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SHEAR STRESS PREDICTION IN SHOCK LOADED COPPER

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The stress-strain behavior during the shock rise of a 30 kbar and 54 kbar shock in copper is modeled using a plastic constitutive model that includes rate and temperature dependent hardening and accounts for the transition from thermally activated to viscous drag controlled deformation at high strain rates. A slight modification to the treatment of the mobile dislocation density within the model from that originally proposed leads to better agreement with the shock data than achieved previously. The results indicate that the deformation mechanism during the shock rise is a drag mechanism.

1. INTRODUCTION

At the 1989 APS Topical Conference on Shock Waves in Condensed Matter, Tonks and Johnson[1] compared measurements of shock wave profiles in copper at 30 kbars and 54 kbars (reported earlier as 61 kbars) with predictions of the profiles using a code that included a plastic constitutive model for high strain rate deformation in copper proposed by Follansbee[2]. Two important conclusions regarding the plastic constitutive model and the rate controlling deformation mechanisms in copper during shock deformation resulted from this study. The investigation indicated that the deformation mechanism in copper during shock loading enters the viscous drag regime. Secondly, the rate dependent hardening term in the plastic constitutive model, which was given a linear strain rate dependence in the original derivation[3], over-predicted the hardening behavior; a square root strain rate dependence was found to give a more reasonable estimate of the hardening at the shock strain rates. However, even with this modified hardening term, the predicted behavior did not show the same rate dependence as observed experimentally. In this paper, we show that a minor modification to the plastic constitutive model leads to much closer agreement with the measurements. The modification is to define a stress-dependent mobile dislocation density, which in the original work was assumed to be constant.

2. THE MODEL

Follansbee and Kocks[3] analyzed room temperature deformation in oxygen-free-electronic copper deformed over a strain rate range of 10^{-4} s^{-1} to 10^{4} s^{-1}. The results were interpreted to prove the absence of a transition to viscous drag controlled deformation at strain rates as low as 10^{4} s^{-1}. This transition in deformation mechanism is inevitable with increasing strain rate, however, and Follansbee extended the Follansbee-Kocks model to incorporate this transition[2] according to the method described by Clifton[4] and Klahn, Mukherjee, and Dorn[5].

The governing equation for the variation of the strain rate, $\dot{\epsilon}$, with stress, σ, and temperature, T, developed previously[2] is written as

$$\dot{\epsilon} = \frac{\dot{\epsilon}_o}{\dfrac{MB\lambda\nu_o}{\sigma b} + \exp\left(\dfrac{\Delta G(\sigma/\theta)}{kT}\right)} \qquad (1)$$

where M is a Taylor factor, B is a drag coefficient (given a stress dependence to account for possible relativistic limits), ν_o is the jump or attempt frequency

(10^{11} s^{-1}), λ is the mean distance between obstacles, b is the Burgers vector, k is the Boltzmann constant, ΔG is an activation energy, $\hat{\sigma}$ is the mechanical stress characterizing the intrinsic strength (state) of material, and $\dot{\epsilon}_o = \alpha M \rho_m b \gamma / \upsilon_o$, where α is a constant (0.5) and ρ_m is the mobile dislocation density. In our previous work and in the Tonks and Johnson[1] analysis $\dot{\epsilon}_o$ was taken to be constant (10^7 s^{-1}). This was partly justified by the expectation that with increasing dislocation density the increase in the mobile dislocation density would be offset by the decrease in the mean spacing between obstacles. Because the spacing γ also is found in the first term in the denominator of Eq. 1, calculations were made[1,2] assuming i) γ equal to a constant and ii) $\gamma = \rho^{-1/2} = \alpha\mu b/\hat{\sigma}$, where μ is the shear modulus and ρ is the total dislocation density. The second case is considered to be more realistic.

Equation 1 describes the dependence of strain rate on stress and temperature at a given state. The rate-dependent evolution of the state is described using[3]

$$\frac{d\hat{\sigma}}{d\epsilon} = \theta_o(\dot{\epsilon})\left[1 - F\left(\frac{\hat{\sigma}}{\hat{\sigma}_s}\right)\right] \qquad (2)$$

where θ_o is the stage II hardening rate and $\hat{\sigma}_s$ is the maximum (saturation) value of the state for a given temperature and strain rate. The original work described the rate-dependence of θ_o using an expression which contained a term linear in strain rate. The evidence now suggests that this over estimates the hardening at strain rates exceeding 10^4 s^{-1} and, as in the Tonks and Johnson work[1], we instead use

$$\theta_o,\ MPa = 2373 + 8.3\ \ln\dot{\epsilon} + 3.51\ \sqrt{\dot{\epsilon}} \qquad (3)$$

Equations 1 through 3 give the plastic constitutive equations used by Tonks and Johnson[1] to calculate the shock rise for copper shock deformed at 30 and 54 kbars. However, in this form the model was unable to predict the details of both shock rises with a single value of γ. In particular the computed strain-rate sensitivity over-predicted the measured increase in stress levels when the shock pressure was increased from 30 to 54 kbars. This result calls attention to the assumption of a constant $\dot{\epsilon}_o$, which implies that the product of the mobile dislocation density and the mean dislocation spacing remains constant with increasing stress or state. The expected variation of the mean spacing with state was given above (case ii). The mobile dislocation density is a fraction, typically a small fraction, of the total dislocation density. Although a constant mobile dislocation density is a good approximation when the strain-rate sensitivity is low, at higher rate sensitivities (as in creep deformation) the mobile dislocation density is often assumed to vary with stress according to a power of between 1 and 3[6]. Assuming that this power is 2, the stress dependent mobile dislocation density can be written as

$$\rho_m = \beta\left(\frac{\sigma}{\hat{\sigma}}\right)^2 \rho \qquad (4)$$

Taking $\beta = 0.02$, which gives $\dot{\epsilon}_o$ values close to 10^7 s^{-1} at low strain rates, Figure 1 shows adiabatic stress-strain curves calculated using Eq. 1 with Eq. 4 substituted for ρ_m in $\dot{\epsilon}_o$ and case ii for γ. These curves differ from those computed previously (see Figure 6 in Reference 2) in that the transition between the viscous drag controlled regime at low strains and the thermally activated regime at higher strains is much more gradual in the curves shown in Figure 1. Figure 2 shows the variation of stress at constant strains of $\epsilon = 0.01$ and $\epsilon = 0.10$ versus strain rate. Comparing these curves with the predictions at a strain of $\epsilon = 0.10$ given in Figure 4 of Reference 3 again shows that the addition of a stress-dependent mobile dislocation term leads to a more gradual transition to the viscous drag regime.

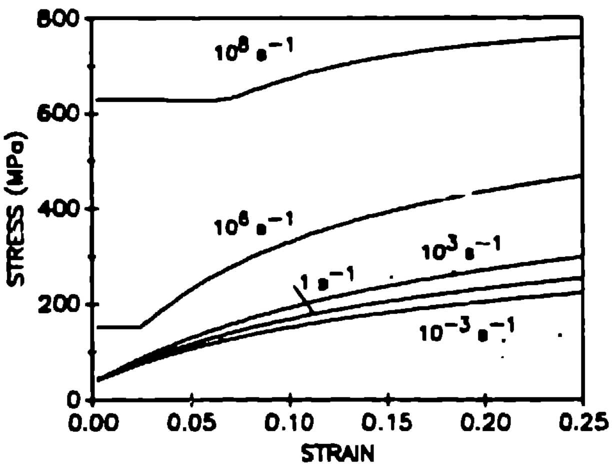

FIGURE 1
Stress-strain curves (adiabatic at $\dot{\epsilon} = 1\ s^{-1}$, $10^3\ s^{-1}$, $10^6\ s^{-1}$, and $10^8\ s^{-1}$, and isothermal at $\dot{\epsilon} = 10^{-3}\ s^{-1}$) as a function of strain rate for an initial temperature of 295K.

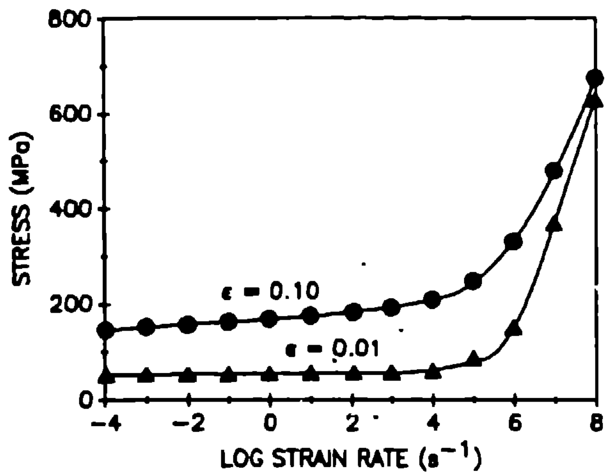

FIGURE 2
Flow stress for deformation at a 295K starting temperature and various strain rates to strains of $\epsilon = 0.01$ and 0.10.

3. CALCULATION OF THE SHOCK RISE IN COPPER

Comparisons are given here between the model predictions and the shock rise profiles for two shock strengths, 30 kbar and 54 kbar The lower pressure shock profile was measured by Wames[7], while the higher pressure profile was reported by Swegle and Grady[8]. The profiles have been analyzed using a steady-wave weak shock analysis by Tonks[9] to give temporal data in the form of plastic strain (ϵ) and deviatoric stress (τ) through the shock rise. These data are plotted in Figure 3. Predictions are made by stepping through the shock in strain increments ($\Delta\epsilon = 0.0001$). At each increment the current shear stress ($\sigma/2$) is calculated using Eq. 1, modified with Eq. 4, and the incremental change in the state is calculated using Eqs. 2 and 3. The predictions, shown in Figure 4 along with the analyzed data, compare favorably with the data, particularly at 30 kbars. The hardening at low strains is slightly underestimated but the peak stresses are calculated reasonably well for both shock pressures.

The stress levels predicted at the end of the shock rise remain above the analyzed data for both shock pressures. However, the reported final stress levels are lower than anticipated. For example, the stress level for room temperature deformation at a uniform strain rate of $10^4\ s^{-1}$ to a strain of 2.5% (which represents the final strain rate and the maximum strain at the end of the 54 kbar shock rise) is roughly 93 MPa. The predicted final stress at the end of the shock rise is 164 MPa, whereas the value reported by Tonks is 116 MPa[9]. The predicted value for the shock exceeds the predicted value for a constant strain rate of $10^4\ s^{-1}$ because most of the shock rise is at a higher strain rate, where, according to Eq. 3, the hardening is higher. The disagreement between the predicted and the measured final stress may indicate that the rate-dependence implied in Eq. 3 is still too high. However, the accuracy of the analyzed shock profiles may not warrant this fine an interpretation.

4. SUMMARY

A simple modification to the plastic constitutive model proposed earlier[2,3] leads to more accurate predictions than found previously for the stress-strain behavior during weak shock loading. The modification

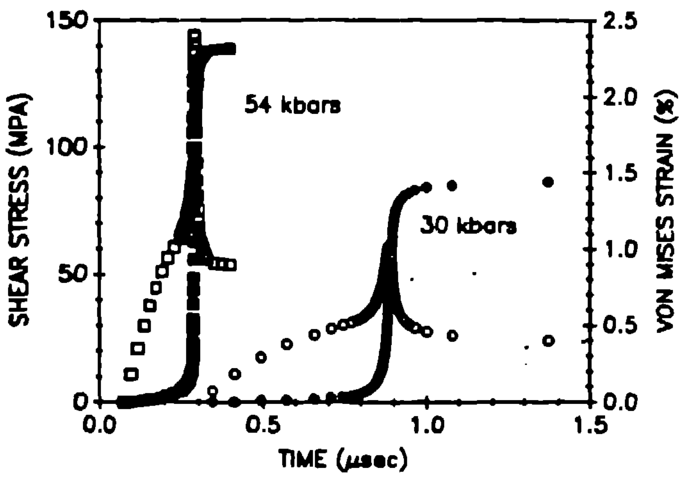

FIGURE 3
Analyzed[9] shear stress and octahedral strain profiles through the 30 kbar[7] and 54 kbar[8] shock rises.

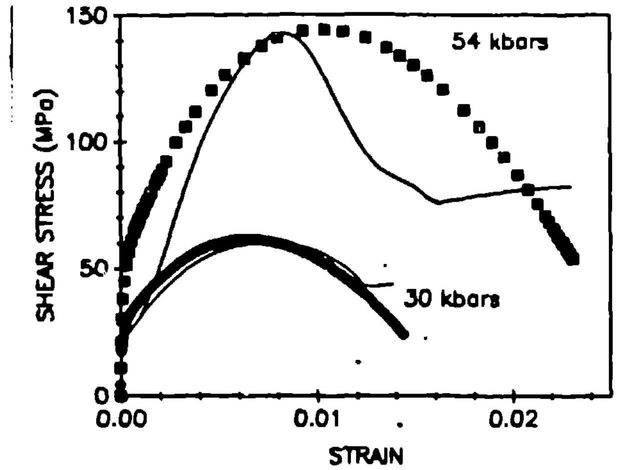

FIGURE 4
Predicted stress (τ) strain profiles (solid lines) for the 30 kbar and 54 kbar shock rises compared to the analyzed[9] profiles (data points).

allows for a mildly stress dependent mobile dislocation density, which does not greatly affect the predicted behavior at low strain rates (< 10^4 s^{-1}) but which introduces the correct strain-rate dependence at strain rates found in the shock rise. The two conclusions from the Tonks and Johnson[1] work regarding plastic constitutive behavior during shock deformation are not altered. That is, the deformation mechanism during shock rise at 30 and 54 kbar is still found to be viscous drag and the linearly rate-dependent stage II hardening proposed in the original Follansbee and Kocks[3] work appears to be too high.

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