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AUTHOR(S) J.N. Johnson
R.S. Hixson
D.L. Tonks
G. T. Gray III

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Los Alamos National Laboratory
Los Alamos, New Mexico 87545

SHOCK COMPRESSION AND QUASIELASTIC RELEASE IN TANTALUM

J. N. Johnson, R. S. Hixson, D. L. Tonks, and G. T. Gray III
Los Alamos National Laboratory
Los Alamos, New Mexico, USA 87545

Previous studies of quasielastic release in shock-loaded FCC metals have shown a strong influence of the defect state on the leading edge, or first observable arrival, of the release wave. This is due to the large density of pinned dislocation segments behind the shock front, their relatively large pinning separation, and a very short response time as determined by the drag coefficient in the shock-compressed state. This effect is entirely equivalent to problems associated with elastic moduli determination using ultrasonic methods. This is particularly true for FCC metals, which have an especially low Peierls stress, or inherent lattice resistance, that has little influence in pinning dislocation segments and inhibiting anelastic deformation. BCC metals, on the other hand, have a large Peierls stress that essentially holds dislocation segments in place at low net applied shear stresses and thus allows fully elastic deformation to occur in the complete absence of anelastic behavior. Shock-compression and release experiments have been performed on tantalum (BCC), with the observation that the leading release disturbance is indeed elastic. This conclusion is established by examination of experimental VISAR records taken at the tantalum/sapphire (window) interface in a symmetric-impact experiment which subjects the sample to a peak longitudinal stress of approximately 7.3 GPa, in comparison with characteristic code calculations.

INTRODUCTION

One of the unique aspects of shock compression science is the ability to make very accurate time-resolved (~1 ns) optical measurements of particle velocity at a position on the surface of an impacted solid sample. If the sample material is transparent these measurements can be made *in situ*. More often, as in the case of metals, measurements are made at the planar interface between the sample and a transparent window material chosen to provide a reasonably good acoustic impedance match with the sample.

These data contain much information concerning variation of elastic moduli with compression, the onset and continuation of high-rate plastic flow, and quasielastic release from the shocked state.

The term "quasielastic" comes from the common observation that release waves in shocked metals (and perhaps most nonmetals) do not exhibit ideal elastic plastic behavior.

The initial treatment of elastic plastic deformation in shock loaded solids was based on rate independent, ideally plastic deformation. However, it was soon recognized that this description was unable to account for time dependent effects, particularly the complex release wave properties observed in almost all, if not all, metals. Measured release waves exhibit a great deal of anelastic, or quasielastic, structure prior to fully plastic, reverse yielding when the stress state encounters the yield surface on the opposite side of the shock hydrostat.

Analysis of quasielastic release in shock compressed solids involves quantifying microstructural effects such as pinned dislocation segments and dislocation pile ups. This includes the evolution of internal stresses that control the motion of these defects as the applied stress is released in the unloading wave propagating into the shocked state [1-4]. In previous work [4] several conclusions are reached in regard to the type of defects related to quasielastic release wave behavior in FCC metals. The first is that pinned segments are responsible for the reverse inelastic deformation observed in the quasielastic regime, and secondly the distance L between pinning points is ~ 500 b, where b the lattice spacing (or magnitude of the Burgers vector). The line density of pinned dislocation segments is on the order of 10^{-5} b^{-2} (approximately 10^{10} cm^{-2}). These observations seem reasonable enough, but the inferred value of the viscous drag coefficient B in the shock-compressed state is unusual. For 6061-T6 aluminum, oxygen-free electronic (OFE) copper, and a silicon-bronze alloy (all shock loaded to 10 - 20 GPa) it is found that the viscous drag coefficient must be approximately 1 dyne s cm^{-2} in order for calculated quasielastic release waves to be in even approximate agreement with measurement. This is three orders of magnitude greater than experimentally observed under ambient conditions [5]. These results pertain explicitly to the assumption that the leading observable release disturbance is fully elastic.

Preliminary molecular dynamics studies [6] suggest that the viscous drag coefficient in copper is essentially unchanged by compression to 10 - 20 GPa. This information served as a strong impetus to investigate quasielastic behavior related to lower values (0.001 - 0.1 dyne s cm^{-2}) of the viscous drag coefficient in FCC metals [6]. The results given in reference [7] show that for B < 0.1 dyne s cm^{-2} the leading observable release wave disturbance in FCC metals does not propagate with the elastic wave speed c_e given by,

$$\rho c_e^2 = K + \frac{4}{3}G , \qquad (1)$$

but with the "relaxed" velocity c_r defined by

$$\rho c_r^2 = K + \frac{4G/3}{1 + (\alpha L^2/4)} , \qquad (2)$$

where ρ is the material density, K and G are the adiabatic bulk and shear moduli, respectively, and nL^2 is the product of density (length per unit volume) of pinned segments and the average pinning separation squared. All of the quantities in Equations (1) and (2) are evaluated in the shock compressed state ahead of the release wave.

For normal values of B, the shear modulus determined from release wave measurements corresponds to the elastic shear modulus G(r) only when $nL^2 << 1$; this is also one of the conditions required of ultrasonic measurement in the accurate determination of elastic moduli [8,9]. Unfortunately, experimentalists have little or no control over nL^2 behind the shock front in shock release experiments.

The specific results obtained so far apply only to FCC metals, for which the Peierls stress provided by the lattice is usually quite small. Other solids may possibly possess sufficient natural resistance to reverse motion of dislocation segments to insure that the initial release from the shocked state is fully elastic.

This brings us to the question of whether or not the leading observable release wave in tantalum, a typical BCC metal, propagates with the fully elastic speed, in contrast with results obtained for FCC metals. The answer to that question is yes, the leading portion of the release wave in tantalum does propagate with the fully elastic wave speed. This is a consequence of the effective pinning of bowed out dislocation segments provided by the finite Peierls stress of the lattice.

MICROMECHANICAL SOURCE OF ANELASTICITY IN THE RELEASE WAVE

The quasielastic nature of release waves in aluminum, copper, and Si bronze comes from internal stresses acting on dislocations due to their curvature between pinning points provided by lattice impurities, point defects, or other dislocations. This is shown schematically in Figure 1. The relationship between radius of curvature R and applied shear stress τ in the shock compressed state is $R = Gb/\tau$ under equilibrium conditions. In the transient situation the back stress β given by

$$\beta = Gb/R \tag{3}$$

is not in equilibrium with τ, but obeys the evolutionary law given by [4]

$$\dot{\beta} = \frac{(1/GB)\beta^2(\tau - \beta)}{[1 - (\beta L/2Gb)^2]^{1/2} - 1} \tag{4}$$

The release of applied stress allows reverse dislocation motion to occur immediately upon arrival of the unloading wave. There are two characteristic times associated with the accumulation of reverse plastic strain [4]: $(B/G)(L/b)^2$ and $B/(nb^2)$.

The first time constant controls the readjustment of the internal stress β due to dislocation curvature and the second controls the rate of accumulation of plastic strain. Because of the small values of B and L/b (~500), the first time constant is very small (~1 ns) and the readjustment of the dislocation segments takes place almost instantaneously, even in comparison to the best VISAR time resolution currently available. This means that the leading observable release disturbance in FCC metals is not fully elastic. Measured release wave speeds are related to an effective elastic shear modulus given by

$$G_{eff} = \frac{G}{1 + nL^2/4} \tag{5}$$

For pure materials, $nL^2 \sim 1$ and the shear modulus, as determined by release wave speeds, is a few tens of percent lower than the true elastic shear modulus G.

As an aside, the above expression for the effective shear modulus suggests a means by which n can be determined in the shock compressed state, once L, G, and G_{eff} are known.

Figures 2 and 3 show two calculations for 6061-T6 aluminum [10], each with B = 0.1 dyne s cm^{-2}. Figure 2 has a value of $nb^2 = 2.5 \times 10^{-5}$, and Figure 3 is for $nb^2 = 0$. The latter case shows the effect of "freezing out" the anelastic contribution from the pinned dislocation segments. The elastic shear modulus in the shock compressed state is taken to be 92 GPa. The vertical arrows indicate the first observable arrival of the quasielastic release wave in each case. It is clear from these results that the presence of mobile dislocation segments has a strong influence on the first arrival of the release wave. The effective shear modulus as obtained from first arrival information alone is G = 65 GPa.

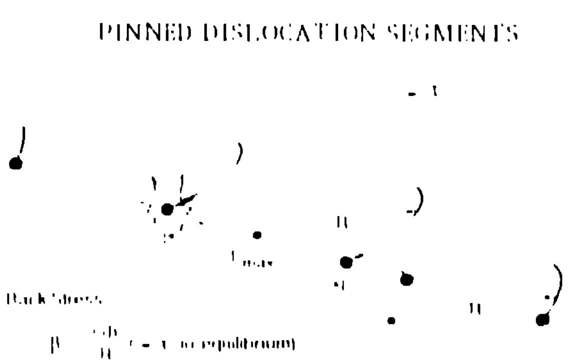

Figure 1. Pinned dislocation segments under the influence of an applied shear stress τ.

RESULTS FOR TANTALUM

The quasielastic release wave in shock loaded tantalum was studied using a 50 mm diameter gas gun. Projectile velocity and tilt were measured immediately before impact by means of a stepped circular array of shorting pins surrounding the target disk. For this experiment the tilt was

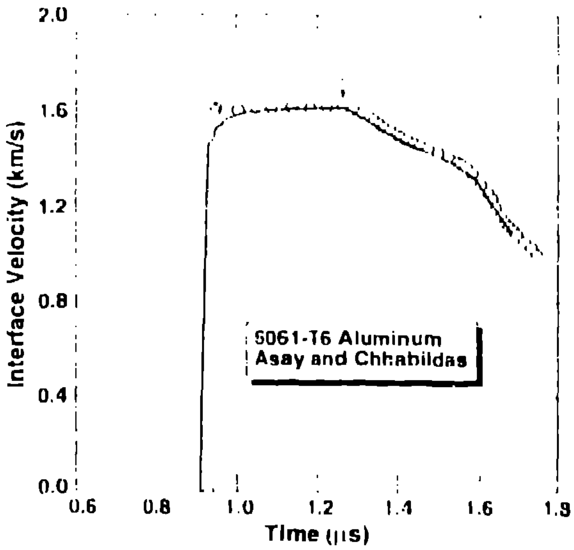

Figure 2. Measured (circles) and calculated (line) wave profiles in 6061-T6 aluminum; $nb^2 = 2.5 \times 10^{-5}$.

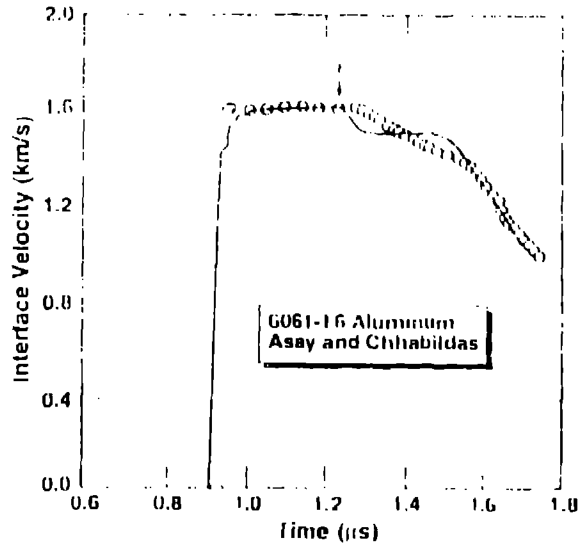

Figure 3. Measured (circles) and calculated (line) wave profiles in 6061-T6 aluminum; $nb^2 = 0$.

approximately 1.0-1.5 mrad, and the impact velocity was very close to 0.256 km/s. Symmetric impact was used to generate an initial 7.3 GPa shock followed by release.

The shock-release profile was measured using a push/pull VISAR [11]. The wave profile was measured at the target-window interface with a Z-cut sapphire window. This target geometry minimized, to the extent possible, hydrodynamic perturbations at the target-window interface.

These data are shown in Figures 4 and 5 in comparison with numerical simulations of this experiment using characteristic methods [4]. Figure 4 shows the effect of quasielastic release with ($nb^2 = 5 \times 10^{-5}$). Figure 5 shows material behavior in the absence of anelasticity in the release wave ($nb^2 = 0$). The viscous drag coefficient is taken to be $B = 0.1$ dyne s cm^{-2} in both cases. The influence of the Peierls stress is taken into account by assuming the dislocation velocity goes very rapidly to zero when the absolute magnitude of the shear stress falls below 0.45 GPa [12]. This is the effect that is responsible for pinning the dislocation segments as unloading begins in the release wave.

The results presented here show the quasielastic nature of the release wave in tantalum and the improved fit to the data when this is taken into account in way described previously [4]. However, it is seen that the first arrival time of the release wave is unaffected by the presence of pinned dislocation segments in the shock compressed state. This is consistent with the existence of a strong lattice resistance (Peierls force) to dislocation motion in BCC materials.

The micromechanical model of pinned dislocation segments for FCC metals has been used in this calculation for tantalum. It is recognized that the shape of expanding dislocation loops in FCC and BCC metals is likely different, and that dislocation storage effects differ greatly between these two crystal structures [13]. However, the essential picture of bowed dislocation segments remains approximately the same. While the fundamental mechanism for generating back stress is unchanged, the effect of varied dislocation morphology will alter global slip activity from areal to lineal glide [14]. This rationale is consistent with substructure observations that show residual long, straight screw dislocations after deformation in tantalum at low temperatures or high strain rates [15].

The elastic properties of tantalum have been described in two ways. The first is the small anisotropy approximation [16], in which the isotropic bulk and shear moduli are assumed to be functions of density. In this approximation the bulk sound speed $c_0 = 3.43$ cm/μs, the slope of the shock velocity/particle velocity relationship is $s = 1.19$, and Poisson's ratio is assumed constant at $\nu = 0.331$ [17]. Because of the relatively large Hugoniot elastic limit in comparison to the peak longitudinal stress of 7.3 GPa (Figures 4 and 5), the small anisotropy approximation may not be as valid as it is for aluminum at 21.7 GPa (Figures 2 and 3).

For this reason we also investigated the effect of using weak shock analysis [16,18] involving second- and third-order elastic moduli. This will provide a check on the reasonableness of the small anisotropy approximation, and will be further confirmation of fully elastic behavior in the initial portion of the release wave. The independent isotropic elastic moduli used in these calculations are given by

$$C_{11} = 289.7 \text{ GPa}, \qquad C_{12} = 145.7 \text{ GPa},$$

$$C_{111} = 1622 \text{ GPa}, \; C_{112} = 1100 \text{ GPa}, \; C_{123} = 838.9 \text{ GPa},$$

based on calculations using the zero pressure bulk sound speed [19] in conjunction with pressure and temperature

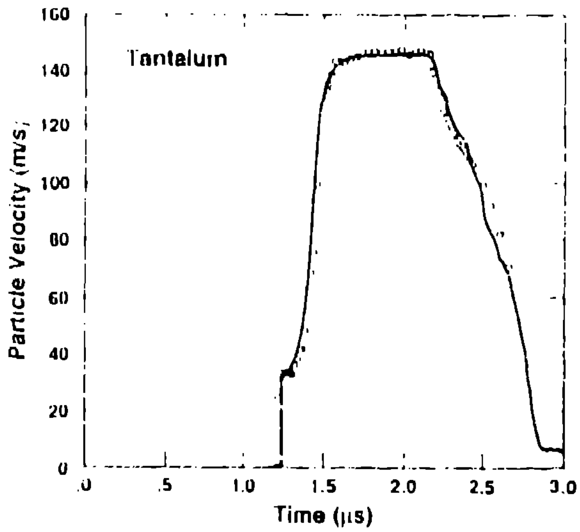

Figure 4. Measured (circles) and calculated (line) wave profiles in tantalum: $nb^2 = 5 \times 10^{-5}$.

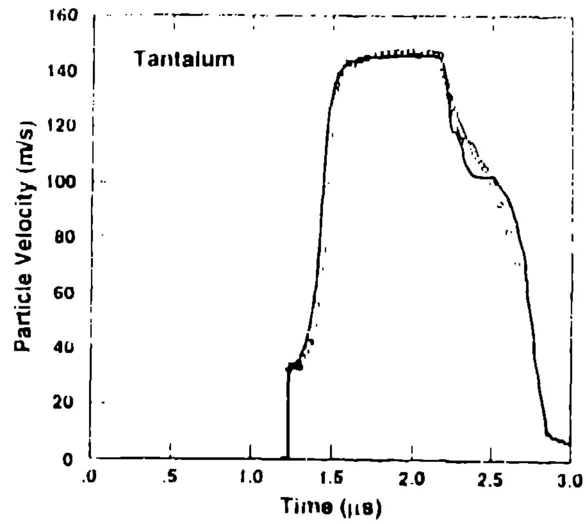

Figure 5. Measured (circles) and calculated (line) wave profiles in tantalum: $nb^2 = 0$.

derivatives of the the bulk and shear moduli [20,21]. Differences between wave profiles (Figures 4 and 5) using either the small anisotropy approximation or the weak-shock analysis are negligible.

SUMMARY AND CONCLUSIONS

Previous work showed that release waves in FCC metals were quasielastic in nature, and that the leading observable release disturbance propagated with a speed controlled by an effective shear modulus defined by $G_{eff} = G/[1 + nL^2/4]$, where G is the fully elastic shear modulus, n is the density of mobile dislocation segments behind the shock front, and L is their pinning separation. Experimentalists have little or no control over n and L in a shock/release cycle, and for this reason release wave arrival times may be of questionable use in obtaining elastic moduli in the shocked state for FCC metals.

For BCC metals, in particular tantalum, the finite Peierls stress serves as an effective additional pinning mechanism that prohibits reverse dislocation motion immediately upon release from the shocked state. The leading observable release wave disturbance in tantalum correctly samples fully elastic material response.

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