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DISLOCATION CONCEPTS OF STRAIN RATE EFFECTS

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ABSTRACT

Whereas von Karman and Taylor independently based their analyses for plastic wave propagation on the deformation concept which assumes the deformation stress is exclusively a function of the plastic strain, Malvern believed that a viscoelastic type of behavior is more realistic, assuming the flow stress is a function of the strain rate as well as the strain. With relatively few exceptions, current investigators of plastic wave propagation phenomena are rather sharply divided into two camps, those that insist on the von Karman-Taylor type of formulation and those that adhere to the Malvern approach. It is the thesis of this paper that both concepts are substantially correct under appropriate circumstances. This thesis will be upheld not only in terms of new experimental evidence but also in terms of deductions arrived at from elementary dislocation theory.

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I. INTRODUCTION

The major problem that is encountered in formulating a realistic mathematical theory of plastic wave propagation concerns the dynamic behavior of materials in the plastic regime. Until recently, the approach to an understanding of the dynamic plastic behavior has been exclusively empirical and experimental.[1] Whereas one group of investigators have continued to insist, in conformity with the von Karman[2] and Taylor[3] assumptions, that the dynamic stress-strain curves are insensitive to the strain rate, other investigators have been strongly convinced that strain-rate effects are significant. In fact a dicotomy of viewpoint has developed between two groups of extremists even though the experimental conditions leading to the two sets of empirical deductions were generally somewhat different. In certain cases, however, the differences in the conditions of test and the materials that were tested appeared to be small.

It is the objective of this summary to reveal that both viewpoints are substantially correct under appropriately different conditions; and in some instances the difference in conditions of the test, differentiating between strain-rate sensitive and strain-rate insensitive results, may

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indeed appear to be small unless subjected to deeper scrutiny. Our interest in this paper will be limited to crystalline materials, particularly metals; therefore, we shall categorically insist, without further proof, that various permissible dislocation mechanisms are responsible for the plastic deformation. When these mechanisms are thermally activated, the stress-plastic strain behavior will be strain-rate sensitive. Under remaining athermal processes, the stress-plastic strain curve will be strain-rate insensitive.

II. SIGNIFICANCE OF THE STRESS-PLASTIC STRAIN BEHAVIOR

The only difference between various mathematical theories of plastic wave propagation concerns the assumptions that are made regarding the plastic behavior of the material.[4] In order to illustrate this issue we consider the simplified problem of a plastic wave moving along a thin cylindrical rod using, as first formulated by von Karman, Lagrangian coordinates. The two equations that satisfy the conditions of continuity and the conservation of momentum, which are common to all possible theories, are

$$\frac{\partial \varepsilon}{\partial t} = \frac{\partial v}{\partial \alpha} \tag{1}$$

$$\frac{\partial \sigma}{\partial \alpha} = \rho \frac{\partial v}{\partial t} \tag{2}$$

where ε = the total axial engineering strain

t = the time

v = the particle velocity

α = the Lagrangian positional coordinate

σ = the engineering stress

ρ = the density of the material.

The additional equation needed to solve for the three dependent variables of σ, ε, and v is a realistic formulation of the plastic behavior of the material. Unfortunately it is not apparent á priori what type of general formulation might be needed. For purposes of discussion, however, we will consider three limiting cases where the time enters the analysis in three different orders of magnitude, namely

$$d\dot{\varepsilon} = p\{\sigma, \varepsilon, \dot{\varepsilon}\} d\sigma + q\{\sigma, \varepsilon, \dot{\varepsilon}\} d\varepsilon + r\{\sigma, \varepsilon, \dot{\varepsilon}\} dt \quad (3a)$$

$$d\varepsilon = \frac{1}{E} d\sigma + g\{\sigma, \varepsilon\} dt \quad (3b)$$

$$\sigma \equiv \sigma\{\varepsilon\} \quad \text{or} \quad d\varepsilon = f(\sigma)\, d\sigma \quad (3c)$$

where E is Young's modulus and the test is done at constant temperature.

Eq. 3a refers to situations where the change in strain rate is significant. An example of such a case concerns the propagation of a shock in stress, which is above the dynamic yield strength, along an originally unloaded bar. Dislocations in the unloaded section of the bar are stationary and, therefore, have zero velocity also at the first instant the shock in stress acts on them. But immediately they begin to accelerate. If they are otherwise unimpeded they will soon acquire the velocity of sound leading to extremely high strain rates.[5] Because of the very small inertia of dislocations the acceleration period will occur over about 10^{-10} sec. Consequently such accelerative phenomenon leading to changes in strain rate will not be detected in experiments where the time is not measured in shorter intervals than 10^{-10} sec. Since no experiments have yet been made in this regime, we will not dwell longer on Eq. 3a.

Eq. 3b is a generalization of Malvern's[6] viscoelastic relationship. Whereas the first term to the right of the equality is the increment in elastic strain, the second term gives the increment in plastic strain, the function $g\{\sigma, \varepsilon\}$ being the plastic strain rate. This formulation of the plastic behavior is substantially consistent with dislocation theory under conditions where the accelerative phenomena can be neglected. When, for example, an increment of stress $d\sigma$ is applied in dt = 0, only an increment in elastic strain results. But as will be shown later, $d\varepsilon$ is not an exact differential and the strain-rate function $g\{\sigma, \varepsilon\}$ will depend on the strain rate history, the physically appropriate variable for g being the substructure in lieu of the strain ε. As shown in Fig. 1, if a bar is held for a long time at σ, and then is subjected to an increment of stress $d\sigma$, the path of deformation will be an instantaneous elastic strain from A to B followed by a time dependent straining from B to C. The curve OAC, therefore, refers to the stress-strain curve conducted at an infinitely slow strain rate.

Eq. 3c assumes that the stress is exclusively a function of the strain as originally assumed by von Karman and Taylor. It demands that when an increment of stress $d\sigma$ is applied, the stress-strain curve is traced directly from A to C, as shown in Fig. 1. This assumption, as we have seen, is basically inconsistent with dislocation theory since the inertia of the dislocations demands that the true path go from A to B to C. Under certain very stringent conditions, however, the plastic behavior might nevertheless be approximated by such a strain-rate independent phenomenon. If, for example, the plastic strain-rate is so rapid that the plastic straining from B to C takes place over a shorter time interval than the least count of time possible with the experimental conditions employed in the test, the plastic behavior will appear to be strain-rate insensitive.

III. STRAIN-RATE EFFECTS FOR THERMALLY ACTIVATED MECHANISMS

Dislocation mechanisms of deformation can be classified into two major groups, those that are thermally activated and those that are athermally activated. When the energy of the unit mechanism is less than about 50kT the process may be thermally activated. But when the energy required to complete the process is greater than about 50kT, successful thermal fluctuations to aid the process are so infrequent that the process can be made to occur only by mechanical means. The flow stress for thermally activated processes decreases rather rapidly with an increase in temperature, whereas in most cases the flow stress for athermal processes decreases slowly with an increase in temperature in a manner proportional to the decrease in the shear modulus of elasticity with an increase in temperature. Furthermore, thermally activated processes always give strain-rate sensitive flow stresses.

Straining results from the slip displacement of dislocation. If the energy required to produce this displacement is high, the process is not thermally activatable. There are a number of such athermal processes, typical examples of which are:

1) Motion of dislocations through long-range back stress fields due to other dislocations,[7]
2) Motion dislocations through very high short-range back stress fields due to other dislocations,[8]
3) Motion of dislocations through alloys having short-range order,[9,10]
4) Unlocking of Suzuki-locked dislocations.[11]

Other mechanisms have much lower energies and are therefore thermally activatable; typical examples are:

1) Peierl's mechanisms of the forward motion of a dislocation from one energy well to the next,[12-14]
2) The intersection of two dislocations,[15-17]
3) Cross-slip in face-centered cubic metals,[18]
4) Motion of jogged screw dislocations leading to formation and diffusion of vacancies,[15, 19-21]
5) Dynamic recovery, resulting from the relief of long-range back stresses due to vacancy formation and diffusion, leading to the climb of edge dislocations.[15, 19, 22-24]

In general, the order given in this partial listing of thermally activated mechanisms is that of increasing activation energies. The last two processes involving the formation and migration of vacancies are the most difficult mechanisms having the highest activation energies. They can only take place at sufficiently high temperatures under conditions of slow deformation or creep, and therefore are not significant in the usual realm of dynamic conditions. But the remaining processes can be strain-rate controlling under certain appropriate dynamic conditions.

IV. THE INTERSECTION MODEL

In order to illustrate the origin of strain-rate sensitive deformation when thermally activated mechanisms control, we will consider the case of the intersection mechanism. This choice is based on the fact that the intersection mechanism has been explored more thoroughly than any other mechanism and on the fact that there are good data available to correlate creep, slow tension, and dynamic test data. Due to limitations in time and space, however, we will omit a detailed description of the factors involved here and refer the reader to some of the original discussions on this mechanism.

Seeger has shown that the plastic shear strain rate, $\dot{\gamma}$, for the intersection mechanism is given by

$$\dot{\gamma} = NAb\nu e^{-\mathcal{U}/kT} \tag{4}$$

where N = the number of points per unit volume of contact between the forest dislocations that are being cut and the glide dislocations

A = the average area swept out per intersection

b = the Burgers vector

ν ≃ the Debye frequency

k = the Boltzmann constant

T = the absolute temperature

$\mathcal{U}$ = the average energy that must be supplied by a thermal fluctuation to complete intersection.

When the dislocations are dissociated into Shockley partials, the average force, F_o, vs displacement, x, diagram at the absolute zero for intersection is given by the solid intersecting curve of Fig. 2.[25] As the dislocations are brought together, x decreases and F_o increases. As x decreases the dislocations first constrict, the constriction being completed at x = b. Between x = b to x = 0, jogs are produced.

At temperatures above the absolute zero, the force is given by

$$F = F_o\, G/G_o \tag{5}$$

where G and G_o are the shear moduli at the test temperature and the absolute zero, respectively. Eq. 5 merely reflects the fact that both the constriction energy and the jog energy are linearly dependent on the shear modulus of elasticity.

If L is the mean spacing of the forest dislocations, a force

$$F = (\tau - \tau^*) L b \tag{6}$$

will act at the point of intersection where τ is the externally applied shear stress, and τ^* is the back stress due to local interactions and long-range stress fields. For our present objectives it will not be necessary to separate these two effects. If a force F less than F_m is applied, the energy that must be supplied by a thermal fluctuation in order to complete intersection is given by the modified Basinski equation

$$U = \int_{(\tau - \tau^*)Lb}^{F_m} x \, dF \tag{7}$$

Eqs. 4 to 7 plus the data of the F_o - x curve given in Fig. 2 then completely describe the intersection mechanism. It is significant that none of these equations contain explicitly the plastic shear strain, γ^ℓ. This arises because the shear strain itself is never a physically significant variable. Rather the physically important variables, at least for the intersection mechanisms are the structurally significant quantities of $\tau^* = \tau_o^* G/G_o$, and L. In general these are not simply related to the strain but are also dependent on the temperature strain-rate and history of straining. As deduced from slow tensile tests for $\dot{\gamma} = 10^{-4}$/sec and T = 90°K, they are given in Fig. 3.[25] When the metal is strained, L decreases and τ^* increases, both factors contributing to strain hardening.

V. SEEGER'S APPROXIMATION

It will prove instructive to first formulate the intersection mechanism in terms of Seeger's approximation which neglects the effects of constriction and is therefore appropriate for intersection of undissociated dislocations. Under this simplification

$$U = U_j - (\tau - \tau^*) L b^2 = U_{jo} G/G_o - (\tau - \tau_o^* G/G_o) L b^2 \tag{8}$$

the subscripts zero are referring to the values at the absolute zero. This arises because the F - x diagram no longer contains the curved part of Fig. 2 due to the constriction force and is then represented by a rectangular area from $0 \leq F \leq F_m$ and $0 \leq x \leq b$. Introducing Eq. 8 into Eq. 4 gives

$$\tau G_o/G = \tau_o^* + \frac{U_{jo}}{Lb^2} - \frac{kTG_o}{Lb^2G} \ln \frac{NAb\nu}{\dot{\gamma}} \quad \text{for } T < T_c \tag{9}$$

and

$$\tau G_o/G = \tau_o^* \quad \text{for } T > T_c \tag{10}$$

where

$$U_{jo} = kT_c \ln \frac{NAb\nu}{\dot{\gamma}} \tag{11}$$

The critical temperature, T_c, is that temperature at which the necessary thermal fluctuations needed to assist the applied stress in order to complete intersection occur practically instantaneously. Below that temperature, as shown by Eq. 9, the flow stress decreases linearly with an increase in the absolute temperature. These trends are shown in Fig. 4 for the case of a mildly strain hardened metal where, as given in Fig. 3, 1/L was selected to be $7.3 \times 10^{+5}$ 1/cm and τ_o^* was taken as 4.8×10^8 dynes/cm^2. For a given $\dot{\gamma}$, T_c remains insensitive to all accessible strain hardened states. Therefore, $NAb\nu$ is substantially a constant. In order to complete the approximation, $NAb\nu$ was estimated to be about 120 1/sec and U_{jo} was approximated to be about 1.45×10^{-13} dyne-cm from Fig. 2 by neglecting the constriction energy. For $0 \le \tau \frac{G_o}{G} \le \tau_o^*$ the behavior is entirely elastic. And over the range $\tau_o^* \le \tau \frac{G_o}{G} \le \frac{U_{jo}}{Lb^2} + \tau_o^*$ plastic deformation occurs by thermally assisted intersection, the flow stress being dependent on the strain rate. As shown by Eq. 9, the higher the test temperature, the more rapidly does the flow stress in this range increase with an increase in strain rate. Therefore, tests conducted at low temperatures on only mildly strain-hardened metals may have flow stresses that are substantially insensitive to the strain rate, whereas the same metal tested in a severely strain-hardened state and at higher temperatures may exhibit significantly sensitive variations of flow stress with strain rate.

An example of the effect of straining on increasing the strain-rate sensitivity of dynamically tested high purity Al is shown in Fig. 5.[26] The log $\dot{\varepsilon}$ vs σ curves, which were obtained using a modification of Kolskey[27] thin wafer technique, are less steep for the material at higher strain-hardened states over the lower ranges of $\dot{\varepsilon}$ where $\ln \dot{\varepsilon}$ is linear with σ, as dictated by Eq. 9.

When $\tau \frac{G_o}{G} > \frac{U_{jo}}{Lb^2} + \tau_o^*$, thermal activation is no longer required to assist intersection because the force F at the point of intersection now exceeds F_m. Consequently the applied stress alone is more than sufficient to effect intersection. Obviously the intersection mechanism is no longer rate controlling and some other dislocation mechanism now becomes significant in dictating the dynamic plastic behavior of the metal.

VI. PREDICTIONS OF DYNAMIC BEHAVIOR CONTROLLED BY INTERSECTION

In this section the dynamic behavior of Al will be predicted up to the limit of the intersection mechanism from the data recorded in Figs. 2 and 3 and the intersection theory given in Section 4. Since these predictions follow directly from the previous discussion the details need not be given here except to state that the assumption that the material obeyed a mechanical equation of state was made. The experimentally determined dynamic behavior and the predicted behavior are recorded in Fig. 6. As noted these predictions are excellent.

Above a strain rate of about $\dot{\varepsilon} = 10^2/\text{sec}$ the applied force is sufficiently great to insure intersection without waiting for a thermal fluctuation and some other dislocation mechanism becomes rate controlling. This region has not yet been explored in sufficient detail to provide a clear picture of the various issues that might be involved. For example, deformation in this region could be due to the thermally aided production of vacancies as a result of the motion of jogged screw dislocations. It is a simple matter to show that for this mechanism also, $\ln \dot{\varepsilon}$ would be proportional to σ, albeit with a different slope than that for the intersection mechanism. Since this prediction is inconsistent with the experimental facts shown in Fig. 6, it becomes obvious that the high strain-rate effects shown in Fig. 6 cannot be ascribed exclusively to the thermally activated motion of jogged screw dislocation. It is, of course, possible that a number of different difficult dislocation mechanisms contribute to the high strain-rate behavior. A short time ago the authors[26] suggested that the leveling off of the strain rate might be due to the fact that the dislocations were reaching their limiting velocity, c, namely the velocity of a shear wave. At this limiting velocity, the limiting strain rate $\dot{\gamma}_{\ell}$ is

$$\dot{\gamma}_{\ell} = m b c \tag{12}$$

where m is the density of dislocations. For the highest experimentally observed strain rates, however, m must be taken to be at least two orders of magnitude less than the actual dislocation density in order to achieve agreement with the experiment. Obviously the assumption of relativistic velocities of dislocations in this case is highly untenable.

When the data given in Fig. 6 are replotted on linear scales, the results shown in Fig. 7 are obtained. Here it is clearly evident that a limiting relativistically controlled strain rate is not being approached. Rather the strain rate appears to be a linear function of the applied stress. The linear dependence of the strain rate on the stress must arise from the operation of some yet unidentified dislocation velocity-sensitive dissipative force.

It is expected that at sufficiently high values of stress, dislocations might acquire the velocity of sound as originally predicted by Frank.[28] An example of an approach to this condition is given by the data of Johnston and Gilman[29] who measured the velocities of individual dislocations as a function of the applied stress. Their data, reproduced in Fig. 8, clearly illustrate that the dislocation velocity at high stresses approaches asymptotically that for a shear wave.

VII. DYNAMIC SUPPRESSION OF HIGH TEMPERATURE THERMALLY ACTIVATED MECHANISMS OF DEFORMATION

Slow mechanisms such as those dependent on diffusion cannot contribute to the dynamic straining. In this section a simple example of the suppression of such a mechanism will be described.

The critical resolved shear yield strength for prismatic slip of single crystals of the hexagonal phase of an alloy containing 67 atomic percent Ag and 33 atomic percent Al for shear strain rates of $\dot{\gamma} = 10^{-4}$/sec are shown by the solid line of Fig. 9.[30, 31] Whereas Regions I and III represent deformation by thermally activated mechanisms, Region II, where the yield strength is insensitive to the temperature, is athermal. Since dislocations on the prismatic plane are undissociated, this athermal behavior cannot be ascribed to Suzuki locking[32] and is, therefore, believed at present to result from short-range order hardening. The thermally activated mechanism responsible for plastic deformation in Region I cannot be intersection and is most likely the Peierl's mechanism since the activation volume is only about $v = 10b^3$. Phonemenologically, at least, it can be represented by the relationship

$$\dot{\gamma}_I = K e^{-\left\{\frac{U_p - v(\tau - \tau^*)}{kT}\right\}} \qquad (13)$$

which is analogous to Eq. 9 for intersection. Here, however, the strain hardening is negligible, v has the constant value of about $10b^3$ and τ^* refers to the stress necessary to disorder the alloy across the slip plane. Rather complete investigations[33, 34] conducted over Region III have shown that

$$\dot{\gamma}_{III} = 1.4\,\tau^{3.6} e^{-\frac{32,500}{RT}} \qquad (14)$$

revealing that this is also a thermally activated mechanism, although it has not yet been definitely identified in detail. The major issue to be made here is that the yield strength for dynamic test conditions, for $\dot{\gamma}$ between 55,000 and 71,000 per sec is given by the broken line. Thus for these dynamic conditions the Mechanism III operative over the high temperature region for slow strain rates appears to be supplanted by the Mechanism I appropriate only to the low temperature region when the strain rate is low.

VIII. STRAIN-RATE INSENSITIVE DEFORMATION

Over ranges where the flow stress is insensitive to the temperature, the flow stress is also insensitive to the strain rate. For intersection this occurs above the critical temperature T_c, as shown schematically in Fig. 4. It should apply also to Region II for prismatic slip of the Ag-Al intermediate phase, as shown in Fig. 9. Perhaps the most striking confirmation of this deduction is contained in the example of basal slip on the 67-33 atomic percent Ag-Al intermediate phase.

As shown in Fig. 10, the critical resolved shear stress for yielding by basal slip at slow shear strain rates of $\dot{\gamma} \simeq 10^{-4}$/sec is independent of the temperature from about 77° to 450°K. Above 450°K, however, the single crystals oriented for slip on the basal planes undertook instead, slip on the prismatic plane giving results in complete agreement with those recorded for Region III in Fig. 9. The slow strain-rate tests revealed a yield point phenomenon with a tremendous yield strain of 135%. The presence of a yield point demands that the dislocations on the basal plane were locked whereas those on the prismatic plane were not locked. Since the yield point was insensitive to the temperature and since the atomic radii of Ag and Al are almost identical, the observed locking could not be ascribed to Cottrell interactions. The only known mechanism that could simultaneously be responsible for the locking and the athermal behavior is the Suzuki mechanism. Such locking could only occur on the basal plane where dislocations dissociate into their partials to provide the stacking fault necessary for chemical locking.

The dynamic yield strength obtained for strain rates up to 6×10^3/sec are shown by the broken curve of Fig. 10. Over the high temperature range, the thermally activated prismatic slip noted in slow strain-rate tests was suppressed and only basal slip was observed. Here, as is only possible for Suzuki locking, the critical resolved shear stress increased somewhat as the temperature increased due undoubtedly to more complete chemical locking of the dislocations. And over the low temperature range from 77° to 450°K, the flow stress obtained in the dynamic tests were only slightly higher than those obtained for slow rates of deformation. This case then represents an extreme example of strain-rate insensitive deformation.

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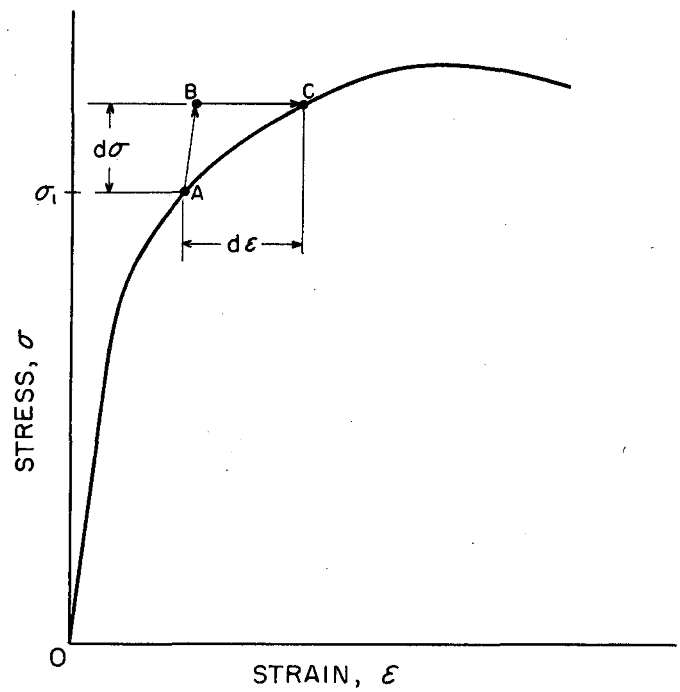

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Fig. 1. Stress-strain curve for infinitely slow deformation rates.

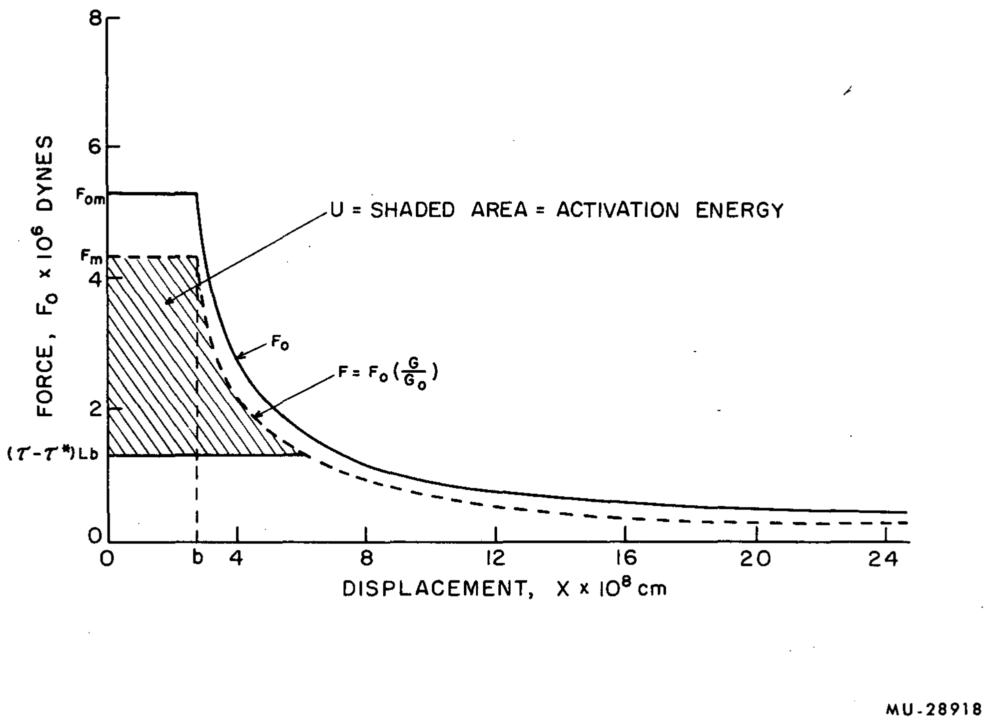

Fig. 2. Force-displacement diagram for intersection in pure Al (25).

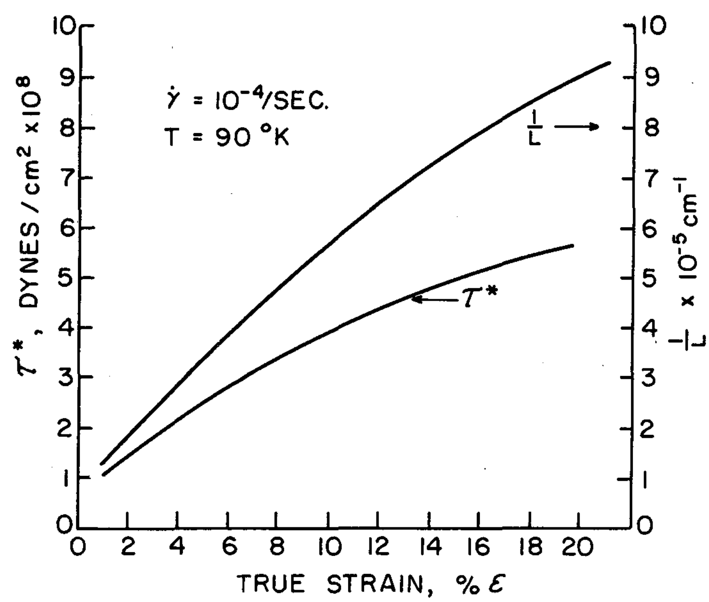

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Fig. 3. Variation of dislocation spacing and back stress fields with strain in pure Al (25).

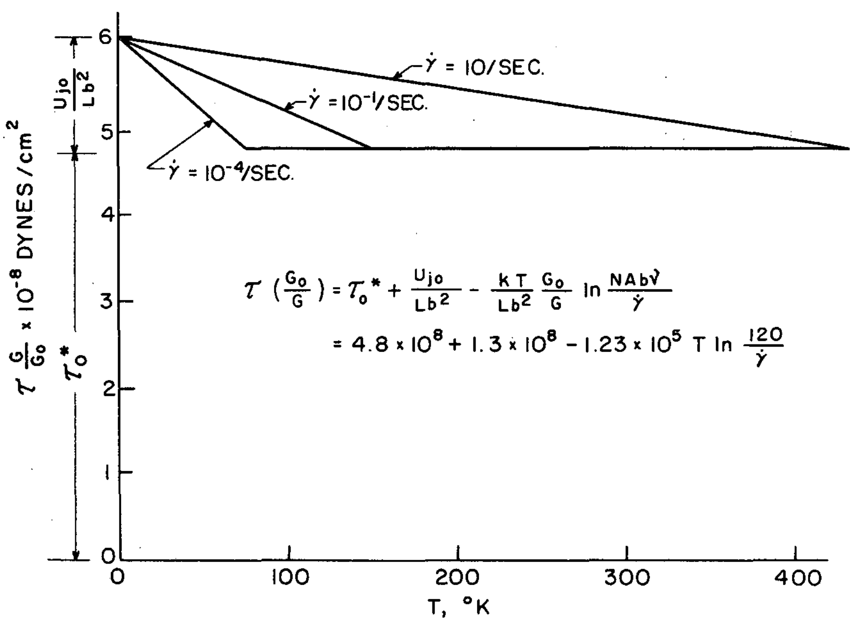

Fig. 4. Representation of the stress-temperature relation for intersection at a given structure.

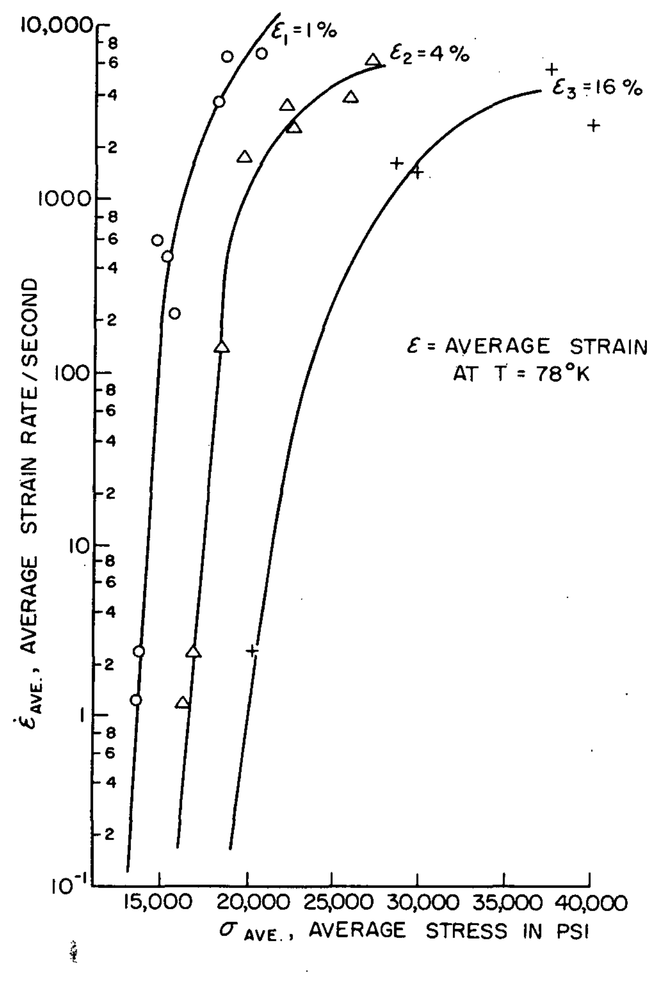

Fig. 5. Effect of stress on strain rate at constant strain (26).

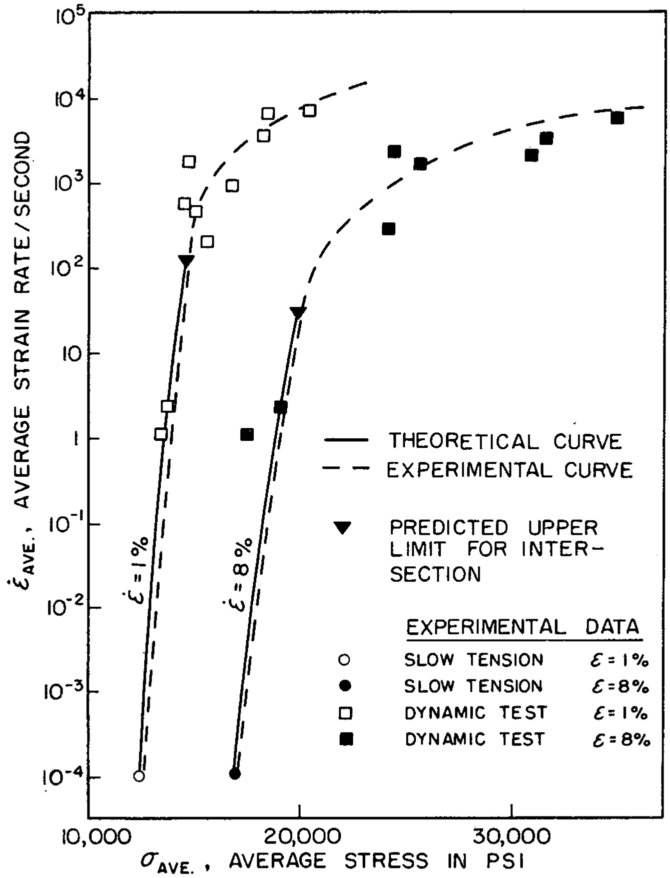

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Fig. 6. Prediction of the dynamic plastic behavior in the intersection region for pure Al at T = 77°K.

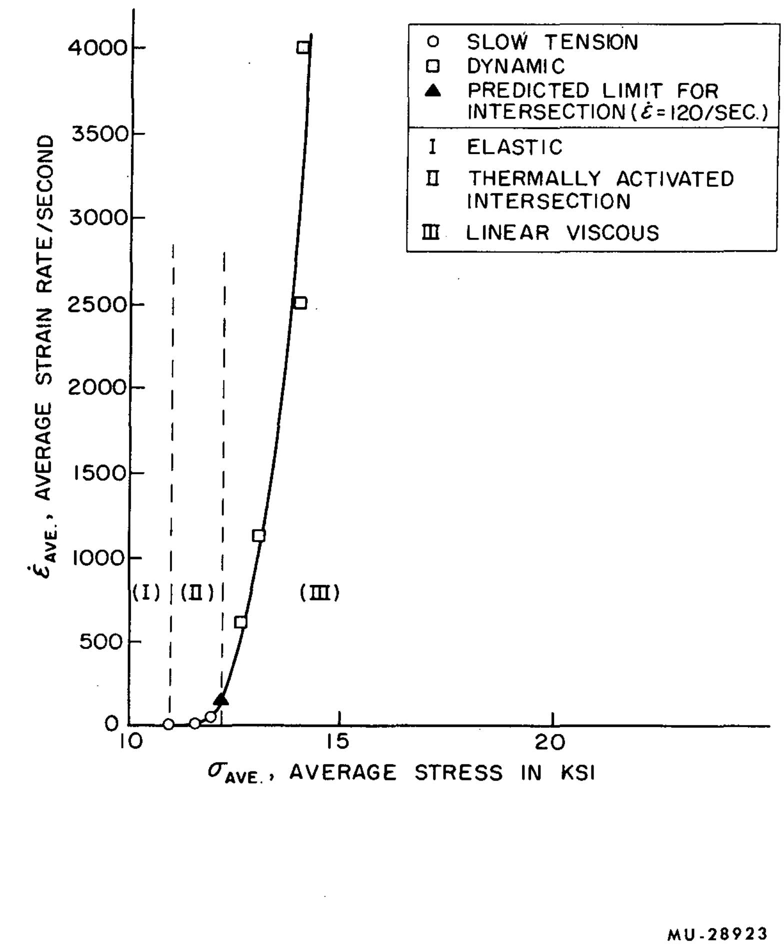

Fig. 7. Effect of stress on the strain rate in dynamic tests in pure Al at 295°K.

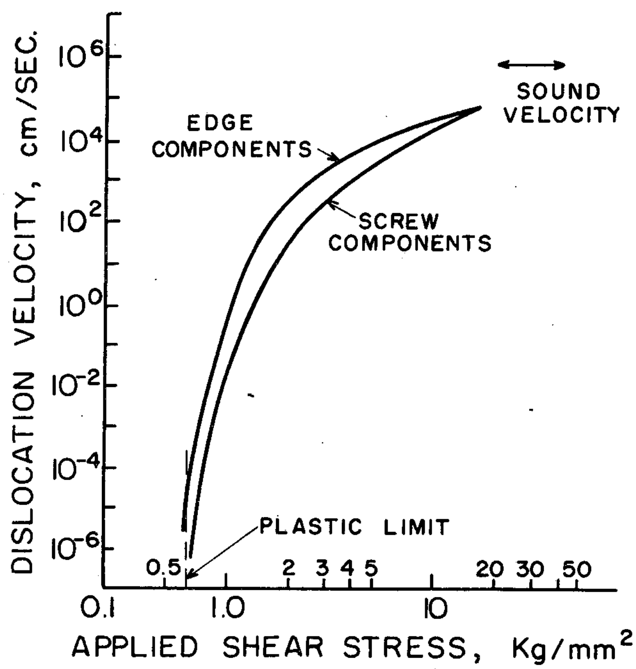

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Fig. 8. Dislocation velocities in LiF as a function of stress (29).

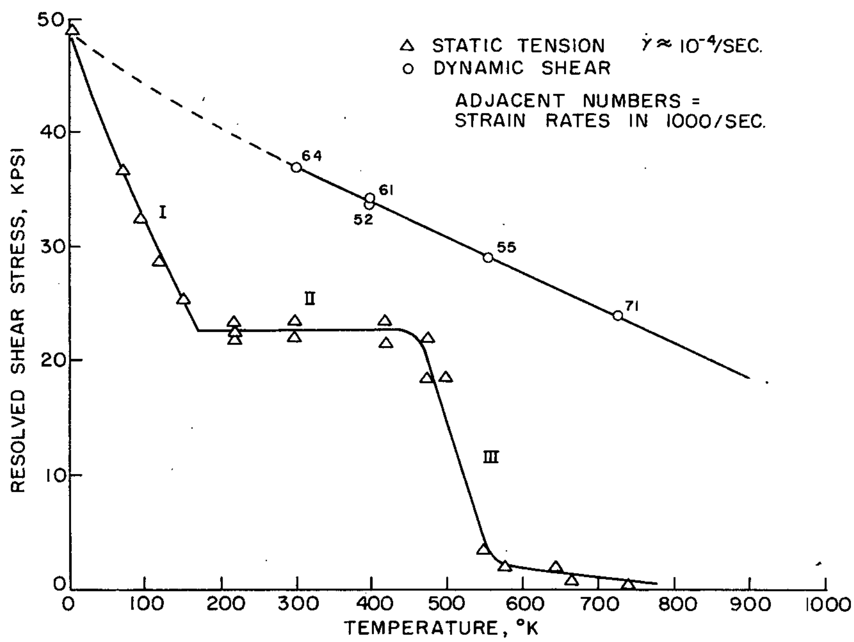

Fig. 9. Prismatic yielding of an Ag (67%)-Al(33%) alloy (30, 31).

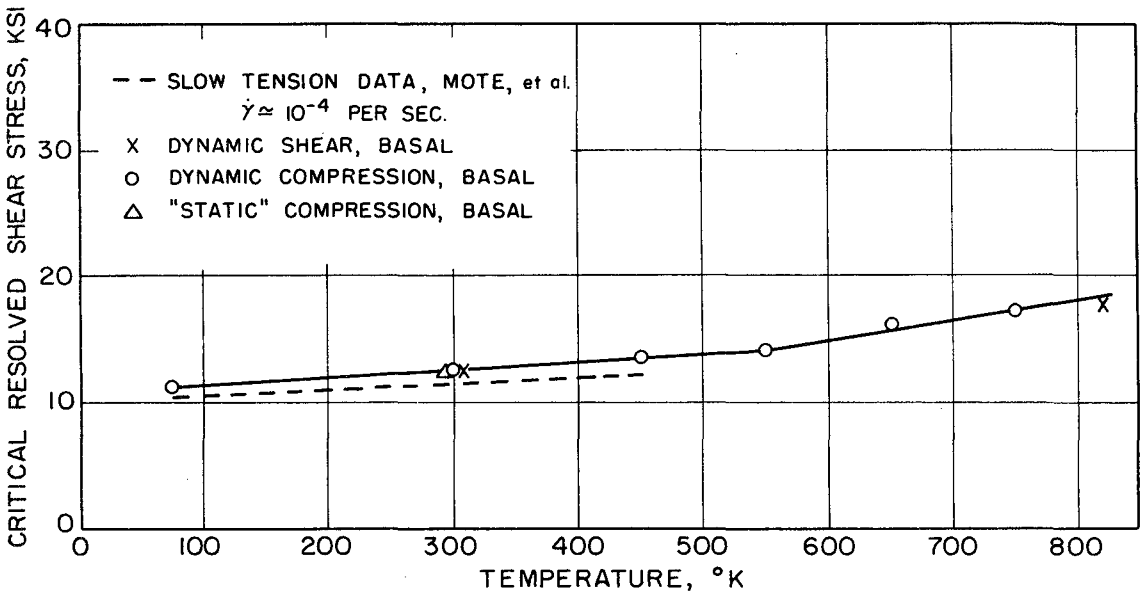

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Fig. 10. Critical resolved shear stress for yielding vs temperature.