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DISLOCATION DYNAMICS

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DISLOCATION DYNAMICS

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PLASTIC RESPONSE OF CRYSTALLINE MATERIALS
TO DYNAMIC CONDITIONS OF STRESSING

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DISLOCATION DYNAMICS

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ABSTRACT

A brief summary is presented on the current status of knowledge regarding the dynamic behavior of crystalline materials in terms of athermal, thermally activated, viscous damping, and relativistic motion of dislocations. It is shown that dislocation concepts have contributed substantially to a better and more unified rationalization of the dynamic behavior of crystalline materials than heretofore possible. Particular emphasis is given to the need for a more complete theoretical understanding of dislocation mechanics. And, the requirements for more detailed and better experiments is expressed particularly for dislocation velocity determinations in the viscous damping and relativistic ranges of conditions.

I. INTRODUCTION

Most of our knowledge on the dynamic behavior of materials is largely empirical and somewhat questionable. This arises because it is analytically impossible to extract the dynamic behavior of materials directly from evidence based on conventional plastic wave propagation experiments. For example, in the von Karman¹ and the Taylor² theories of plastic wave propagation the *a priori* assumption is made that the stress is only a function of the strain; and in the Malvern³ type of approach the strain rate is assumed to depend on the stress rate, the stress and the strain. Only after having made such assumptions is it possible to integrate the equations for equilibrium and continuity along characteristics to obtain a solution to the wave propagation problem. Although other laboratory methods, such as the Kolsky thin wafer technique,⁴ provide the basis for experimental approaches that do not demand *a priori* assumptions regarding the dynamical behavior of materials, they have other limitations such as inaccuracy, especially over the first few microseconds duration of the passage of a wave. Thus it is not surprising that much confusion continues to persist regarding whether or not the dynamic behavior of a given material is strain rate sensitive.⁵ Furthermore, no systematic or otherwise logical correlations on the dynamic behavior of materials seems to have matured solely as a result of empirical approaches to the problem.

However, as a result of recent and rapid advances in knowledge, both experimental and theoretical, of dislocation behavior, a new insight has been gained in our understanding of the dynamic plastic behavior of crystalline materials. It is the purpose of this paper to summarize

briefly some of the major advances that have already been made and to indicate those areas which appear to be especially deserving of additional attention. For this purpose, the following types of mechanisms will be discussed in sequence, athermal, thermally activated, high velocity damping, and relativistic motion; this is the usual order of behavior with increasing stress or strain rate. The separation between these types of mechanisms is not always sharp, however, and several may often be effective simultaneously.

II. ATHERMAL MECHANISMS

Plastic deformation of single crystals is the result of glide displacement of dislocations. During such glide, moving dislocations react with each other, with other dislocations, and with the lattice itself. When the reaction fields are high, the energy to cause continued motion of the dislocation cannot be supplied by thermal fluctuations and the work of moving the dislocation is done completely mechanically by the applied stress. The stress necessary to cause plastic deformation (dislocation motion) in this event is not dependent on the strain rate except perhaps at high strain rates where other types of phenomena intervene. Since the energies of the reactions that take place are usually proportional to the shear modulus of elasticity, the critical resolved shear stress for deformation generally decreases slightly with an increase in temperature just as does the shear modulus of elasticity. A typical example of such athermal behavior is illustrated over Region II of Fig. 1 where the flow stress at yielding is independent of the strain rate and decreases mildly with an increase in temperature. The deformation stress for most athermal processes may increase either slowly or rapidly with continued straining depending on the redistribution of dislocations. Consequently when deformation occurs exclusively by an athermal mechanism, the von Karman assumption,¹ that the deformation stress is a function of the strain alone, is good.

Various known athermal dislocation mechanisms are summarized in Table I.

Under appropriate conditions several of these mechanisms might be operative simultaneously. In this event, the deformation stress is a

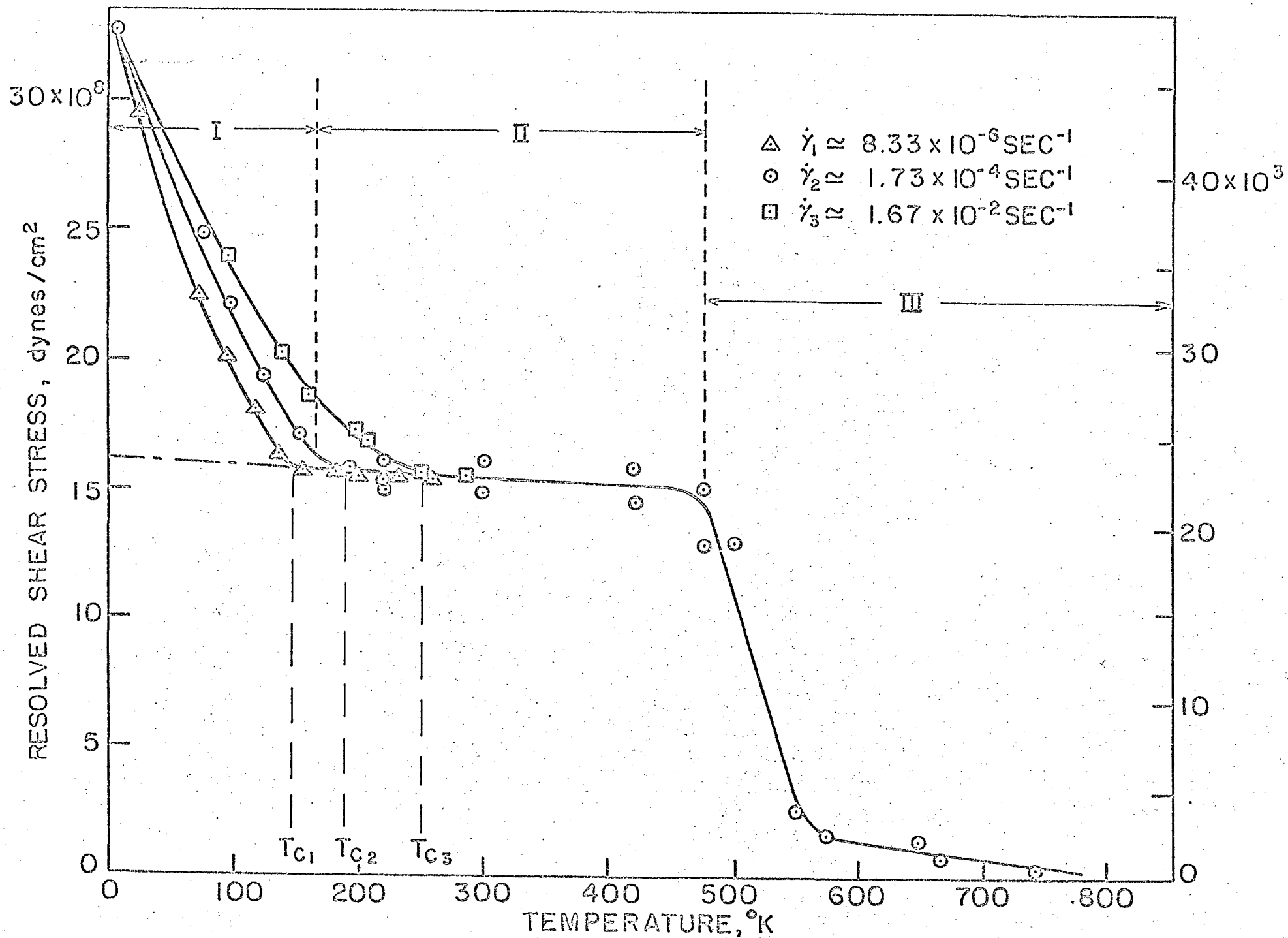


FIG. 1 EFFECT OF TEMPERATURE ON THE CRITICAL RESOLVED SHEAR STRESS FOR PRISMATIC SLIP OF Ag-33 AT. % Al SINGLE CRYSTALS.

TABLE I

Athermal Mechanisms

-
-
1. Surmounting Long-Range Stress Fields⁶
 2. Breaking Attractive Junctions⁷
 3. Making Antiphase Boundary in Long-Range Ordered Alloys⁸
 4. Breaking Bonds in Short-Range Ordered Alloy⁹
 5. Overcoming Strong Solute Atom Locking¹⁰
 6. Overcoming Suzuki Locking Effects¹¹
-
-

complicated function of the operative processes. For example, a short-range ordered alloy (No. 4) might at first reveal a slight decrease in the deformation stress with straining because of a decrease in order with deformation but it may later exhibit strain hardening because of dislocation interactions arising from mechanisms Nos. 1 and 2. Mechanisms Nos. 5 and 6 usually reveal a yield point caused by a decrease in the deformation strength once dislocations are unlocked or new dislocations are generated at points of high stress concentration.

III. THERMALLY ACTIVATED MECHANISMS

Frequently the stress is not sufficiently high to push dislocations past the barriers they encounter. If, however, the barriers are localized and their energy is not excessively high, thermal fluctuations over restricted regions, assisted by the applied stress, can cause dislocations to surmount them and thereby continue to move. Two equivalent expressions give the resolved shear-strain rate, $\dot{\gamma}$, namely

$$\dot{\gamma} = \rho b \bar{v} \quad (1a)$$

or

$$\dot{\gamma} = N A b \nu \quad (1b)$$

where ρ is the density of the mobile dislocations, b is their Burger's vector, $\bar{v}$ their mean velocity, N is the number of points per unit volume at which dislocations are arrested at barriers, A is the area swept out per activation and ν is the frequency of thermal activation. Whereas ρ and N always depend on dislocation arrangements, they may also be functions of the stress; the values of $\bar{v}$ and ν always depend on the temperature and stress and in some cases depend also on the details of arrangements of dislocations. For example in Seeger's formulation of the intersection mechanism,¹² it is assumed that N and A are independent of the stress and that the frequency of activation is given by

$$\nu = \frac{\nu_0 b}{L} e^{-U/kT} \quad (2)$$

where

$$U = U_j - \tau^* L b^2 \quad (3)$$

$$\tau^* = \tau - \tau_A \quad (4)$$

and where ν_0 is the Debye frequency, L is the mean spacing between the forest dislocations that are being intersected, U is the energy that must be supplied by a thermal fluctuation to complete intersection, U_j is the energy to form a jog, and kT has its usual meaning of the Boltzmann constant times the absolute temperature. Equation 3 reveals that the activation energy U is equal to the energy to form jogs on the intersected or intersecting dislocations minus the work, τ^*Lb^2 that the stress contributes during intersection. The stress τ^* , however, is the average local stress and, as shown by Eq. 4, is equal to the applied stress, τ , minus the athermal stress τ_A . Combining Eqs. 1b, 2, 3 and 4 gives

$$\tau = \tau_A + \frac{U_j}{Lb^2} - \frac{kT}{Lb^2} \ln \left\{ \frac{NAb^2\nu_0}{L\dot{\gamma}} \right\} \quad T \leq T_c \quad (5)$$

$$\tau = \tau_A \quad T \geq T_c \quad (6)$$

Although, as shown by Dorn and Mitchell,¹³ the activation energy for intersection is somewhat more complicated function of τ^* than that given by Eq. 3 when the dislocations are dissociated into partials, the validity of Eq. 5 has been well established for Al where the constriction energy is negligible. This is revealed in Fig. 2 by data for polyslip in Al where the tensile axis was in the [111] crystallographic direction.¹⁴ At and above T_c thermal fluctuations of energy U_j are so frequent that intersection takes place without delay even when τ^* is vanishingly small. Furthermore as shown by Eqs. 5 and 6 and the data of Fig. 2, as the strain rate is increased, T_c is also increased. Thus an exclusively athermal behavior might exist at one temperature over a range of slow strain rates

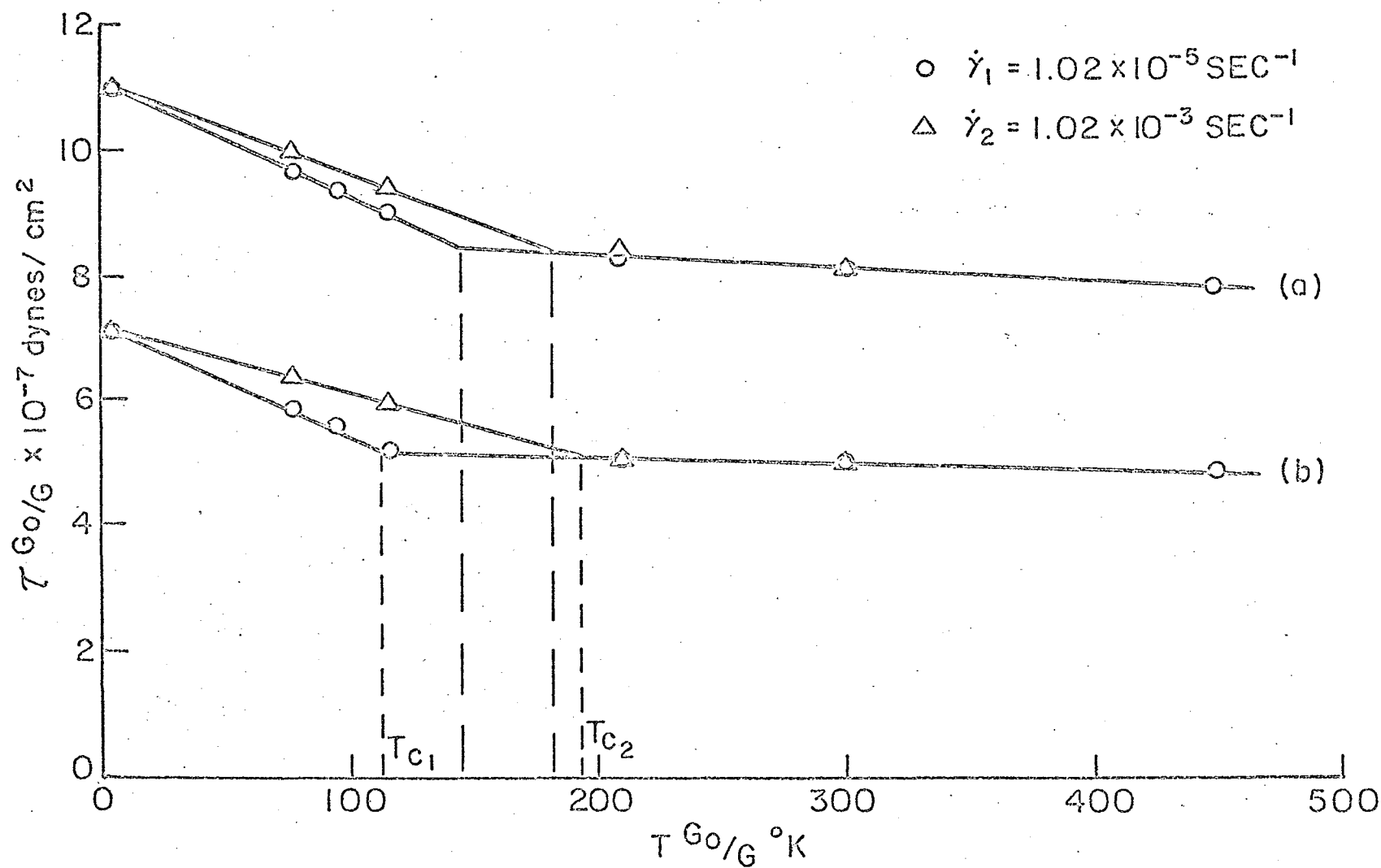


FIG. 2 EFFECT OF TEMPERATURE ON THE FLOW STRESS OF AL SINGLE CRYSTALS IN [111] ORIENTATION IN TWO PRE-STRAINED STATES. (14)

which will revert at higher strain rates to a thermally activated mechanism superimposed on the athermal behavior.

It is significant that Eqs. 5 and 6 do not contain the strain explicitly. The data of Fig. 2, however, show that τ_A increases and L decreases with strain hardening. Furthermore Mukherjee, Mote and Dorn¹⁴ have shown that $NAb \frac{\nu_{ob}}{L}$ also changes with deformation as illustrated in Fig. 3. Although the origin of these various changes in the dislocation network variables is yet under discussion, current evidence supports Saada's thesis that $\tau_A \simeq \frac{Gb}{L'}$ where $L' \simeq 2L$ mean distance between attractive junctions.⁷ Obviously the strain will enter Eqs. 5 and 6 as determined by the modification of dislocation arrangements during deformation. Before much further progress can be made in using dislocation theory for determination of plastic wave propagation effects, it will be necessary to ascertain how the values of τ_A , L and $NAb \frac{\nu_{ob}}{L}$ depend on the strain so that Eqs. 5 and 6 or their equivalent can be formulated as $\dot{\gamma} \equiv \dot{\gamma} \{ \gamma, T, \tau \}$. At present, the only recourse is the experimental evaluation of the needed relationships and these are somewhat questionable because it is usually assumed that they are independent of the stress.

Various thermally activated mechanisms of deformation are summarized in Table II.

Region III of Fig. 1 has been shown to be due to a diffusion controlled mechanism.¹⁶ The various diffusion-controlled mechanisms, however, are not enumerated here because they prevail only at temperatures above about one-half of the melting temperature and then only for slow strain rates;

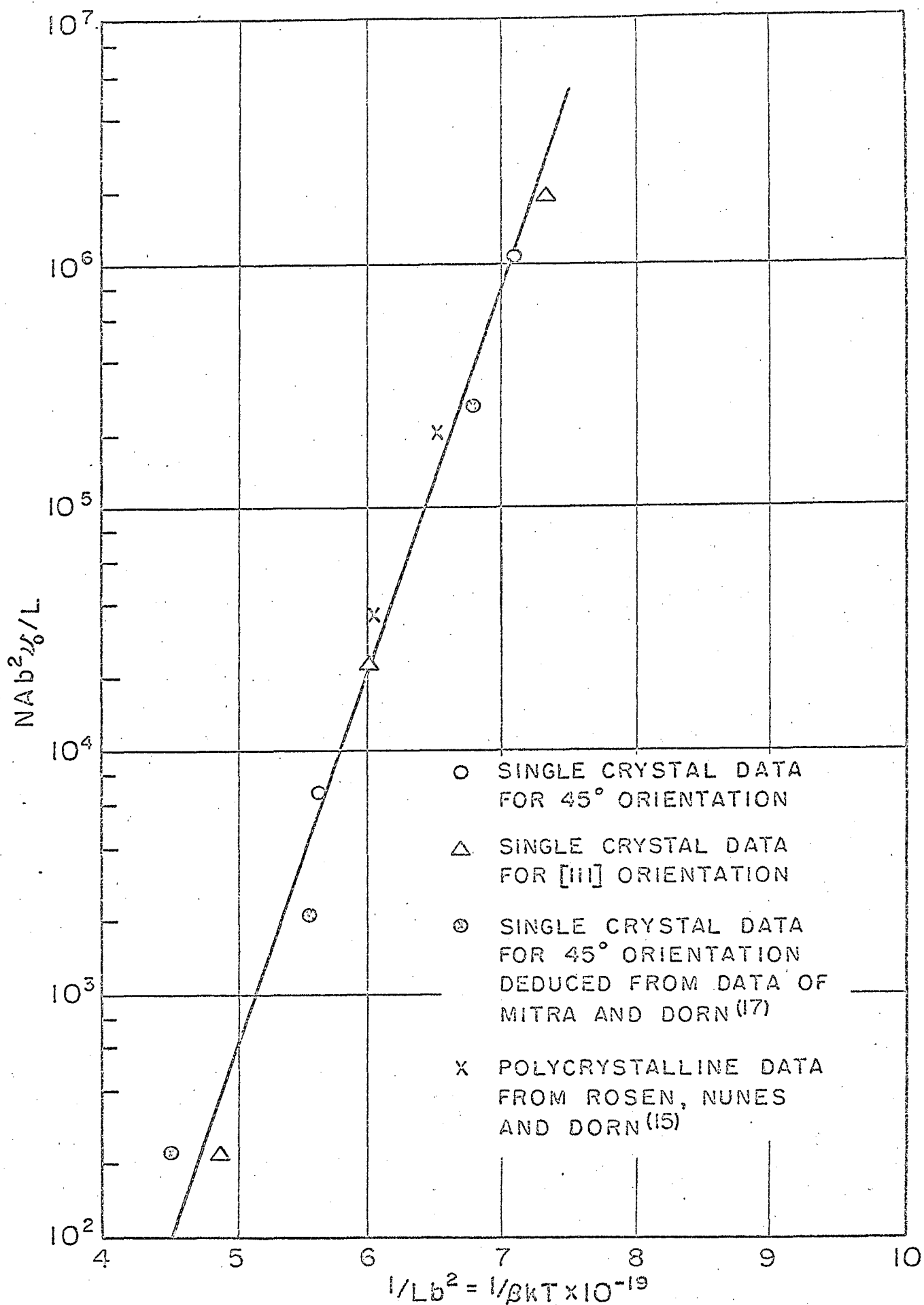


FIG.3 CHANGES IN DISLOCATION STRUCTURE VARIABLES WITH DEFORMATION

TABLE II

Thermally Activated Mechanism

Intersection¹⁷

Peierls Mechanism¹⁸

Cross-Slip¹⁹

Solute Atoms^{20,32}

Many Diffusion Controlled Mechanisms²¹

consequently, they are of little significance to the dynamic behavior of materials.

Whereas the intersection mechanism provides only modest strain-rate effects, the Peierls process is much more strain-rate sensitive. A typical example is the range over Region I of Fig. 1 where the flow stress decreases rather rapidly as the temperature increases and the strain rate decreases. This behavior is also typical for BCC metals at low temperatures in contrast to FCC metals which have very low Peierls stresses and obey the then more difficult but less temperature and strain-rate sensitive intersection mechanism over about the same low temperature range.

The mechanics of the Peierls process is illustrated in Fig. 4. Under zero stress, dislocations prefer to lie in valleys parallel to rows of atoms on the slip plane where their line energy, Γ_0 , is least. As shown by the potential energy diagram, the line energy is almost sinusoidal and has its maximum value, Γ_c , for displacements $a/2$ where a is the spacing between parallel rows of atoms. At the absolute zero a stress τ_p^* is required to push the dislocation over the Peierls hill. If a stress $\tau^* < \tau_p^*$ is applied, the dislocation will be moved partway up the Peierls hill from ABC to A'B'C' where ordinarily it will come to rest. Above the absolute zero, however, thermal fluctuations will cause loops such as A'B''C' to be formed periodically. Small loops will collapse but for sufficiently high energy fluctuations where B'' is sufficiently over the Peierls hill the applied stress τ^* will cause the loop to expand sideways and cause the dislocation to advance to A'''B'''C''' along its length. Using the line energy concept, Dorn and Rajnak¹⁸ calculated from the

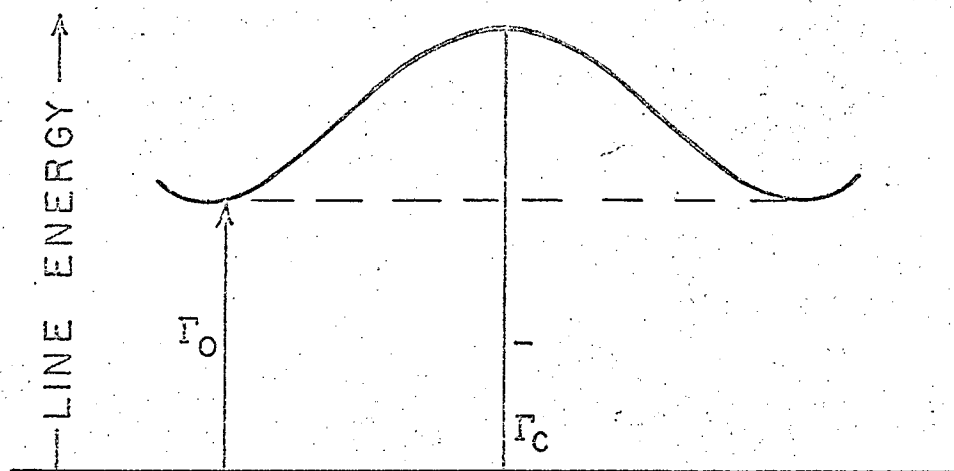
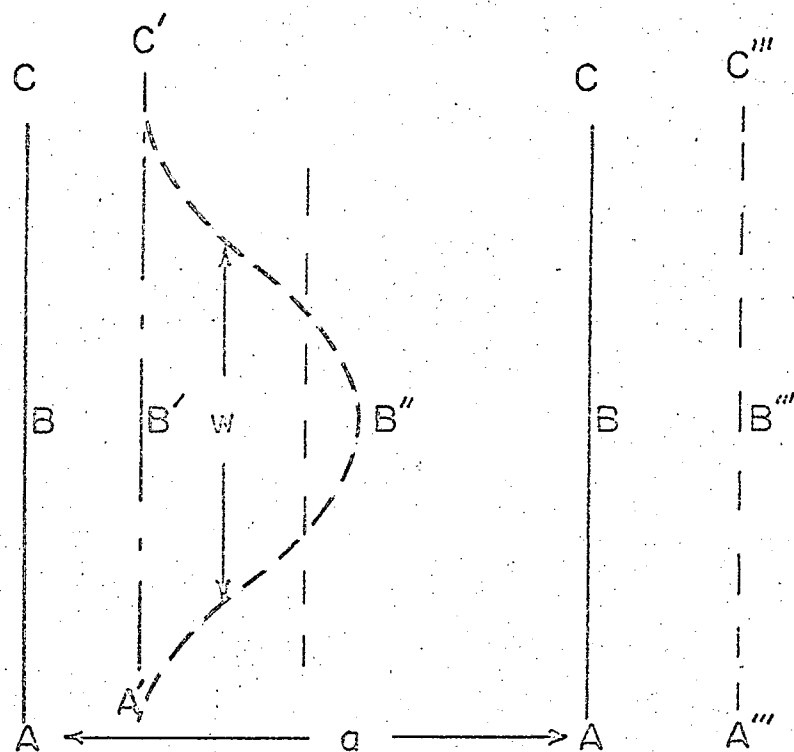


FIG. 4 NUCLEATION OF A PAIR OF DISLOCATION KINKS IN THE PEIERL'S PROCESS.

calculus of variations the critical shape loop and the extra energy required for such an event. Their results are shown in Fig. 5 where U_n is the activation energy to nucleate a pair of kinks and U_k is the energy of a single isolated kink in the unstressed metal. If w is the average separation of the pair of kinks upon nucleation and L is the distance the kinks separate before their motion is arrested, the mean velocity v of the dislocation is given by

$$v = \frac{vaL}{w} e^{-U_n/kT} \quad T \leq T_c \quad (7)$$

and the plastic shear strain rate given by Eq. 1a is

$$\dot{\gamma} = \frac{\rho baLv}{w} e^{-U_n/kT} \quad T \leq T_c \quad (8)$$

As shown in Fig. 5, $U_n = 2U_k$ when $\tau^* = 0$ and U_n decreases toward zero as τ^* approaches τ_p^* .

Whereas slightly above T_c , $\tau = \tau_A$, and the mechanism is athermal, below T_c , $\tau = \tau^* + \tau_A$, and the process consists of a thermally activated mechanism. At T_c , $\tau^* \equiv 0$, whence $U_n = 2U_k$ and

$$\dot{\gamma} = \frac{\rho baLv}{w} e^{-2U_k/kT_c} \quad (9)$$

Consequently for the same conditions of $\dot{\gamma}$ and $\frac{\rho baLv}{w}$ as shown by Eqs. 8 and 9

$$\frac{U_n}{2U_k} \approx \frac{T}{T_c} \quad (10)$$

Thus, as already documented in Fig. 5, τ^*/τ_p^* is a universal function of T/T_c for the Peierls mechanism. As shown by Eq. 9, the only effect of increasing strain rate alone is to increase T_c .

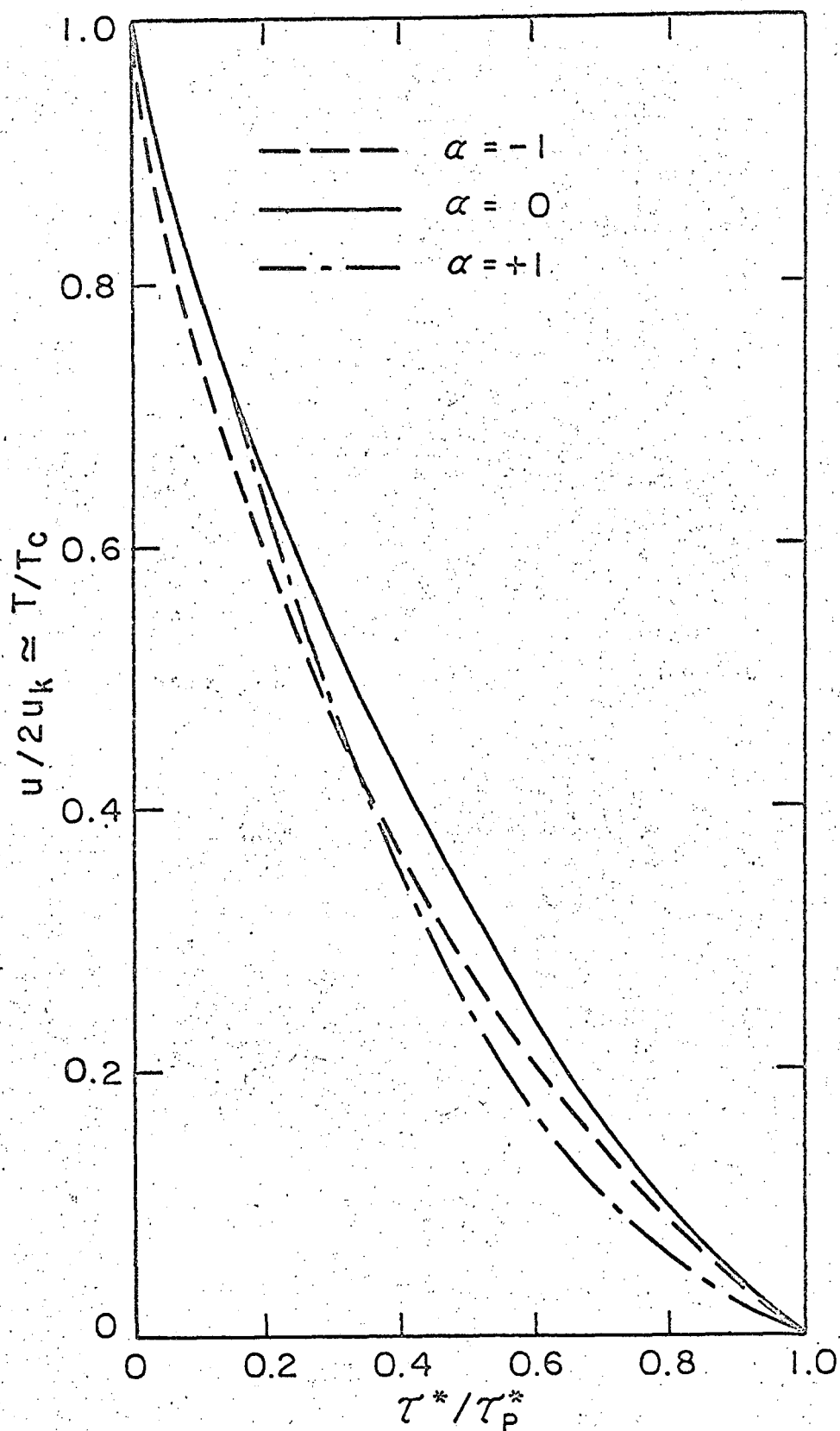


FIG.5 EFFECT OF STRESS ON THE
ACTIVATION ENERGY FOR KINK
NUCLEATION.

The data for two slow strain rates given in Fig. 1 can be used to predict the dynamic behavior. $\tau_p^* = \tau^*(\text{at } 0^\circ\text{K}) = (\tau - \tau_A)(\text{at } 0^\circ\text{K})$. The kink energy can be obtained from the values of T_{c_1} and T_{c_2} at $\dot{\gamma}_1$ and $\dot{\gamma}_2$ respectively by applying Eq. 9, according to which

$$\frac{\dot{\gamma}_1}{\dot{\gamma}_2} = \frac{e^{-2U_k/kT_{c_1}}}{e^{-2U_k/kT_{c_2}}} \quad (11)$$

Knowing in this way $2U_k$, the value of $\frac{\rho abLv}{w}$ can be determined from Eq. 9. Thus from Eq. 9, the new T_c for any high strain rate $\dot{\gamma}$ is now determinable. And thus applying Fig. 5, τ^* can be determined as a function of T the values of τ_p^* and T_c being now known. For good accuracy in prediction, it is necessary to correct for the modulus change with temperature according to

$$\begin{aligned} U_k &= U_{k_0} G/G_0 \\ U_n &= U_{n_0} G/G_0 \\ \tau_A &= \tau_{A_0} G/G_0 \end{aligned} \quad (12)$$

where the subscript zero refer to the absolute zero and G is the shear modulus at the test temperature. Figure 6 gives the slow strain rate data, shown by solid lines as determined on an Instron testing machine, the predicted dynamic yield strengths as shown by broken curves, and the dynamic test data points as obtained by the Kolsky technique⁴ for prismatic slip in Ag_2Al . The predictions are within the accuracy and reproducibility of the dynamic test data. Equally good agreement has recently been obtained for $\{123\}[111]$ slip in AgMg by Barmore.²²

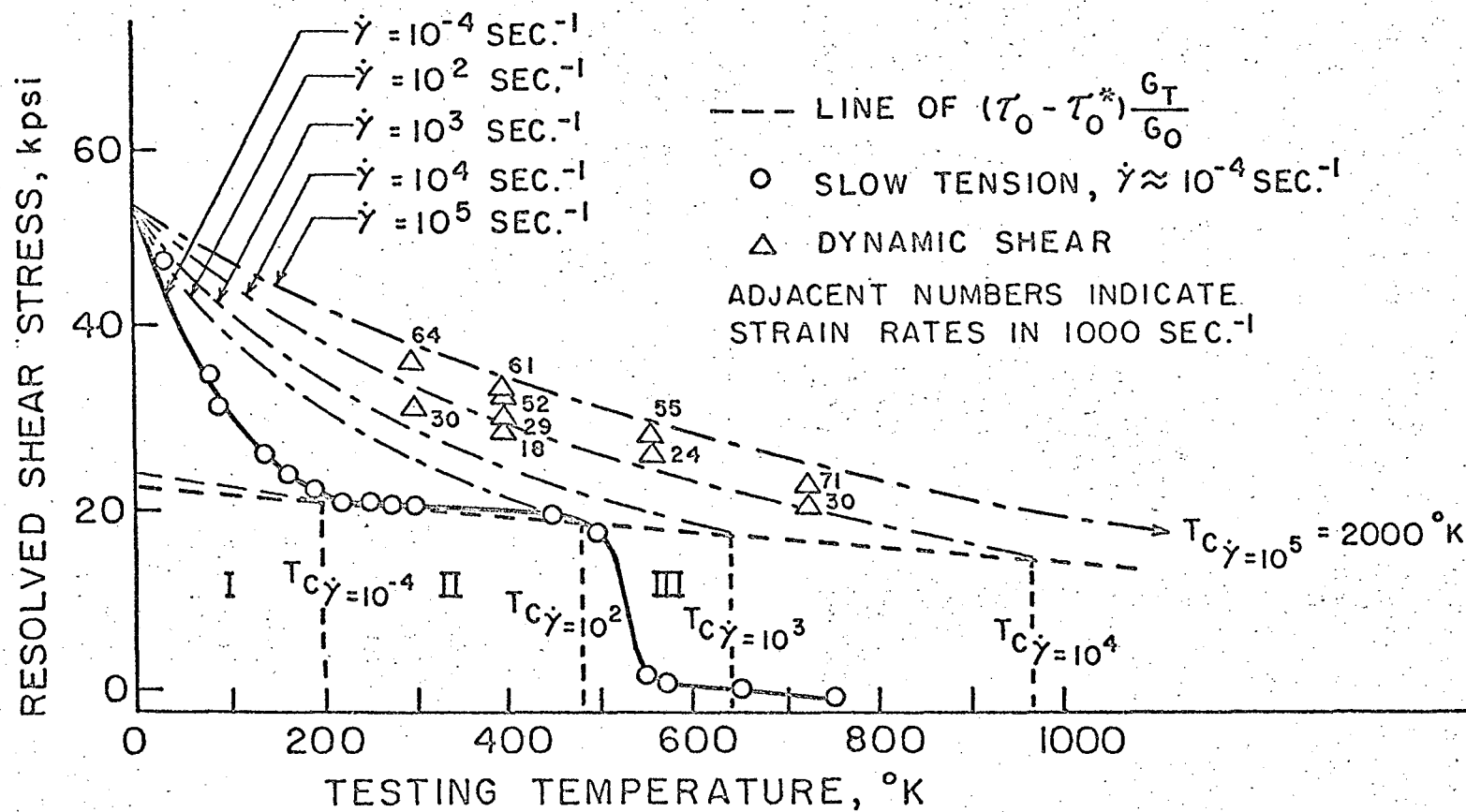


FIG. 6 RESOLVED SHEAR STRESS vs. TEMPERATURE AND STRAIN RATE FOR PRISMATIC SLIP IN $\text{Ag}_2\text{-Al}$.

The activation energy for the intersection mechanism (Eq. 3) is dependent on the mean spacing of the forest dislocations, L , and therefore increases for the same stress as L increases with straining. In contrast, as shown in Fig. 5, the activation energy for nucleation of pairs of kinks by the Peierls mechanism is physically determined and independent of dislocation patterns. Consequently only the factors ρ , L and τ_A are expected to be dependent on strain when the Peierls mechanism applies.

IV. HIGH VELOCITY DAMPING MECHANISMS

When the applied stress is less than $\tau^* + \tau_A$ (as determined at the absolute zero) dislocations are momentarily arrested at barriers and their mean velocities are determined by the frequency of thermal activation over such barriers. If, however, the applied stress exceeds $\tau^* + \tau_A$ (as measured at the absolute zero) the stress itself can push the dislocations past the barriers without delay. In this range, various energy dissipative mechanism can operate which serve to prescribe the velocity of dislocations. Unfortunately not all of these mechanism are presently known and only a few that are known have been as yet subjected to detailed theoretical scrutiny. Furthermore only relatively few experimental studies have been conducted under conditions where such high velocity damping mechanisms are significant. In Table III is enumerated an incomplete list of various damping mechanism that might serve to limit dislocation velocities.

Thermoelastic damping arises from the changes in hydrostatic stress and associated temperatures that occur at each point in a crystal as a dislocation moves through it. The inhomogeneous hydrostatic stress field of a moving dislocation sets up a temperature gradient which produces a rise in entropy and a dissipation of the mechanical energy of the system. The phonon viscosity effect results from changes in the energies of the modes of lattice vibrations due to the shear stress field of a moving dislocation. These energy changes are equalized by an energy exchange between the quantized sound waves (phonons) producing a dissipation of mechanical energy and a damping of dislocation motion. Anharmonic damping originates from changes in the vibrational energies of the atoms

TABLE III
Dislocation Damping Mechanisms

Thermoelastic ²³	$(\tau - \tau_B) \approx \frac{C_V T}{25} \times \frac{\omega}{c}$	(13)
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Phonon viscosity ²⁴	$(\tau - \tau_B) \approx \frac{C_V T}{20} \times \frac{\omega}{c}$	(14)
--------------------------------	--	------

Anharmonic radiation ²⁵	$(\tau - \tau_B) \approx \frac{\epsilon}{12} \times \frac{\omega}{c}$	(15)
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Phonon scattering ²⁶	$(\tau - \tau_B) \approx \frac{\epsilon}{8} \times \frac{\omega}{c}$	(16)
---------------------------------	--	------

where

$$\epsilon \approx \frac{3h\nu_0}{b^3(e^{-h\nu_0/kT} - 1)}$$

h = Plank's constant

ν_0 = Debye frequency

c = transverse sound velocity

in the highly distorted lattice region of the dislocation core. As a dislocation moves through the crystal, the atoms in the region of the core increase their vibrational energy and radiate this excess energy to the surrounding matrix when they return to their normal lattice positions. The amount of vibrational energy radiated depends on the velocity of the dislocation and the applied shear stress must supply this energy to maintain a uniform dislocation velocity. The phonon scattering effect arises from two distinct processes. One contribution to this damping mechanism results from the scattering of sound waves, produced by the natural vibration of the lattice, by the inhomogeneous stress field of moving dislocations. The second contribution involves the ability of a dislocation to move in a slip plane under the action of a shear stress. Sound waves that cross this plane and produce a shear stress on it will cause the dislocation to vibrate and radiate energy in the form of elastic waves.

An example of strain-rate phenomena in slightly cold-worked polycrystalline Al is recorded in Figs. 7 where the low strain rate data were obtained from tests conducted on an Instron testing machine and the high strain-rate data from tests using the Kolsky technique.²⁷ Since Al undertakes the intersection mechanism, Eq. 5 applies; the various parameters in Eq. 5 were evaluated from two slow strain rate tests in a manner analogous to that already described for the Peierls process. On this basis, the dynamic behavior over the range where the intersection mechanism controls was predicted as shown by the solid line in Fig. 7a where the yield stress increases linearly with $\ln \dot{\epsilon}$ as required by Eq. 5, where the true strain rate $\dot{\epsilon}$ is taken to be $\approx \dot{\gamma} \sqrt{3}$. The terminal point, shown by ∇ , refers to

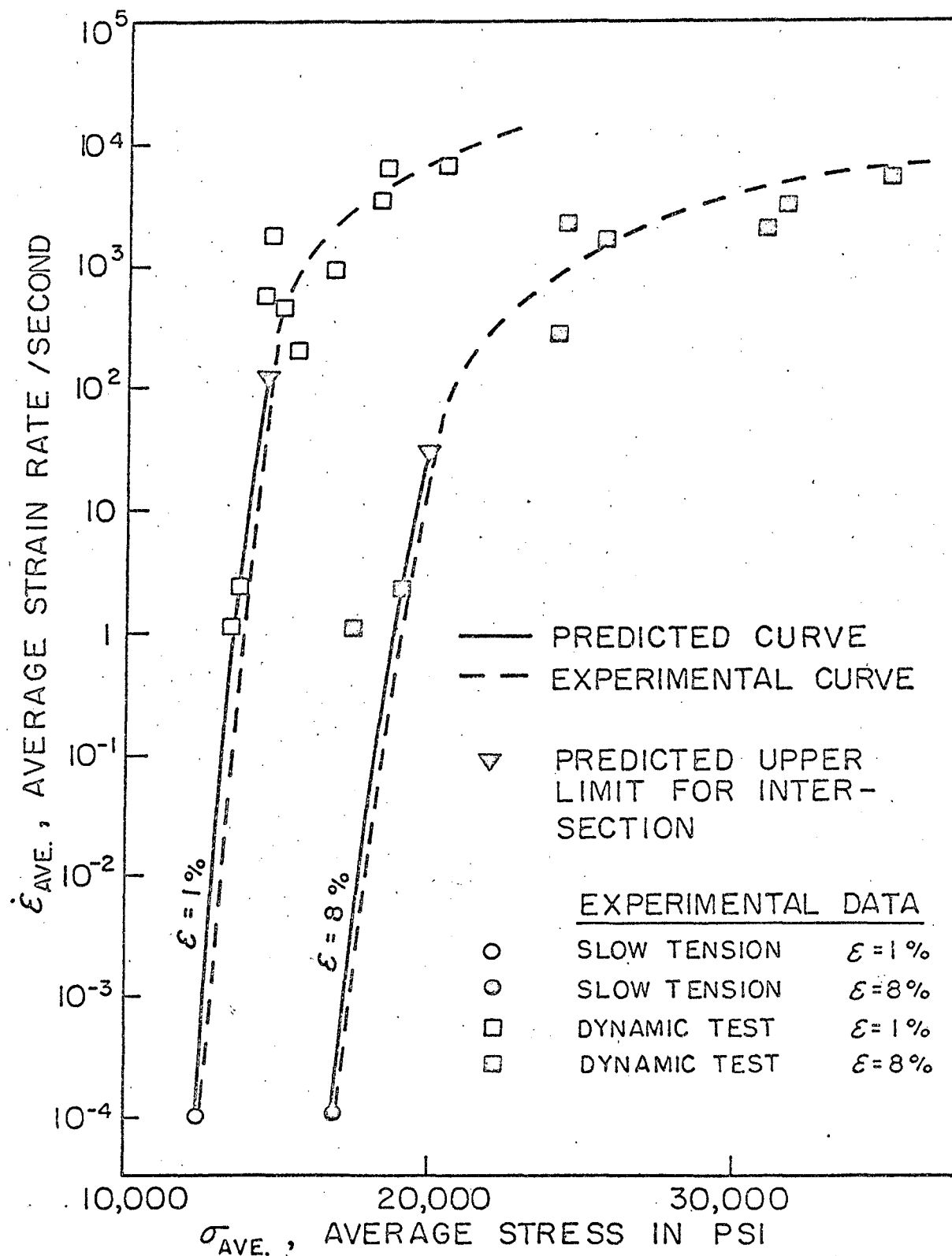


FIG. 7a PREDICTION OF THE DYNAMIC PLASTIC BEHAVIOR IN THE INTERSECTION REGION FOR PURE Al AT $T = 77^{\circ}\text{K}$.

the value obtained for $\tau = \tau^* + \tau_A$ for the slow strain rate tests as extrapolated to the absolute zero. Up to this point, Eq. 5 provides accurate predictions of the dynamic behavior of polycrystalline Al. When the stress at any temperature exceeds $(\tau^* + \tau_A)_0$ as evaluated at the absolute zero, the applied stress itself is sufficiently high to induce intersection and the velocity of the dislocation is no longer controlled by the thermally activated intersection mechanism. It is just at this point where the curve of Fig. 7a deviates from linearity suggesting that a new mechanism indeed applies for the higher range of stresses shown by the broken curve. Although it was thought at one time that the strain rate for the higher stresses was reaching some limiting strain rate, perhaps as dictated by the relativistic velocity of the dislocations, the same data replotted on linear scales, as shown in Fig. 7b, reveal that such a limiting condition was not achieved over the range of stresses that were examined. Rather the strain rate appears to increase almost linearly with some overstress $\tau - \tau_B$ in the high stress region as might be expected for some damping mechanism.

Assuming thermoelastic damping, we might expect

$$(\tau - \tau_B) \simeq \frac{C\nu T}{25} \frac{\nu}{c} \quad (17)$$

whence from Eq. 3b

$$(\tau - \tau_B) \simeq \frac{C\nu T}{25c} \frac{\dot{\gamma}}{\rho b} = \frac{C\nu T \sqrt{3} \dot{\epsilon}}{25c\rho b} \quad (18)$$

Using $\dot{\epsilon} = 4 \times 10^3/\text{sec}$, $(\tau - \tau_B) = 1.17 \times 10^9 \text{ dynes/cm}^2$, $C_\nu = \frac{2.14 \times 10^7 \text{ ergs}}{\text{cm}^3 \cdot \text{K}}$, $T = 300^\circ\text{K}$, $c = 3.2 \times 10^5 \frac{\text{cm}}{\text{sec}}$ and $b = 2.86 \times 10^{-8} \text{ cm}$, suggests that $\rho \simeq 9.6 \times 10^4$.

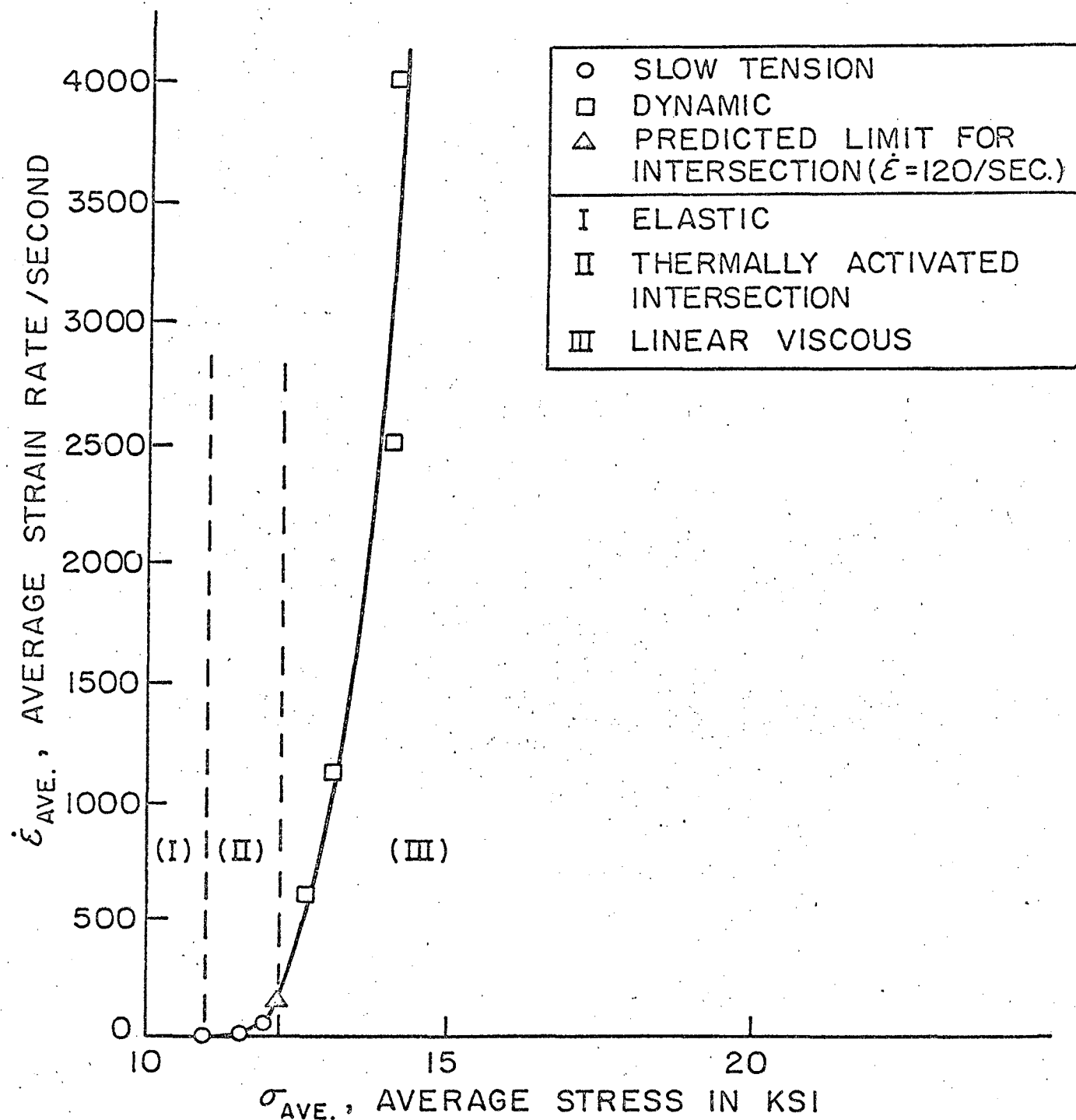


FIG. 7b EFFECT OF STRESS ON THE STRAIN RATE
IN DYNAMIC TESTS IN PURE Al AT 295°K.

However, this value of ρ is orders of magnitude below that known to be correct in the thermally activated region for much lower stresses.

Consequently, these data suggest the existence of some yet unknown but much more effective damping mechanism than any of those documented in Table III.

V. RELATIVISTIC EFFECTS

Theoretical analyses²⁸ have demonstrated that the line energy Γ of a screw dislocation depends on its velocity, v , according to the relationship

$$\Gamma = \Gamma_0 / \{1 - (v/c)^2\}^{1/2} \quad (19)$$

where Γ_0 is the rest energy and c is the velocity of a transverse sound wave. A similar but slightly more complicated expression applies for an edge dislocation. Thus as v approaches c the line energy approaches infinity and c therefore represents the maximum velocity a dislocation can achieve. Expansion of Eq. 19 into a Taylor's series reveals that the effective mass of a dislocation when $v < c$ is given by the very small quantity Γ_0/c^2 . Thus dislocations can accelerate in less than about 10^{-11} sec to extremely high velocities for modest overstresses. Such velocities might of course be limited by viscous damping at high stresses or by thermally activated mechanisms at lower stresses.

Additional theoretical investigations have uncovered interesting modifications of the stress fields about high velocity dislocations which on occasions serve to change drastically their interaction energies and therefore their behavior from those expected for more leisurely moving dislocations.²⁹ Unfortunately, as will be discussed more fully later, there has not yet been reported any experimental data to confirm these deductions.

VI. VELOCITIES OF DISLOCATIONS

The plastic strain rate which must be known in order to formulate mathematical descriptions of plastic wave phenomenon depends on two factors (vide Eq. 1b) namely the density of mobile dislocations and their mean velocity. Each factor may depend on the stress-strain and temperature history to which the material was subjected. More realistic theories of plastic wave propagation therefore depend on more detailed knowledge about the factors that affect ρ and v . Whereas little progress has yet been made in studying the density of mobile dislocations under dynamic stressing, a number of experimental investigations have been directed toward obtaining v by the Johnston-Gilman technique.³⁰ Mobile dislocations are introduced by several different methods and their initial position and final position following stress pulsing are determined by etch pitting. The velocity is determined by dividing the displacement by the duration of the pulse. Johnston and Gilman have shown that the displacement in one long stress pulse is about the same as that in a series of shorter pulses of equal magnitude and total duration, thus confirming deductions based on the relativistic motion of dislocations that their effective mass at velocities much less than that of sound is very small.

In spite of the simplicity of the technique many auxiliary factors are present that serve to qualify the accuracy of the results.

1. Velocities scatter widely; part of this scatter is known to arise from stress fields of other nearby dislocations which depends upon the method of introducing mobile dislocations; other factors arising from localized volume and surface effects have also been noted.

2. The stress pulse particularly for the higher stresses and shorter times deviates seriously from the ideal rectangular shape.

3. Whereas some investigators report mean velocities others report maximum velocities under conditions where perhaps neither is justified.

Despite the numerous difficulties some interesting results have been obtained:

1. Most investigators find that the dislocation velocity becomes vanishingly small near a limiting lower stress value which correlates with the yield stress. We suggest that this might be an athermal stress τ_A and that the velocity of dislocations should then be correlated with $\tau^* = \tau - \tau_A$ as previously discussed rather than with τ as is now commonly done.

2. Usually dislocation velocities at least for the lower stress range, depend on the temperature in a way that suggests the operation of thermally activated mechanisms. The data of Johnston and Gilman on LiF³⁰ (Fig. 8) and those of Gutmanus, Nadgornyi and Stepanov³¹ (Fig. 9) on NaCl are too meagre to permit rationalization of the possible mechanism. It appears, however, that the thermally activated motion of dislocation in LiF is associated with surmounting stress fields of impurities.³²

3. Erickson³³ has shown that much higher stresses are required to move dislocations on the {112} plane in silicon iron than on the {110} plane as recorded by Stein and Low³⁴ thus suggesting that the Peierls mechanism might be operative. On the other hand, Stein and Low's data are seriously at variance with expectations of the simple Peierls process described in Section III of this survey. In fact (vide Fig. 9 of Ref. 34)

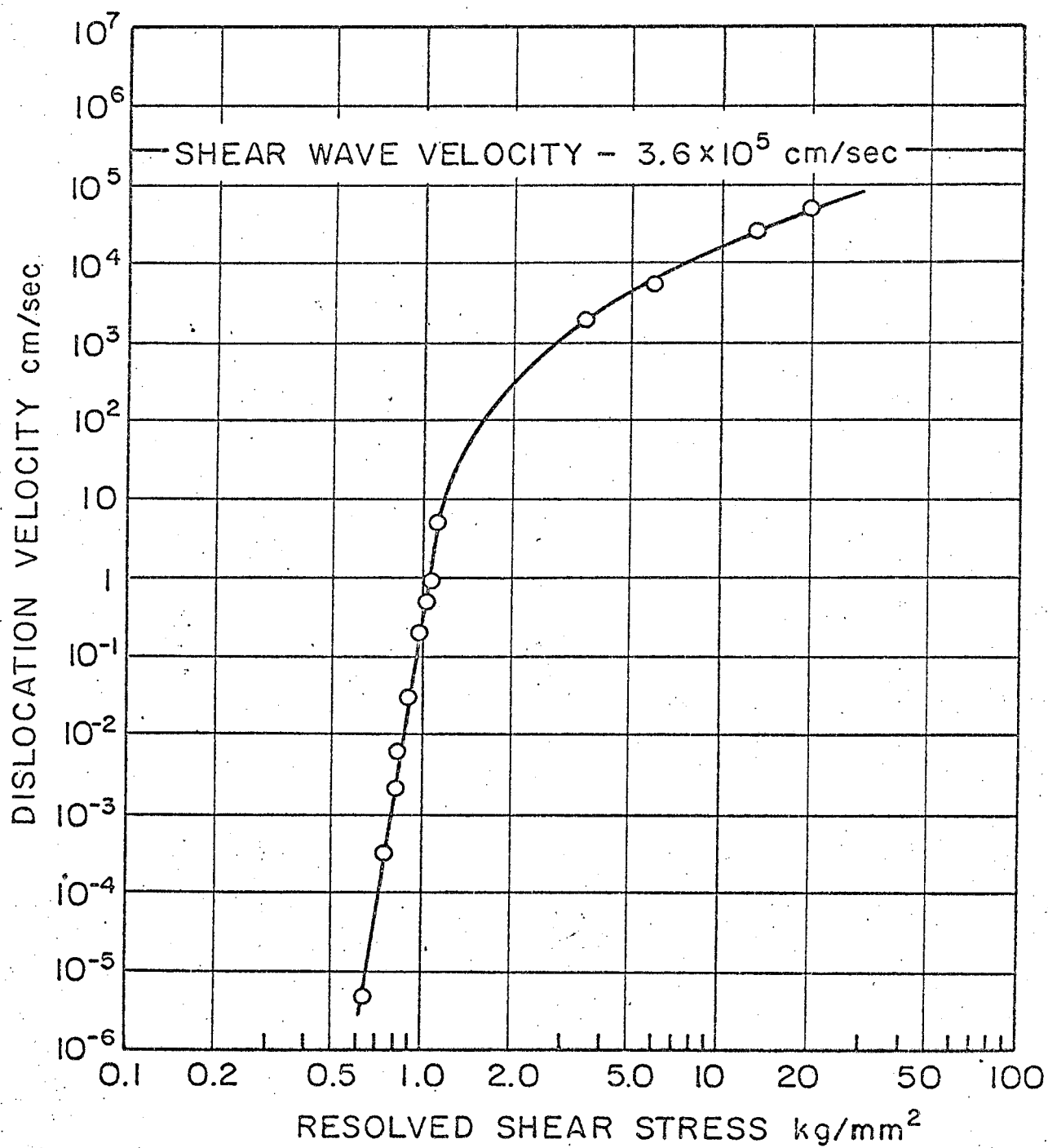


FIG. 8a EDGE DISLOCATION VELOCITY vs.
RESOLVED SHEAR STRESS FOR LiF. (30)

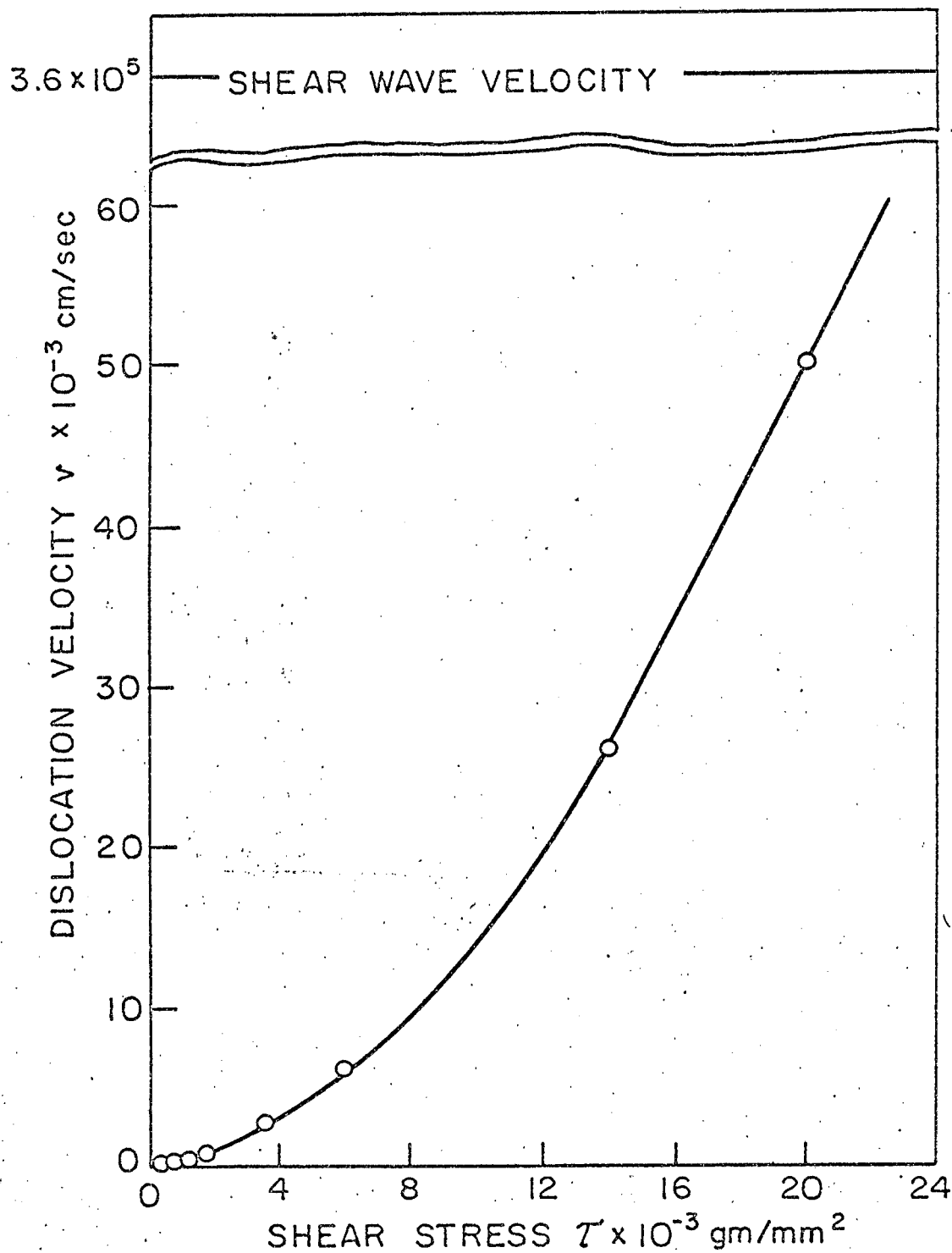


FIG. 8b EDGE DISLOCATION VELOCITY vs.
RESOLVED SHEAR STRESS FOR LiF. (30)

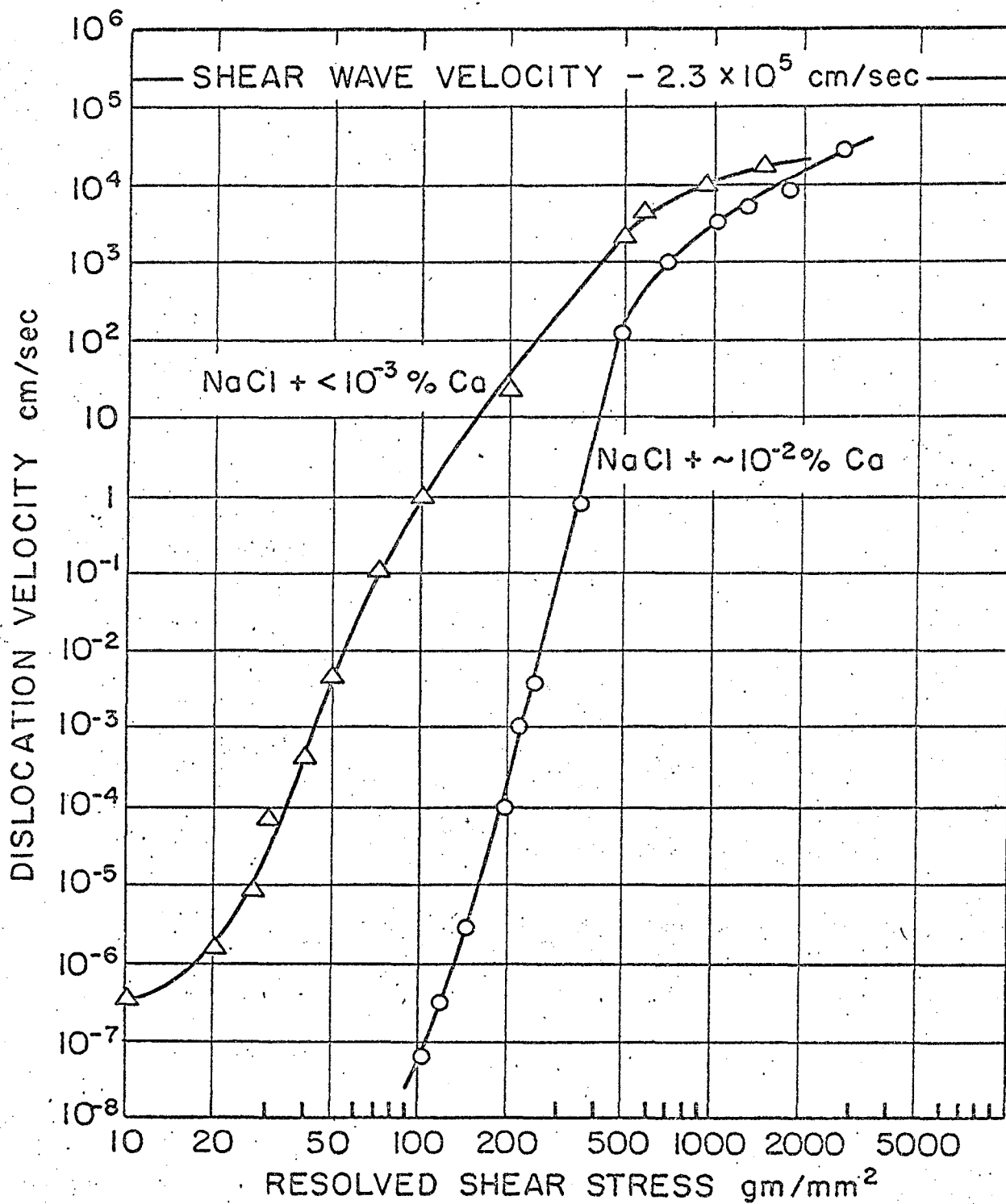


FIG. 9a EDGE DISLOCATION VELOCITY vs. RESOLVED SHEAR STRESS FOR SODIUM CHLORIDE. (31)

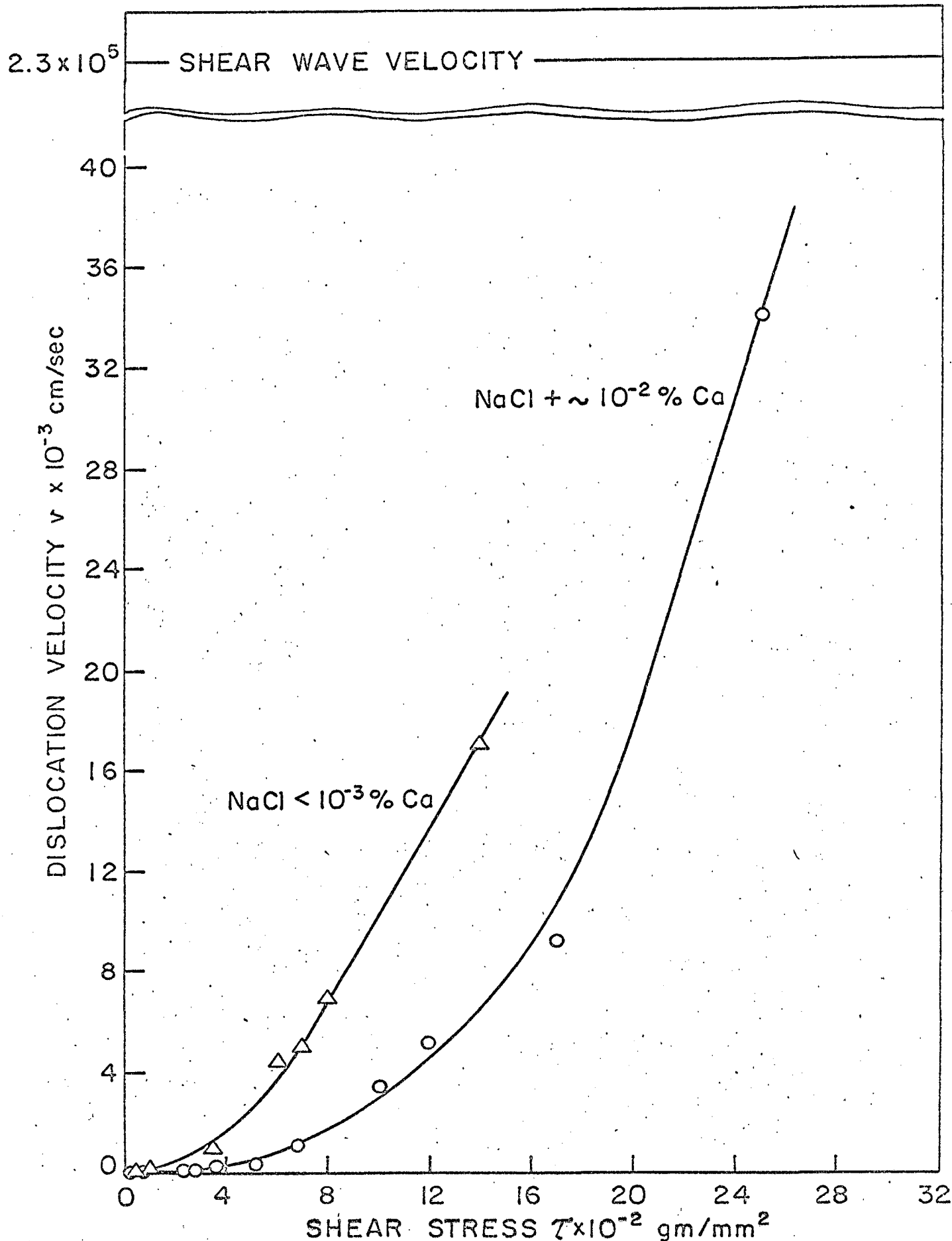


FIG. 9b EDGE DISLOCATION VELOCITY vs. RESOLVED SHEAR STRESS FOR SODIUM CHLORIDE. (31)

Stein and Low's data suggest an apparent activation energy for the higher constant stresses that decreases as the temperature decreases, a trend that is wholly inconsistent with any series of thermally activated dislocation mechanism.

4. Dislocation velocities have been determined in Ge, Si, InSb and GaSb,³⁵ all of which indicate similar trends with stress and temperature. The data of Chandhuri, Patel and Rubin³⁵ on Ge give

$$v = 2.24 \times 10^{-4} \tau^{1.5} e^{-\frac{36,600}{kT}} \text{ cm/sec}$$

over a wide range of stress and temperature when τ is in dynes/cm².

Somewhat similar results have been reported by Kabler.³⁶ This trend, however, does not correspond to any known dislocation mechanism. Celli, Kabler, Ninomiya and Thomson³⁷ have suggested that a Peierls process in which the nucleation and motion of kinks is modified by closely spaced barriers to the kink motion might be the responsible mechanism.

5. Data such as those reported in Figs. 8a and 9a have been interpreted to mean that the dislocation velocity is approaching the limiting relativistic velocity c at high stresses. When, however, these data are replotted on a linear scale as shown in Figs. 8b and 9b it becomes clear that no such limiting velocity is being approached. In fact no dislocation velocity above about $v \approx c/2$ has yet been measured.

6. Johnston and Gilman's data³⁰ reveal that below a stress of about $3.9 \times 10^8 \frac{\text{dynes}}{\text{cm}^2}$ (the maximum used over a range of temperatures) LiF undertakes a thermally activated mechanism. As shown in Fig. 8b, below this stress the dislocation velocity increases somewhat exponentially with

the stress whereas above this stress level the increase in velocity is, within experimental accuracy, linear with an increase in stress. The same general trends are observed for NaCl (Fig. 9b). It is therefore reasonable to suspect that the linear region is due to some damping mechanism.

Calculations of damping coefficients B for several of the mechanisms listed in Table III have been made by Mason²⁴ and Suzuki, Ikushima and Aoki³⁸ where b is defined by

$$F = (\tau - \tau_B)b = B\nu \quad (20)$$

These theoretical values are compared with the values of B determined from the dislocation velocity measurement in LiF and NaCl in Table IV. Also shown in this table are values of B for several other materials determined from the Granato Lücke theory³⁹ of damping of high frequency acoustical waves. As shown in Table IV, the theoretical B values are in satisfactory agreement, within an order of magnitude, suggesting that one or more of these mechanisms might be responsible for the linear velocity-stress behavior at high stresses.

TABLE IV
Dislocation Damping Coefficients
(B of Eq. 20)

Material	LiF	NaCl	KCl	Quartz	Cu	Cu-0.13%Mn
OBSERVED VALUES						
Dislocation velocity measurements	$7 \times 10^{-4}(30)$	$2 \times 10^{-4}(31)$				
Accoustical attenuation measurements	$1.3 \times 10^{-3}(38)$	$2 \times 10^{-3}(40)$	$3.5 \times 10^{-4}(38)$	$6 \times 10^{-4}(24)$	$6 \times 10^{-5}(24)$ $7.9 \times 10^{-5}(38)$ $8 \times 10^{-4}(41)$ $6.5 \times 10^{-4}(42)$	$4.5 \times 10^{-4}(38)$
THEORETICAL VALUES						
Thermoelastic	$1.7 \times 10^{-4}(24)$	$7.1 \times 10^{-5}(24)$ $7.7 \times 10^{-5}(38)$		$2.5 \times 10^{-5}(24)$	$1 \times 10^{-5}(24)$	
Phonon viscosity	$9 \times 10^{-4}(24)$ $4.5 \times 10^{-5}(38)$	$2.1 \times 10^{-3}(24)$ $1.1 \times 10^{-4}(38)$	$1.9 \times 10^{-4}(38)$	$6 \times 10^{-4}(24)$	$5 \times 10^{-5}(24)$ $7 \times 10^{-5}(38)$	$1 \times 10^{-4}(38)$
Phonon scattering	$5.6 \times 10^{-5}(24, 38)$	$7.7 \times 10^{-5}(24)$	$5.9 \times 10^{-5}(38)$	$4 \times 10^{-5}(24)$	$1 \times 10^{-4}(24)$ $1.1 \times 10^{-4}(38)$	$9 \times 10^{-5}(38)$

SUMMARY

Substantial progress has already been made in rationalizing the dynamic behavior of crystalline materials in terms of dislocation theory. Many athermal and thermally activated dislocation mechanisms have been identified and characterized. Major emphasis need yet be placed on a better understanding of strain hardening; in addition new types of thermally activated mechanisms must be rationalized and the effect of the simultaneous operation of several thermally activated mechanisms must yet be better formulated. Our theoretical understanding of viscous damping effects on high speed dislocations is yet very immature and only a few experiments have been performed in this range. Whereas the LiF and NaCl data suggest relations that might arise from the damping mechanisms given in Table III, those for Al appear to require the operation of some much more effective mechanism. Although it has been suggested in the past that dislocations in LiF, NaCl and Al might be approaching their terminal relativistic velocities under high stress, a re-evaluation of these data prove that this is not so; the trends of a linear dependence of dislocation velocity with stress in the high stress region is suggestive of the operation of damping mechanisms. Up to the present, there does not appear to be any experimental data at sufficiently high dislocation velocities to reveal their relativistic behavior.

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