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INFLUENCE OF STRAIN RATE ON FLOW STRESS OF TANTALUM*

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Abstract

The influence of strain rate on the flow stress of 99.9% pure, fully recrystallized tantalum is examined. Stress-strain curves in both tension and compression are obtained at strain rates from 10^{-5} to 5,000 sec^{-1}. The dependency of the upper and lower yield stress on strain rate is closely predicted by the Johnston-Gilman model. However, this model does not adequately describe post-yield hardening behavior. The Dorn-Rajnak model appears to describe the dynamic behavior better; it not only predicts yield behavior accurately but provides an excellent description of post-yield behavior.

Introduction

The most widely accepted theory of the upper and lower yield points in crystalline materials is that of Johnston and Gilman (1). This theory attributes the yield drop to the multiplication of dislocations associated with accumulated plastic strain. The rate dependence of the yield stress is described in terms of the stress sensitivity to average dislocation velocity. The basic model may be modified to include post-yield hardening effects. The mathematical representation of this model is empirical.

A more fundamental theory describing strain rate effects on bcc metals is that proposed by Dorn and Rajnak (2). Their model is based on the kinetics of dislocation motion associated with the Peierls' mechanism.

This paper compares the flow behavior predicted by these theories to the actual flow behavior of tantalum, chosen because of its remarkably high sensitivity to strain rate effects. The rheological behavior of high purity tantalum was studied over a range of approximately nine decades of strain rate. Both tension and compression tests were conducted. Data showed that tantalum is extremely sensitive to strain rate. Over the regime of strain rates tested, the upper yield stress increases by almost 500%. Post-yield behavior is slightly less sensitive. In tension tests, the elongation at fracture decreases by about 60%.

Experimental Technique

Tension and compression tests at strain rates below 10 sec^{-1} were performed on an MTS hydraulically actuated universal test machine (20,000-lb capacity) programmed for a constant displacement rate. For tests slower than 0.01 sec^{-1}, strain was measured by an Instron Model G-51-11 extensometer. Load and strain were plotted as functions of time on a two-pen strip chart recorder. For the higher rate tests, an Optron

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Model 680 optical extensometer was used to measure strain. Data were plotted versus time on a Tektronix Model 555 oscilloscope. In addition, load versus strain was recorded on a Tektronic Model 536 oscilloscope.

Tests at strain rates from 10 to 200 sec^{-1} were conducted on a modified Dynapak metal-working machine (3). Specimen stress was measured by a strain-gaged load cell placed in series with the specimen. Strain was measured by the Optron extensometer. Data were recorded on oscilloscopes, as described above.

The Hopkinson split-bar test technique was used to obtain data at rates above 500 sec^{-1} (4,5). Test data were recorded as a function of time on a Tektronix Model 555 oscilloscope. A typical oscilloscope record is shown in Fig. 1.

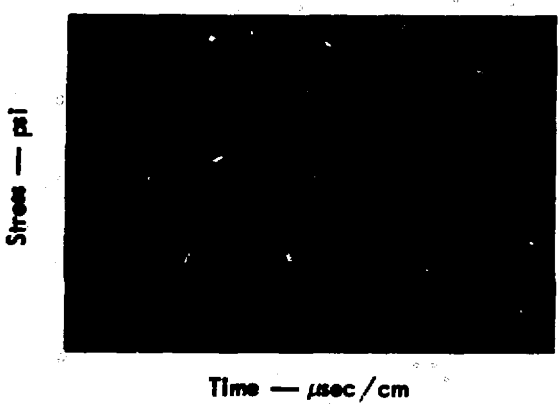

FIG. 1
Sample data trace for a Hopkinson bar tension test.

Test specimens were machined from a single rod of 99.9% pure, fully recrystallized tantalum obtained from Kawecki-Berylco Industries. Compressive specimens were disk-shaped, 0.4 in. in diam and from 0.3 to 0.6 in. thick. For tensile tests on the MTS and Dynapak machines, round specimens with a gage section 1.25 in. long and 0.252 in. in diam were used. For Hopkinson bar tests, specimens had a gage length of 1/4 to 1/2 in. and a diam of 0.16 in.

Theoretical

Johnston-Gilman Model. The Johnston-Gilman theory is based on the assumption that dislocation multiplication causes yield drop in metals. The authors hypothesize that aged dislocations remain locked, and that yield drop and subsequent plastic flow result from mobile dislocations being rapidly nucleated. Their theory also attempts to describe the sensitivity of flow stress to strain rate through the stress sensitivity of the dislocation velocity. Hahn used the Johnston-Gilman theory to analyze the rate sensitivity of several bcc metals (6). However, to simplify his analysis, he assumed the absence of pre-yield microstrain. In this paper, that assumption is not allowed.

The governing equation for the Johnston-Gilman model is

$$\dot{\epsilon}_p = \phi b \, (\rho_o + \alpha \epsilon_p) \, v^* \exp\left[-(D + H\epsilon_p)/\sigma\right]. \qquad (1)$$

The derivation of the equation and the technique used to solve it have been described in detail by Hoge and Gillis (7).

The following values of constants were used to evaluate equation (1) (references are indicated when appropriate): modulus $E = 27 \times 10^6$ psi, Burgers vector $b = 2.86$ Å, orientation factor $\phi = 1$, initial dislocation velocity $\rho_o = 10^6$ cm^{-2}, multiplication coefficient $\alpha = 1.5 \times 10^{12}$ cm^{-2} (8), and limiting dislocation velocity $v^* = 8.14 \times 10^4$ in./sec (the shear wave velocity) (9). The two remaining parameters, D and H, were adjusted to agree with upper and lower yield stresses obtained from experiments. Values of $D = 5.3 \times 10^5$ psi and $H = 8.1 \times 10^7$ psi provided the best representation of experimental data.

Dorn-Rajnak Model. The Dorn-Rajnak model predicts the flow characteristics of bcc metals from the kinetics of dislocation motion described by the Peierls' mechanism. Using the Boltzmann approach, the frequency of nucleation of a pair of kinks in a length L (10) is

$$\nu_n = \frac{\nu b L}{2w^2} \exp(-U_n/kT) \tag{2}$$

where ν is the Debye frequency, w the width of a kink loop, U_n the energy to nucleate a pair of kinks, k the Boltzman constant, and T the absolute temperature. If a is the distance between Peierls' valleys, the average velocity of a dislocation moving as a result of nucleation is $\nu_n a$, and the average strain rate becomes

$$\dot{\epsilon} = \frac{\rho L a b^2}{2W^2} \nu \exp(-U_n/kT). \tag{3}$$

Guyot and Dorn (10) have shown that

$$\frac{U_n}{2U_k} = \left(1 - \frac{\sigma^*}{\sigma_p}\right)^2 \tag{4}$$

where U_k is the single kink energy under zero effective stress, σ_p is the Peierls' stress, and σ^* is the component of stress necessary to help dislocations surmount the Peierls' barrier. Thus, the stress necessary for plastic flow is

$$\sigma = \sigma^* + \sigma_a \tag{5}$$

where σ_a is the athermal stress component which overcomes long range obstacles. σ_a is independent of strain rate. The governing equation for the Dorn-Rajnak is obtained by combining equations (3) and (4) as follows:

$$\dot{\epsilon} = \frac{\rho L a b^2}{2w^2} \nu \exp -\left[\frac{2U_k}{kT}\left(1 - \frac{\sigma^*}{\sigma_p}\right)^2\right]. \tag{6}$$

The following values were used in evaluating equation (6): $L = 10^{-4}$ cm (2), $a \approx b = 2.86$ Å (2), $w = 24b$ (10), $U_k = 0.31$ ev (10), and $\nu = 5 \times 10^{12}$ H_3. A dislocation density of 10^8 cm^{-2} was found to give the best representation of experimental data. σ^* and σ_a were evaluated from a series of low temperature tests conducted at a constant strain rate of 10^{-4} sec^{-1}. Results of these tests are presented in Fig. 2, which shows σ_p = 155,000 psi and σ_a = 18,000 psi.

Results and Discussion

Figure 3 compares the results of experimental and theoretical data. Although both the Dorn-Rajnak and Johnston-Gilman models provide a reasonable description of the rate effect on upper yield stress, the Dorn-Rajnak model is felt to give a better representation. At rates above 10^3 sec^{-1}, this model based solely on the Peierls' mechanism deviates from experimental data. The discrepancy is attributed to a dislocation damping mechanism which tends to reduce the average dislocation velocity (11). When damping is the controlling mechanism, stress is a linear function of strain rate. For tantalum at rates of about 10^3 sec^{-1}, $\sigma^* < \sigma_p$, so it appears that both damping and

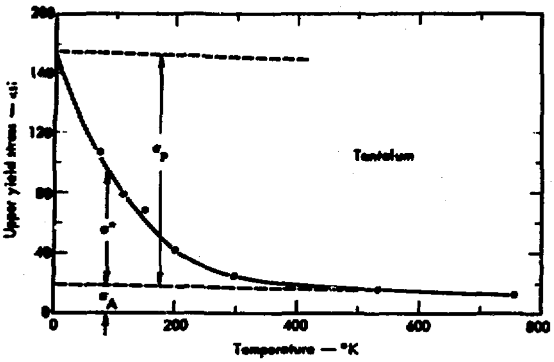

FIG. 2
Upper yield stress versus temperature.

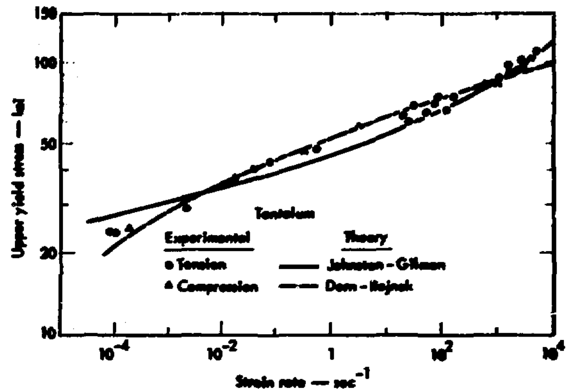

FIG. 3
Effect of strain rate on the upper yield strength of tantalum.

Peierls' mechanisms are operative. Both mechanisms can be incorporated in a single equation by calculating an average dislocation velocity equal to the distance between Peierls' valleys divided by time spent at a barrier plus time between barriers (12).

The Johnston-Gilman model predicts the rate effect on lower yield stress with approximately the same accuracy as upper yield (7). However, it fails to describe post-yield behavior accurately.

The best method to check the accuracy of the Dorn-Rajnak model in predicting flow behavior is through a rate-temperature parameter P. Taking the logarithm of both sides of equation (6) and letting A equal the pre-exponential constant, one obtains

$$T \log (A/\dot{\epsilon}) = \frac{2U_k}{2.3k}\left(1 - \frac{\sigma^*}{\sigma_p}\right)^2 = P. \qquad (7)$$

Equation (7) predicts that tests conducted at various temperatures and rates should have similar stress-strain curves for equal values of P, under the assumption that strain hardening is a function only of σ^* and σ_p. Figure 4 shows the stress-strain curves obtained at a strain rate of 10^{-4} sec^{-1} at various temperatures. Figure 5 pictures stress-strain curves performed at various rates at room temperature. Since curves with approximately equal P values are quite similar, this approach appears to be valid, closely approximating ductility as measured by fracture strain.

Finally, it should be pointed out that in evaluating the two models some of the constants were evaluated to provide a best fit to data. In the Johnston-Gilman model, D and H were arbitrarily selected. It would be difficult to find legitimate values for these two constants in literature. However, in the Dorn-Rajnak model, only the value of dislocation density was arbitrarily determined. The selected value of 10^8 cm^{-2} is a value often observed in literature. This value is also very close to the density calculated from the Johnston-Gilman model at the upper yield point. In this respect, the Johnston-Gilman dislocation multiplication concept appears valid.

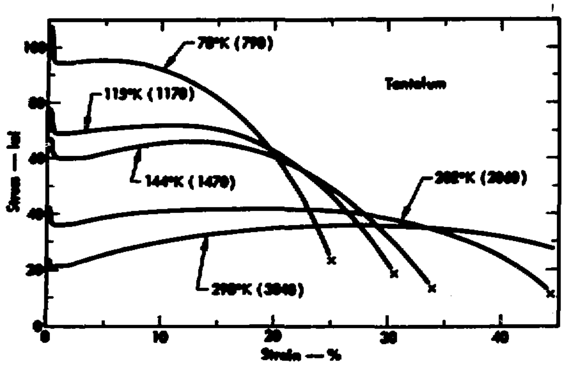

FIG. 4
Effect of temperature on stress-strain curve. Tests were performed at a rate of 10^{-4} sec^{-1}. P values are indicated in parentheses.

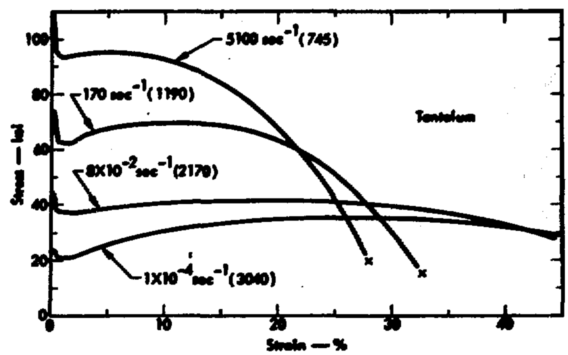

FIG. 5
Effect of strain rate on stress-strain curve. Tests were performed at room temperature. P values are indicated in parentheses.

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References

1. W. G. Johnston and J. J. Gilman, J. Appl. Phys. 30, 129 (1959).
2. J. E. Dorn and S. Rajnak, Trans. Met. Soc. AIME 203, 1052 (1964).
3. K. G. Hoge, J. Basic Eng. 88, 509 (1966).
4. K. G. Hoge, J. Appl. Polymer Sci., Appl. Polymer Symposia 5, 19 (1967).
5. F. E. Hauser, Exp. Mech. 6, 395 (1966).
6. G. T. Hahn, Acta Met. 10, 747 (1962).
7. K. G. Hoge and P. P. Gillis, Lawrence Radiation Laboratory, Livermore, Report UCRL-72316 (1970).
8. C. N. Reid, A. Gilbert, and A. R. Rosenfield, Phil. Mag. 12, 409 (1965).
9. H. L. Dunegan, internal correspondence (1969).
10. P. Guyot and J. E. Dorn, Can. J. Phys. 45, 983 (1967).
11. J. E. Dorn, J. Mitchell, and F. Hauser, Exp. Mech. 5, 357 (1965).
12. A. Kumar and R. G. Kumble, J. Appl. Phys. 40, 3475 (1969).