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Discussion: Activation volumes of plastic deformation of crystals

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ABSTRACT

Two data sets reporting apparent activation volumes V_{app} of plastic deformation of ultrafine-grained Cu are compared. They differ strongly in magnitude and stress dependence of V_{app}. It is suggested that the difference results from differences in testing method. While V_{app} from stress relaxations appears to be consistent with expectations for the activation of thermally activated glide of dislocations in the bulk of the material, V_{app} from rate change tests appears to characterize the quasi-stationary deformation where dynamic recovery plays a major role. This underscores the necessity to specify which of the various thermally activated subprocesses of deformation are tested.

Plastic deformation of materials generally is thermally activated. Consider a material deforming at a temperature T under an applied stress $\sigma = \sigma_0$ at a strain rate $\dot{\epsilon} = \dot{\epsilon}_0$. When the stress is changed by a small amount $d\sigma$, the material reacts with a small relative change in strain rate given by $d \ln \dot{\epsilon}$. The parameter

$$V_{app} \equiv M\, k_B\, T\, \frac{d \ln \dot{\epsilon}}{d\sigma}, \tag{1}$$

where k_B is the Boltzmann constant and M is the geometrical factor converting the applied stress σ to the resolved shear stress $\tau = \sigma/M$ sensed by the dislocations, defines a volume. V_{app} contains information about the obstacles to dislocation motion that are overcome under stress with support by thermal activation and is called apparent (or operational) activation volume. It equals the true activation volume

$$V \equiv k_B\, T\, \frac{\partial \ln \dot{\epsilon}_{pl}}{\partial \tau}, \tag{2}$$

if

i) $\dot{\epsilon}$ is the *plastic* (irreversible) strain rate $\dot{\epsilon}_{pl}$, and
ii) among the parameters determining $\dot{\epsilon}_{pl}$ the stress τ is the only one that changes, i.e., changes of microstructural parameters and internal stresses are negligible.

Both these conditions are not trivial. $\dot{\epsilon}$ in Eq. (1) is usually taken to be the measured total strain rate $\dot{\epsilon}_{tot}$. As such, it is the sum

$$\dot{\epsilon}_{tot} = \dot{\epsilon}_{el} + \dot{\epsilon}_{inel} \tag{3}$$

of the elastic strain rate $\dot{\epsilon}_{el}$ and the inelastic strain rate $\dot{\epsilon}_{inel}$ (dependent on time t) that in turn has a reversible (anelastic) component $\dot{\epsilon}_{anel}$ and an irreversible (plastic) component $\dot{\epsilon}_{pl}$:

$$\dot{\epsilon}_{inel} = \dot{\epsilon}_{anel} + \dot{\epsilon}_{pl}. \tag{4}$$

$\dot{\epsilon}_{tot}$ equals $\dot{\epsilon}_{pl}$ only if $\dot{\epsilon}_{el}$ and $\dot{\epsilon}_{anel}$ are negligible. However, this is not certain. Making the realistic assumption of linear elastic behavior, the elastic strain is $\epsilon_{el} = \sigma/E_{eff}$, where E_{eff} is the effective elastic modulus of the material, accounting for the elastic reaction of the gauge length of the test specimen as well as all other elastic contributions to the measured length. The elastic strain rate

$$\dot{\epsilon}_{el} = \dot{\sigma}/E_{eff} \tag{5}$$

becomes large for large rates $\dot{\sigma}$ of stress change even if the elastic strains are negligible compared to the accumulated plastic strain. An analogous argument holds for the anelastic strain rate (resulting e.g. from reversible bowing/unbowing of stressed dislocation segments), as the anelastic strain generally scales with the elastic strain. So the rates of reversible strains may not be negligible. The assumption of constant microstructure is critical as well. The microstructure is not only defined by slowly changing parameters such as grain size

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and grain texture, but also comprises fast changing parameters such as internal stresses varying sensibly with the local arrangement of structure elements. For instance, unbowing of dislocations may have a significant relaxing effect on long-range internal back stresses of athermal nature that reduce the thermal stress component available for overcoming thermal obstacles to dislocation motion.

Despite these restrictions it has become customary in the literature to tacitly assume that V_{app} can be interpreted in terms of the process of bulk dislocation glide that in most instances makes the dominant contribution to plastic deformation. If that were true, V_{app} should be a well and uniquely defined parameter, independent of the experimental technique used for its measurement. However, this is not the case. To give another example, two recent data sets [1,2] of apparent activation volumes of ultrafine-grained Cu are compared in this work. The reported values of V_{app} differ greatly in magnitude and stress dependence. This difference is explained in terms of different experimental procedures capturing different subprocesses of plastic deformation.

In motor-driven machines deformation usually occurs at constant rate $\dot{l}_{tot}$ of change of total measured length l_{tot}. If the machine is stopped at a certain stress $\sigma = \sigma_0$, the change of inelastic length continues so that the elastic length component decreases. This causes relaxation of stress σ. Usually the testing assembly behaves linear elastically and the gauge length of the specimen makes the only contribution to the inelastic length change. Then the rate $\dot{\epsilon}_{inel}$ of inelastic deformation of the specimen is oppositely equal to the elastic deformation rate $\dot{\epsilon}_{el}$ (Eq. (5)):

$$\dot{\sigma} = -E_{eff}\ \dot{\epsilon}_{inel}. \tag{6}$$

With the aid of Eq. (6) the measurement of the inelastic strain rate of the specimen is reduced to a measurement of the rate at which the stress σ relaxes when the total strain is held constant. As the stress is measured with high accuracy, the experimental scatter in σ is low. This allows one to determine $\dot{\sigma}$ as well as the related value of $\dot{\epsilon}_{inel}$ at high accuracy. For constant E_{eff}, the strain interval, where σ relaxes from a starting value σ_0 at an inelastic strain $\epsilon_{inel,0}$ to some lower value σ_1 at $\epsilon_{inel,1}$, follows by integration of Eq. (6) as $\epsilon_{inel,1} - \epsilon_{inel,0} = (\sigma_{inel,0} - \sigma_{inel,1})/E_{eff}$. As E_{eff} is usually large compared to σ, the strain interval necessary for determination of V_{app} is relatively small. This means that microstructural changes needing plastic strain are minimized in stress relaxation tests. V_{app} may be determined in two ways (see e.g. the book of Caillard and Martin [7]). One is to fit the measured $\sigma(t)$-curve of stress relaxation. The other one is to determine $\dot{\epsilon}_{inel}$ via Eq. (6) and apply Eq. (1). Isaev et al. [2] used the first method in their investigation of ultrafine-grained (UFG) Cu. Their results for V_{app} are shown in Fig. 1 (a). It is seen that the V_{app}-values of Isaev et al. (black circles) are about $4b^3G/\sigma$ at stresses around $10^{-2}G$. The two data points result from two somewhat different UFG materials of similar strength produced by direct and equal channel hydroextrusion.[1] Within experimental uncertainty they are in fair agreement with the values recently reported in [4] (gray plus symbols) for UFG Cu at 373 K. These data confirm an inverse stress dependence of V_{app} measured by stress relaxation.

Tests with changes of total strain rate $\dot{\epsilon}_{tot}$ are more commonly used for V_{app}-measurement than stress relaxation tests. Using the identity $d\ln\sigma = d\sigma/\sigma$, Eq. (1) is rewritten as

$$V_{app} \equiv M\ k_B\ T\ /\ (m\sigma), \tag{7}$$

[1] With decrease of T down to 25 K Isaev et al. report a decrease of V_{app} by a factor of 4 to 5, so V_{app} remains larger than $b^3\sigma/G$; this limited, T-dependent decrease of V_{app} is regarded irrelevant for the present comparison of data.

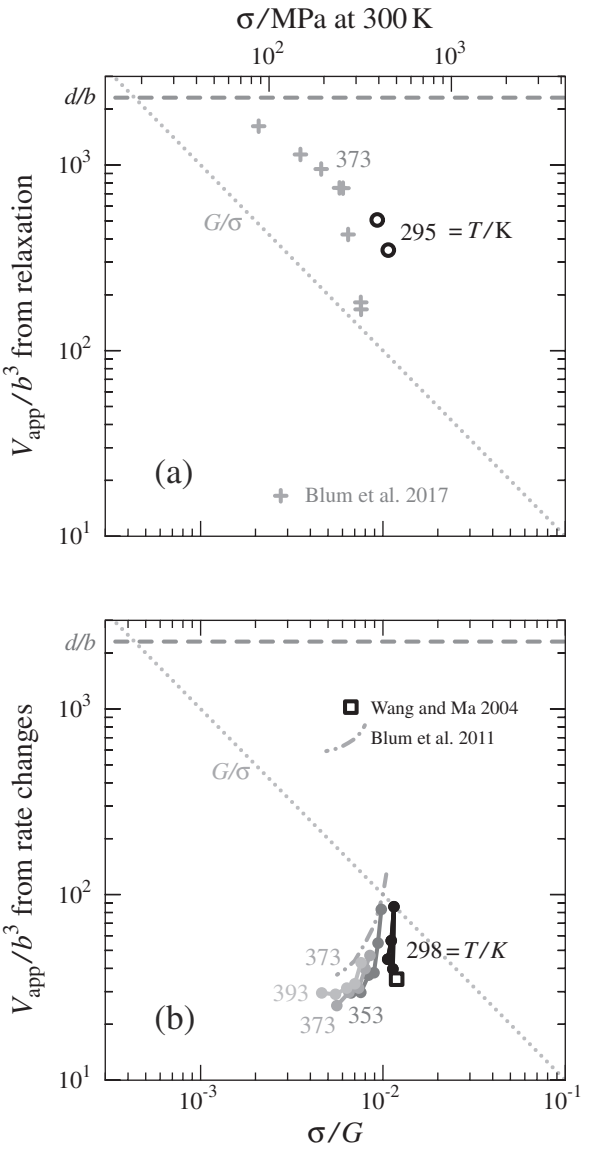

Fig. 1. Apparent activation volumes V_{app} of UFG Cu (in units of b^3, where $b = 2.56\times10^{-10}$ m is the Burgers vector length of Cu [3]) as function of stress σ (in units of shear modulus $G(T)$ [3]) (a) from stress relaxations of Isaev et al. [2] with data from [4], (b) from rate change tests of Duhamel et al. [1] (filled circles); with datum from Wang and Ma (cryo-rolled Cu) [5] and data set of Blum et al. [6] (dash-dotted, 24 passes of equal channel angular pressing); dashed line for grain size $d = 6\times10^{-7}$ m in units of b; dotted reference line G/σ represents estimate of the activation volume of thermally activated glide in pure materials (see text).

where $M = 3.06$ is the Taylor factor and

$$m = \frac{d\ln\sigma}{d\ln\dot{\epsilon}} \equiv \frac{1}{n} \tag{8}$$

is the strain rate sensitivity of the flow stress; n is the stress exponent of the strain rate. m is approximated as

$$m = \frac{\Delta\ln\sigma}{\Delta\ln\dot{\epsilon}} = \frac{\ln(\dot{\epsilon}_0) - \ln(\dot{\epsilon}_1)}{\ln(\sigma_0) - \ln(\sigma_1)} = \frac{\log(\dot{\epsilon}_0/\dot{\epsilon}_1)}{\log(\sigma_0/\sigma_1)}; \tag{9}$$

the subscripts 0 and 1 denote the values of strain rate $\dot{\epsilon}$ and stress σ at the point where the rate change starts and the point where the rate change is finished. Fig. 1 (b) shows the results for V_{app} obtained in the rate change tests of Duhamel et al. [1] for UFG Cu (filled circles) at four temperatures $295 \le T/K \le 393$. V_{app} is mostly lower than