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On dislocation mechanism of dynamic deformation of uranium

V.A. Pushkov *, D.V. Tsisar

Russian Federal Nuclear Center – VNIIEF, Dynamical strength, Mira ave., 37, Sarov 607190, Nizhni Novgorod Region, Russia

A B S T R A C T

In this work, based on results of the tests with study of dynamic diagrams of compression of uranium–molybdenum alloy, we made an attempt to determine dislocation velocity, length of free run of dislocations, and increase of dislocation density during plastic deformation of alloy at dynamic strain rates.

The study of the dynamic mechanical properties of uranium and its alloys is of significant interest both in Russia and abroad, for example, in [1–6]. Such characteristics as diagrams of compression, tension, spall and shear strengths, shock adiabats and other characteristics have been determined over a wide range of the temperature–velocity conditions of loading [1–6]. Also the authors of these works studied from microstructural changes in tested samples up to changes of dislocation structures. For example, large number of twins was revealed in [5], where shear deformation of alloy of uranium with 6% of niobium was studied at strain rate of $\sim$2000 s^{-1}. At the same time, the kinetics of plastic deformation of uranium and its alloys have been poorly studied. Using the results of the tests from a study of dynamic diagrams of compression of uranium–molybdenum alloy ($\sim$1% Mo) [1], we made an attempt to determine dislocation velocity, length of free run of dislocations, and increase of dislocation density during plastic deformation of alloy at strain rates of 600–1400 s^{-1}. In the work [1], using the split Hopkinson bar technique, we studied tension and compression diagrams σ–ε ('stress–strain') of uranium and its alloy with molybdenum at dynamic strain rates of 100–1800 s^{-1} and temperatures 20–600°C.

It is known that slip and twinning are the basic mechanisms of plastic deformation [7]. The dislocation mechanism of twinning is rather complicated. Therefore this work is associated with kinetics of deformation mechanism by slip. In [8], the formula was obtained for calculation of dislocation velocity depending on tangential stress applied to the slip plane. Force f per unit of dislocation length is,

$$f = \sigma_s \cdot b \tag{1}$$

where σ_s is the tangential stress, b is the Burgers vector. Three forces are acting on rectilinear dislocation line during dynamic equilibrium. One force tends to move it forward. This force is the moving force,

$$f = \sigma_s \cdot b \tag{2}$$

which is caused by the relevant stress. The other two forces act against its motion. These forces are the inertia force,

$$f_i = m \cdot a \tag{3}$$

where m and a are the dislocation mass and the dislocation acceleration accordingly; and the force of viscous deceleration,

$$f_v = B \cdot v \tag{4}$$

where B is the attenuation constant, v is the dislocation velocity. Thus,

$$f = f_i + f_v \text{ or } \sigma_s \cdot b = m \cdot dv/dt + B \cdot v \tag{5}$$

After separation of variables and integration, we obtain the following:

$$v(t) = \left(\frac{\sigma_s b}{B}\right)\left[1 - e^{-(B/m)_t}\right] \tag{6}$$

In the other words, dislocation velocity equals to zero at $t = 0$. As velocity grows, value of the viscous force approaches value of the moving force, acceleration is reduced, and velocity approaches value of the stationary velocity,

$$v_{ss} = \sigma_s \cdot b/B \tag{7}$$

Dislocation motion occurs during loading. Since the flow stress is changed during plastic deformation, velocity of dislocations is changed as well. Tests in [1] were performed during uniaxial tension and compression. So, for determination of tangential stresses, which cause slip, the authors used flow stresses from 'stress–strain' σ–ε diagrams presented in [1]. We used for this case averaged diagrams of alloy compression at strain rates of 600–880 s^{-1} (average value 740 s^{-1}) and 1000–1400 s^{-1} (average value 1200 s^{-1}) from [1], which are presented in Fig. 1.

In order to convert flow stress into tangential stress, we used the Shmid criterion from [7]:

$$\sigma_s = \sigma \cdot k \tag{8}$$

* Corresponding author. Tel.: +7 83130 450 09; fax: +7 83130 459 58.
E-mail address: root@gdd.vniief.ru (V.A. Pushkov).

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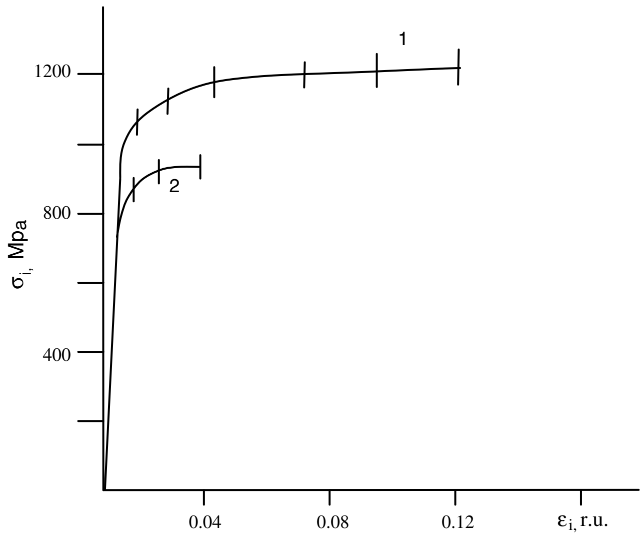

Fig. 1. Averaged diagrams of alloy compression for temperature 20 °C and different strain rates $\acute{\varepsilon}$: 1–$\acute{\varepsilon}$ = 1000–1400 s^{-1} (average value 1200 s^{-1}); 2–$\acute{\varepsilon}$ = 600–880 s^{-1} (average value 740 s^{-1}).

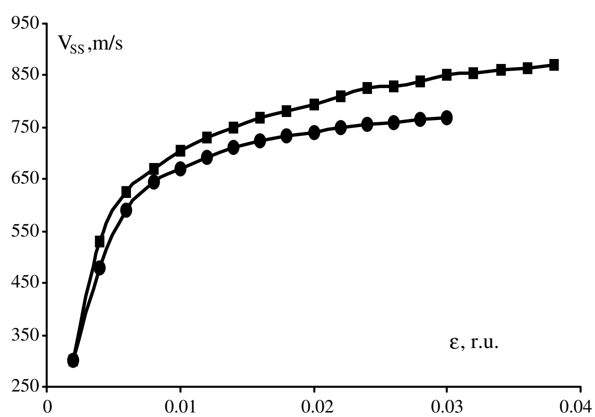

Fig. 2. Dislocation velocity versus degree of deformation of uranium–Mo alloy for 20 °C and various $\acute{\varepsilon}$: –●– $\acute{\varepsilon}$ = 740 s^{-1}; –■– $\acute{\varepsilon}$ = 1200 s^{-1}.

where k is the Shmid factor. The Shmid factor takes account for dependence of tangential stress on plane orientation, and it is changed from 0 to 0.5 when orientation is changed. For a polycrystal we chose an average value $k_{av.}$ Assuming absence of texture, we selected $k_{av.}$ = 0.25. Because of the absence of data on the alloy, B was assumed to be 10^{-4} Pa s (that is typical for metals with technical purity [8]), and the Burgers vector is $b = 2.8538 \times 10^{-10}$ m [9]. Fig. 2 presents the dependence of dislocation velocity on alloy deformation for two different strain rates.

If dislocation moved for all grain length ℓ_1 with formation of a step having value b at its edge, the value of plastic deformation is,

$$\varepsilon = b/\ell_1 \tag{9}$$

during run of one dislocation, and

$$\varepsilon_n = n \cdot b/\ell_1 \tag{10}$$

during run of n dislocations. However, not all n dislocations are able to run through all crystal totally, i.e. dislocation stopped by a barrier will travel the distance $\ell_{fr} < \ell_1$, where ℓ_{fr} is length of free run of dislocation. In this case, value ε_n will be by ℓ_{fr}/ℓ_1 less, i.e.,

$$\varepsilon_n = b\ell_{fr} \cdot n/\ell_1^2 = b\ell_{fr} \cdot N_n \tag{11}$$

where

$$N_n = \acute{n}/\ell_1^2 \tag{12}$$

is density of movable dislocations, their number is $\acute{n} < n$. Further it is possible to determine increase of density of dislocations as plastic deformation grows by the formula:

$$\Delta N = \varepsilon/b \cdot \ell_{fr} \tag{13}$$

To account for the theoretical background for calculation of ℓ_{fr} [7,8] and available experimental data, the dependence,

$$\ell_{fr} = f(\varepsilon) \tag{14}$$

was formulated. It is presented in Fig. 3. It is possible to see that ℓ_{fr} is sharply reduced in the beginning of plastic deformation. It facilitates formation of 'forest' of dislocations when, after traveling ℓ_{fr}, dislocations are stopped, and successive dislocations are reacting with primary dislocations. Then after ~1.5% of deformation, ℓ_{fr} is smoothly reduced. Usually it occurs, if there is no a flowability site at the 'stress–strain' σ–ε diagram (alloy diagrams from [1] are such diagrams) after transition of yield strength, when cellular structure starts being formed. The type of diagrams with a site (or a zone) of flowability is characterized by the fact that, after reaching the yield strength, strain grows, but stress is not increasing for some period of deformation. Cellular structure is not formed here. Then as strain grows further, stress starts increasing. Fig. 4 presents the dependence of dislocation density growth ΔN on strain ε obtained with account for ℓ_{fr} reduction as ε grows. There the values of dislocation

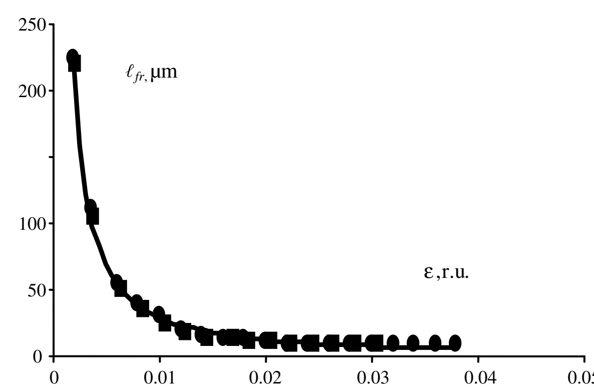

Fig. 3. Length of free run of dislocations versus degree of deformation of uranium–Mo alloy: –●– $\acute{\varepsilon}$ = 740 s^{-1}; –■– $\acute{\varepsilon}$ = 1200 s^{-1}.

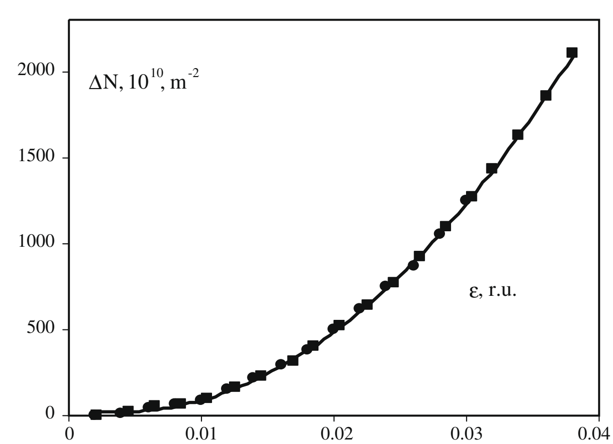

Fig. 4. Graph of dislocation density growth versus degree of deformation of uranium–Mo alloy with account for reduction of ℓ_{fr}: –●– $\acute{\varepsilon}$ = 740 s^{-1}; –■– $\acute{\varepsilon}$ = 1200 s^{-1}.

density ΔN are calculated in comparison with the initial dislocation density $\Delta N_{init.} = 1200 \times 10^{12}$ m^{-2}, which was obtained in the uranium–molybdenum alloy before the dynamic compression tests [1] by the metallography.

Thus, we considerated some aspects of dislocation mechanism of uranium dynamic deformation. Some parameters for the dependences of dislocation mechanism are approximate, but for elaborating more accurate dependences we have to have more experimental data.

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