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A. Rosen and A. Ahmadieh

It has been suggested that the controlling mechanism for the strain hardening of aluminum at low temperatures is the thermally activated intersection of dislocations.[1,2] The two important parameters of this model which determine the strain hardened state are the total back stress, τ^*, and the mean spacing between the forest dislocations, L. L can be calculated from experiments developed by Mitra, Osborne and Dorn.[1] The technique is to change the strain rate rapidly at various stress levels and to measure the simultaneous change of the stress. The quantity which can be obtained experimentally is

$$\beta = \frac{\partial \ln \dot{\gamma}}{\partial \tau} \simeq \frac{\Delta \ln \dot{\gamma}}{\Delta \tau} \tag{1}$$

where $\dot{\gamma}$ is the strain rate and τ is the resolved shear stress.

The Seeger equation of strain rate when the deformation is controlled by the intersection mechanism is:[3]

$$\dot{\gamma} = K e^{-U/kT} \tag{2}$$

when $K = NAb\left(\frac{\nu b}{L}\right)$

where N is the number of points per unit volume at which intersection can take place, b is the Burger's vector, ν is the Debye frequency, k is the Boltzmann constant, T is the absolute temperature, and A is the average area swept out per activated dislocation segment.

It is shown by Basinski[4] that the activation energy, U, for intersection is:

$$U = \int_{F}^{F_m} x \, dF \tag{3}$$

where x is the distance the dislocations have to travel to complete intersection, and:

$$F = (\tau - \tau^{*})\ Lb \tag{4}$$

assuming that K in Eq. 2 is constant, from Eqs. 1, 2, 3, and 4 we can derive:

$$\beta = \left(-\frac{1}{KT}\right)\left(\frac{\partial U}{\partial F}\right)\left(\frac{\partial F}{\partial \tau}\right) = \frac{xLb}{KT} \tag{5}$$

At constant strain rate, stress, and temperature, x is constant when NA is constant and therefore

$$\beta = \text{const. } L \tag{6}$$

Identical single crystals differing only in diameter are known to exhibit different stress-strain curves.[5, 6, 7] It was the authors interest to ascertain whether specimens having different sizes will have different β values or not.

Single crystals of high purity aluminum (99.994% Al) were grown in a graphite mold under argon atmosphere using the Bridgman method. The crystals were cylindrical, 4" long and 11/16" diam and were favorably oriented for (111) [$\bar{1}$01] single slip. They were reduced in a 2" gage section by the spark cutting method, etched in a 9 parts HCl, 3 parts HNO_3, 2 parts HF and 5 parts H_2O reagent, annealed under argon at 500°C for 15 min., etched again in the same solution and subsequently tested. The final diam of the specimens were 0.231, 0.387 and 0.550 in., the surface to volume ratio of the small specimen is approximately 2.5 times that of the large specimen. The experiments were carried out on an Instron Testing machine at 77°K with a strain rate of 1.67×10^{-4} per sec. After yielding, the strain rate was rapidly changed to 1.37×10^{-5} per sec. which resulted in a certain drop of the stress. These changes in the strain rate were repeated several times during straining. Figure 1 shows the true stress-shear strain curves for the three different size specimens at a strain rate of 1.37×10^{-4} per sec. In Fig. 2, the values of β are plotted vs the true stress for the same specimens.

The small and the medium size specimens have the same stress-strain curve and only the large specimen shows somewhat lower values of stress for the same strain. This can be also a result of slight variations between specimens and it seems that the three specimens have the same stress-strain curves within the limitations of the experiment. Figure 2, however, shows that the dependence of β on the flow stress is influenced by the size of the specimen. The smaller the diam the lower the value of β at the same stress and this tendency is more significant at lower stresses. Since $\beta kT = xLb$, according to Eq. 5 and xLb is the activation volume, it may be concluded that the activation volume is higher for large specimen at the same stress provided NA is not changed.

When the specimen deforms dislocations pile up on the surface due to irregularities or surface oxides and a debris layer develops on the surface, which is responsible for a back stress in addition to the back stress due to the dislocation entanglements in the bulk of the specimen. Since the smaller specimen has a larger surface to volume ratio, this surface back stress has a more significant influence on the activation volume of a small specimen. As the deformation proceeds, the distribution of the forest including the surface debris gets more and more uniform and the size effect diminishes, as clearly seen from the tendency of Fig. 2.

The fact that specimens of different diam have different β or L values at the same shear stress indicates that the strain hardened state of a crystal does not only depend on the stress, the temperature and the strain rate. The size which is responsible for the non-uniform distributions of the anchored dislocations may have a minor effect when the specimen is relatively large, but is certainly an important variable for small specimens having a large surface to volume ratio.

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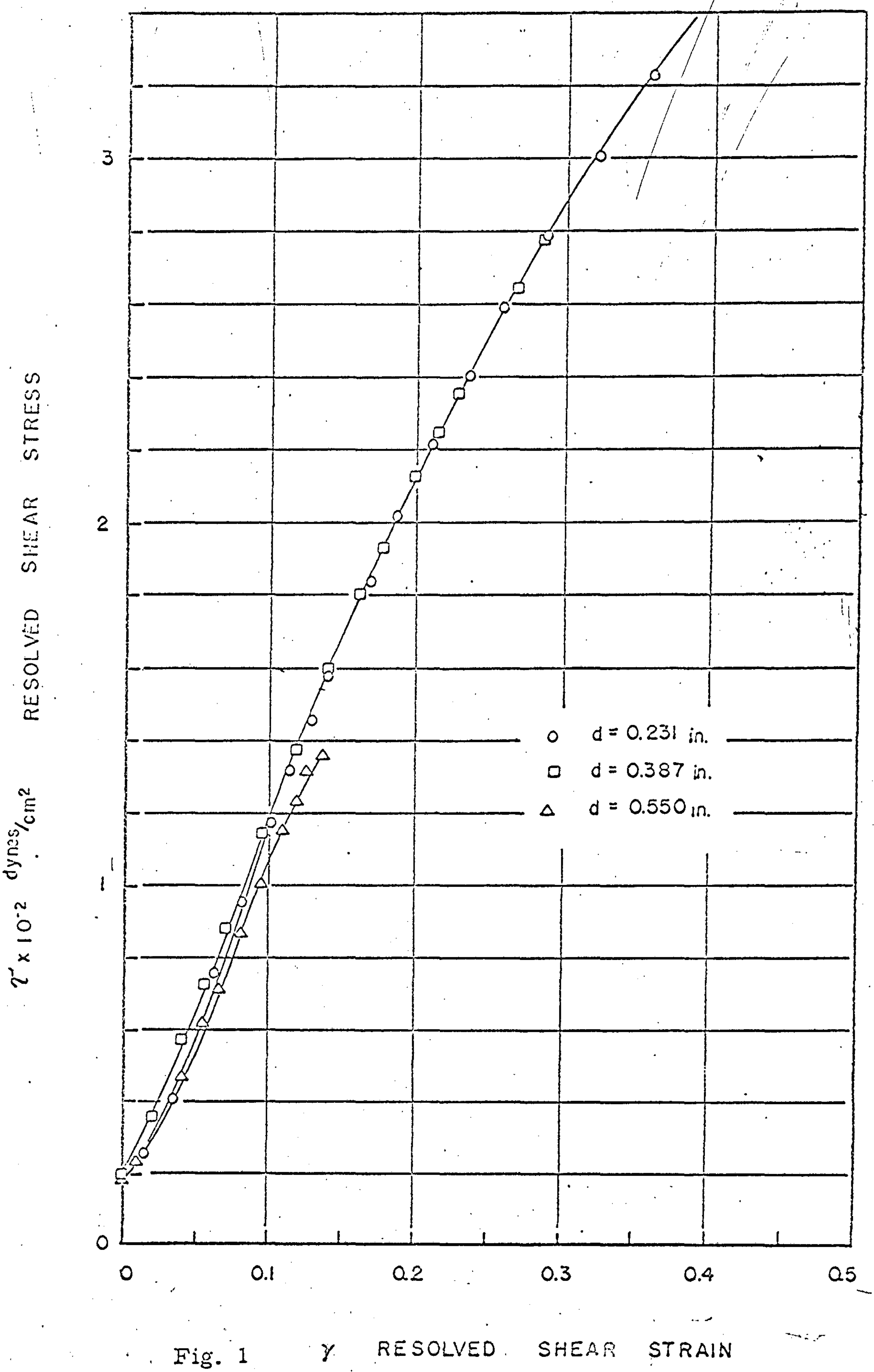
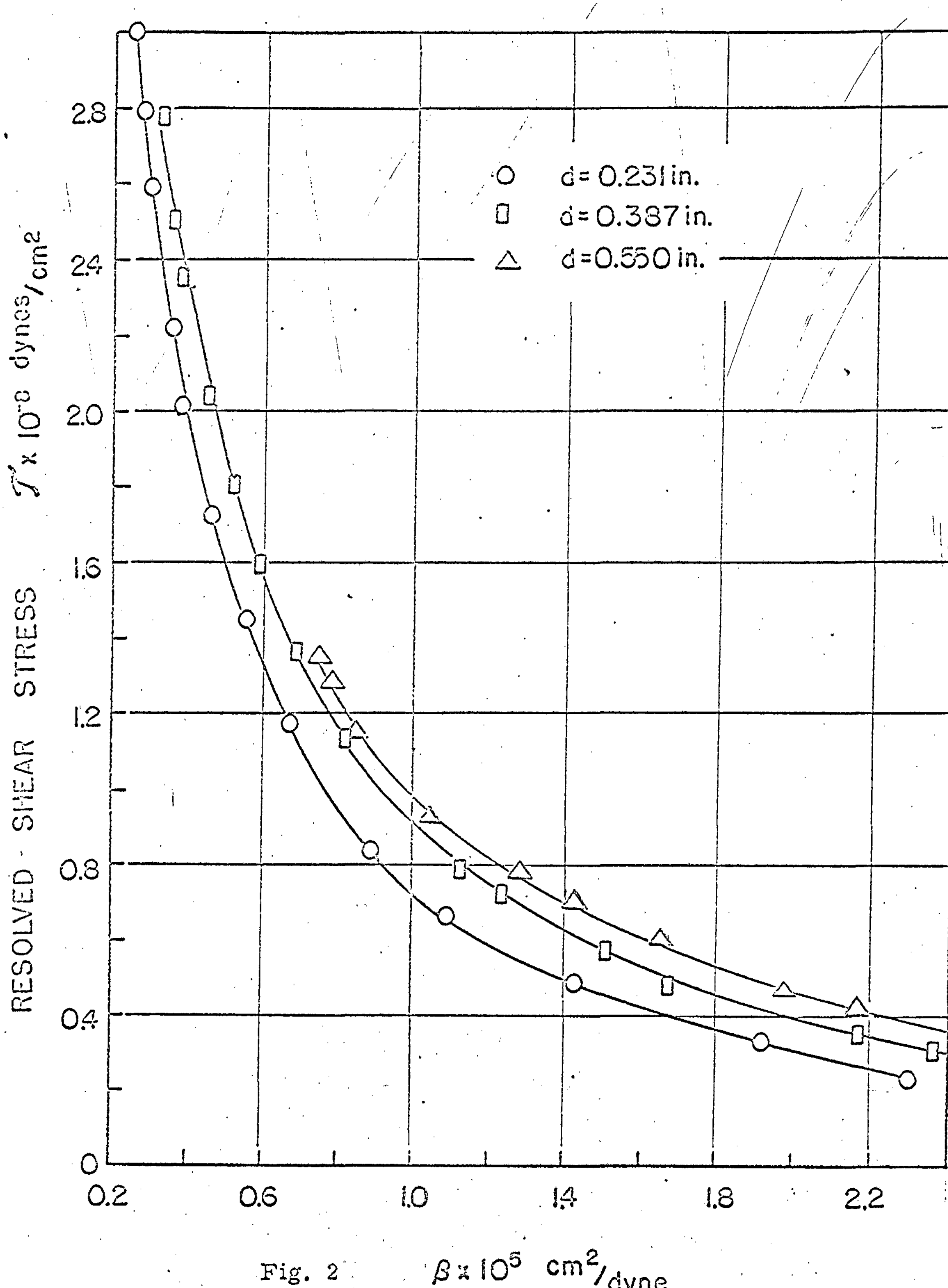

3
2
1
0
RESOLVED SHEAR STRESS
$\tau \times 10^{-2}$ dynes/cm²
d = 0.231 in.
d = 0.387 in.
d = 0.550 in.
0
0.1
0.2
0.3
0.4
0.5
γ RESOLVED SHEAR STRAIN

Fig. 1



d= 0.231 in.
d= 0.387 in.
d=0.550 in.
RESOLVED - SHEAR STRESS $\tau' \times 10^{-8}$ dynes/cm²
$\beta \times 10^5$ cm²/dyne

Fig. 2