_1 = -bU_c/2(2\Gamma R^3)^{1/2}$. The data for pure Mg and the 7.9 at.% Li alloy plotted in this way are given in Fig. 7 where it is assumed that at $\tau \approx \tau^*$. This appears to be justified by the decrease of stress

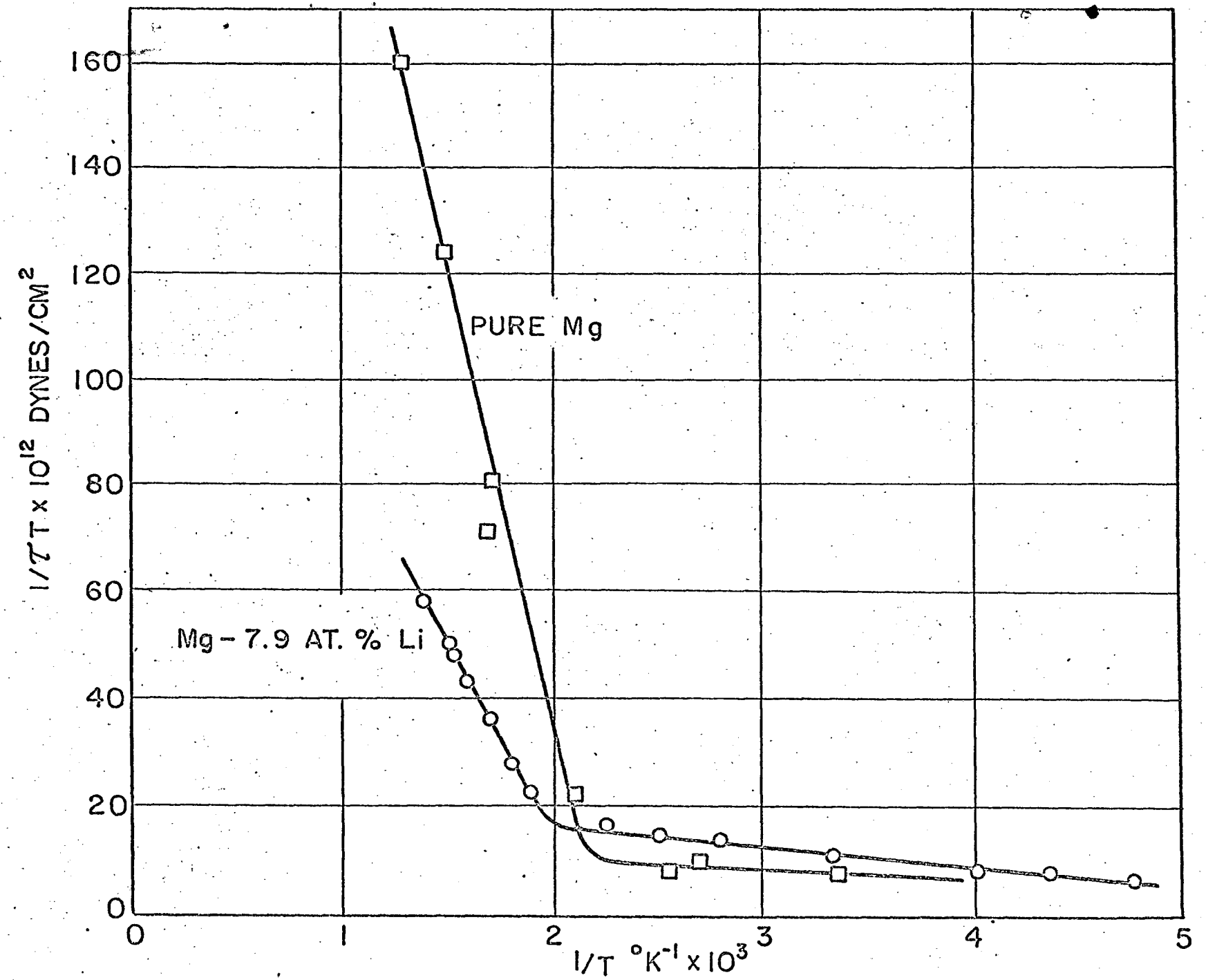

FIG. 7 $1/\tau T$ vs. $1/T$ FOR PRISMATIC SLIP IN Mg AND Mg-7.9 AT. % Li.

above about 450°K to very low stress as shown in Fig. 1a indicating the athermal contribution to the flow stress τ_A is negligible.

From the effect of strain rate on the flow stress, it also follows that

$$\beta kT = kT \frac{\partial \ell n \dot{\gamma}}{\partial \tau^*} = \frac{2kT}{\tau^*} + \frac{2(2\Gamma R^3)^{1/2}}{\tau^{*2} b} \tag{17}$$

or

$$(\beta - \frac{2}{\tau^*})\ kT = \frac{2(2\Gamma R^3)^{1/2}}{\tau^{*2} b} \tag{18}$$

revealing that $(\beta - 2/\tau^*)kT$ increases linearly with $1/\tau^{*2}$ when the deformation is controlled by the cross-slip mechanism, where the slope of the curve is $S_2 = 2(2\Gamma R^3)^{1/2}/b$. The data for pure Mg and the 7.9 at.% Li alloy plotted in this way are given in Fig. 8. The data in Figs. 7 and 8 reveal that both pure Mg and the 7.9 at.% Li alloy follow closely the expected trends for thermally activated prismatic slip as controlled by the cross-slip mechanism from the basal to the prism plane for temperatures in excess of about 450°K. ($\frac{1}{T} < 2.1 \times 10^{-3}$)

Values of $(\Gamma R^3)^{1/2}$ and U_c were deduced from the slopes of Figs. 7 and 8 and are shown in columns 3 and 4 of Table IV. The estimate of the constriction energy, given by Stroh,[17] is

$$U_c \simeq \frac{Gb^2 d}{30} \left(\ell n \frac{d}{b}\right)^{1/2} \tag{19}$$

where d is the equilibrium separation of the Shockley partial dislocations on the basal plane. Furthermore, Nabarro[18] has shown that

$$R/\Gamma \simeq \frac{1}{16} \ell n \frac{d}{eb} \tag{20}$$

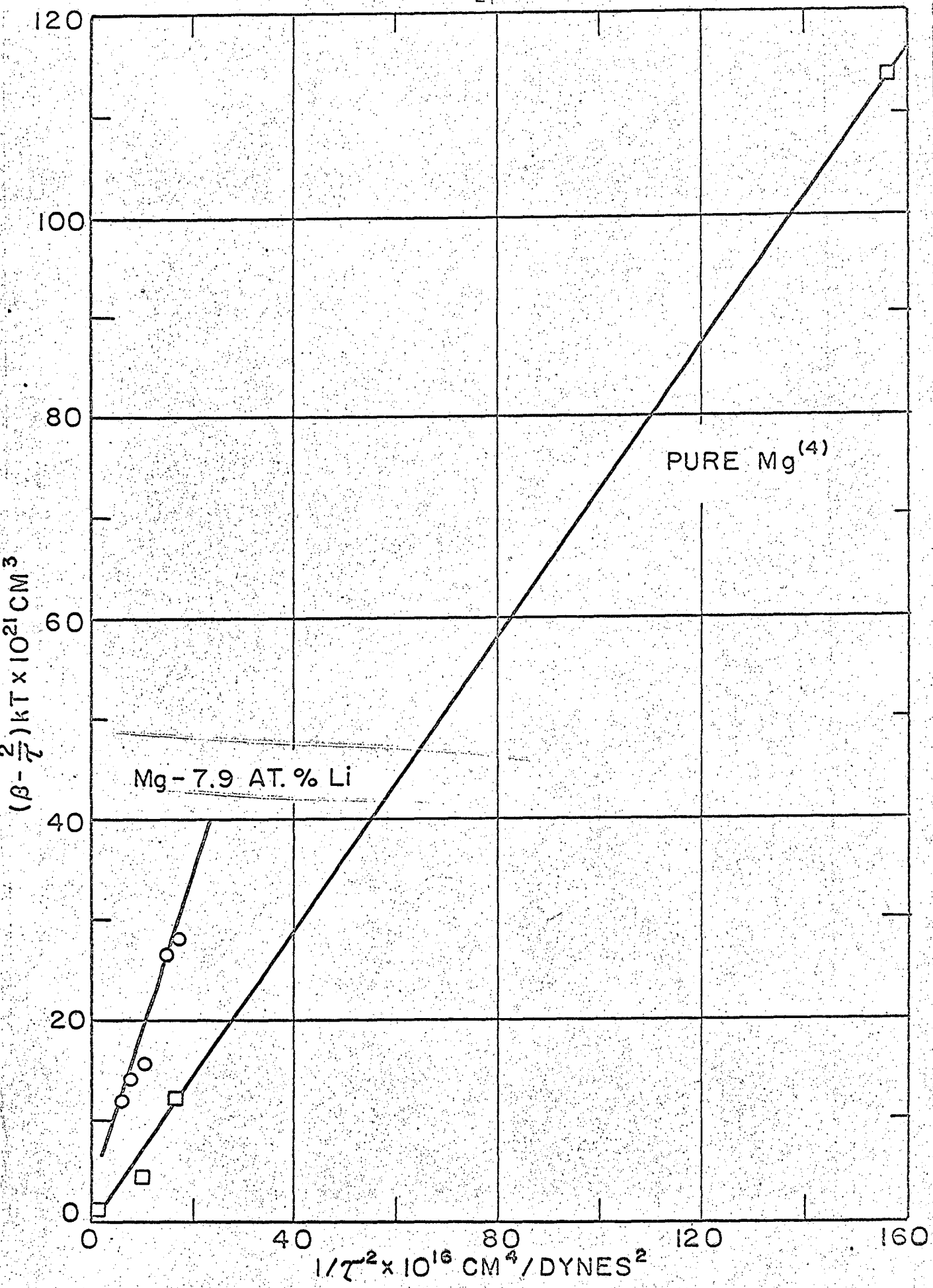

FIG. 8 $(\beta - 2/\tau)kT$ vs. $1/\tau^2$ FOR PURE Mg AND Mg - 7.9 AT. % Li

TABLE IV

Constriction Energies and Stacking Fault Widths
Deduced from the Cross-Slip Analysis

(1)	(2)	(3)	(4)	(5)	
Alloy	$U_c/(\Gamma R^3)^{1/2}$ $\frac{cm}{dyne}$	$(\Gamma R^3)^{1/2}$ $dynes^2$	U_c ergs	d/b	
From	Fig.1	Fig. 8	(2) and (3)	Eq. 19	Eq. 20
Mg[4]	15.2	0.84×10^{-13}	1.28×10^{-12}	~ 6.5	~ 3.0
Mg at. 7.9 at.% Li	6.3	1.70×10^{-13}	1.07×10^{-12}	~ 5.5	~ 3.0

where Γ is given by Eq. 14. The values of d/b deduced by use of these equations is given in the 5th column of Table IV. In view of the approximations of Eqs. 19 and 20, the agreement between the stacking fault widths deduced from these equations are as good as might be expected. These values also agree with the $d/b \approx 2$ deduced by Dorn and Mitchell[15] from basal slip data for pure Mg where the dislocation intersection mechanism is rate controlling. These data further confirm the thesis that the high temperature thermally activated dislocation mechanism for prismatic slip in pure Mg and the 7.9 at.% Li alloy is controlled by the cross-slip mechanism and they also reveal that pure Mg and the 7.9 at.% Li alloy have about the same high stacking fault energies.

CONCLUSIONS

1. Increased Li content in Mg-Li solid solutions results in a substantial decrease in the critical resolved shear stress for prismatic slip at low temperatures. Whereas at the higher temperatures, addition of Li increases the critical resolved shear stress for slip.
2. Over the range of temperatures and compositions investigated the alloys exhibit several deformation controlling mechanisms.
3. Prismatic slip in the 12.9 and 15.9 at.% Li alloys over the temperature region from 4.2 to about 300°K is controlled by the Peierls mechanism. Between about 300°K and 400°K the deformation is controlled by an athermal process which might arise from the short-range ordering mechanism suggested by Fisher.[7]
4. The rate controlling mechanism of prismatic slip in the 7.9 at.% Li alloy as well as in pure Mg above about 450°K appears to be the result of cross-slip of dislocation segments lying in screw orientation on the basal planes as proposed by Friedel.[9] Over the temperature region from 450°K to 200°K below which twinning and fracturing predominate, the trends of the τ - T curves as well as the activation volumes suggest that the Peierls mechanism controls.
5. The correlation of the data for the Mg-7.9 at.% Li alloy and pure Mg in the region where the cross-slip mechanism controls the rate of prismatic slip indicates that Li additions up to 7.9 at.% have only a minor effect on the stacking fault width.

6. The major effect of Li additons on the low temperature shear stress for prismatic slip appears to be due to its effect in lowering the Peierls stress. The theoretical reason for this effect is not known at present.

7. The increase in the shear stress for the athermal prismatic slip with increasing Li content in the higher temperature region of the alloys containing over 7.9 at.% Li might be due to short-range order strengthening.

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REFERENCES

1. F. E. Hauser, P. R. Landon, and J. E. Dorn, Trans. ASM, 48, 589-593 (1956).

2. F. E. Hauser, P. R. Landon, and J. E. Dorn, Trans. ASM, 50, 856-883 (1958).

3. W. F. Sheely and R. R. Nash, Trans. AIME, 218, 416-423 (1960).

4. P. Ward Flynn, J. D. Mote and J. E. Dorn, Trans. AIME, 221, 1148-1162 (1961).

5. R. M. Quimby, J. D. Mote and J. E. Dorn, Trans. ASM, 55, [1], 149-157 (1962).

6. C. S. Roberts, Magnesium and Its Alloys, (J. Wiley, New York, 1960) pp. 82-83.

7. J. C. Fisher, Acta Met., 2, 9-17 (1954).

8. S. K. Mitra and J. E. Dorn, Trans. AIME, 224, 1062-1071 (1962).

9. J. Friedel, "Dislocations Integrations and Internal Stresses," Physiques des Solides (Sorbonne, Paris, 1959) pp. 220-262.

10. J. E. Dorn and S. Rajnak, Trans. AIME, 230, 1052-1064 (1964).

11. A. Rosen, J. D. Mote and J. E. Dorn, Trans. AIME, 230, 1070-1074 (1964).

12. A. K. Mukherjee and J. E. Dorn, Trans. AIME, 230, 1065-1070 (1964).

13. P. Wynblatt, A. Rosen and J. E. Dorn, "The Effect of Temperature, Strain Rate and Structure on the Flow Stress of an Iron-2% Manganese Alloy," (to be published Trans. AIME, 1965).

14. J. Friedel, Electron Microscopy and Strength of Crystals, (J. Wiley, New York and London, 1963) pp. 605-649.

15. J. E. Dorn and J. B. Mitchell, Proceedings Second International Symposium on Materials, University of California, Berkeley, California (to be published).

16. J. Friedel, Dislocations, (Addison-Wesley Publishing Co. Inc., 1964) p. 62.

17. A. N. Stroh, Proc. Phys. Soc. (London)B., 67, 427-436 (1954).

18. F. R. N. Nabarro, Advances in Physics, (Interscience Publishers, 1962) pp. 271-334.

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