ANELASTIC RELEASE FROM THE SHOCK-COMPRESSED STATE

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ANELASTIC RELEASE FROM THE SHOCK-COMPRESSED STATE

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Abstract - Anelastic release from the shock-compressed state is analyzed in terms of internal self stresses acting on dislocation pile-ups and pinned dislocation loops. Reverse anelastic deformation occurs immediately upon release due to these internal stresses and is responsible for departure from ideal elastic-plastic unloading wave behavior: anelastic release is commonly observed in all shock-loaded metals. The rate at which this readjustment of internal stress occurs is determined by the viscous drag coefficient in the shock-compressed state and has important consequences regarding determination of elastic moduli from unloading wave data.

1.- Introduction

A common experimental observation is that unloading waves from a shock-compressed state in metals never exhibit ideal elastic-plastic response; that is, the transition from elastic unloading to fully plastic unloading is always observed to be a gradual process. We call this process anelastic to distinguish it from ideal elastic-plastic behavior. To illustrate these ideas, we imagine a finite-duration shock pulse arriving at some point in the material. The strain as a function of time is shown in figure 1 for ideal elastic-plastic response (solid line) and the anelastic response (dash-dot line) generally observed. The maximum volume strain $\epsilon = 1 - \rho_0/\rho$ is designated ϵ_0.

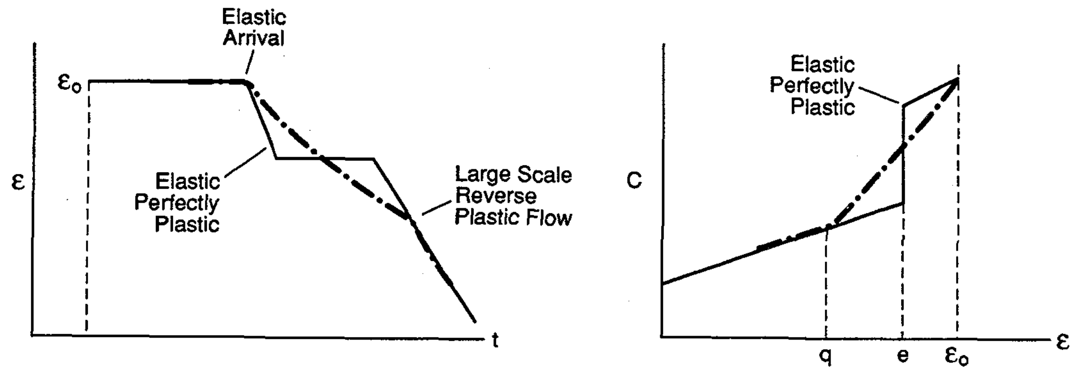

Fig. 1 - Anelastic release-wave profile.

When these observations are translated into Lagrangian wave speed, the data appear as shown in the diagram on the right of figure 1. Points e and q represent volume strains at which ideal elastic-plastic release (e) and anelastic release (q) would undergo transition to large-scale, reverse plastic flow (the normal reverse yield point). In this paper a physical explanation for anelastic release is proposed: this is based upon internal self stresses resulting from dislocation pile-ups and pinned dislocation segments in the shocked state. Calculated anelastic unloading wave speeds /1/ are compared with data for 6061-T6 aluminum shock-loaded to 20.7 GPa /2/, and the influence of the viscous drag coefficient in the shocked state on data interpretation is critically examined.

2.- Lagrangian Unloading Wave Speed

The Lagrangian unloading wave speed C is obtained in the following heuristic way. The bulk and shear moduli in the shocked state are K and G, respectively. The longitudinal stress σ in a planar shock wave is related to the mean stress p and the maximum resolved shear stress τ according to

$$\sigma = p + (4/3)\tau \quad . \tag{1}$$

The Eulerian sound speed c is then given by

$$\rho c^2 = \rho(d\sigma/d\rho) = \rho(dp/d\rho) + (4\rho/3)(d\tau/d\rho) \quad , \tag{2}$$

where ρ is the material density. The shear stress τ is governed by

$$\dot{\tau} = G(\dot{\rho}/\rho - \dot{\gamma}) \quad , \tag{3}$$

where $\dot{\gamma}$ is the anelastic (plastic) strain rate, which is related to the micromechanics of internal stresses acting on dislocations. For $K = \rho(dp/d\rho)$ and $\rho_0 C = \rho c$, eq. (2) becomes

$$\rho_0 C^2 = (\rho/\rho_0)[K + (4G/3)(1 - \rho\dot{\gamma}/\dot{\rho})] \quad . \tag{4}$$

3.- Calculated Wave Speeds

Johnson, Lomdahl, and Wills /1/ perform extensive calculations of C for 6061-T6 aluminum shocked to 20.7 GPa and compare results with data of Asay and Chhabildas /2/. These calculations are carried out for straight dislocation pile-ups and

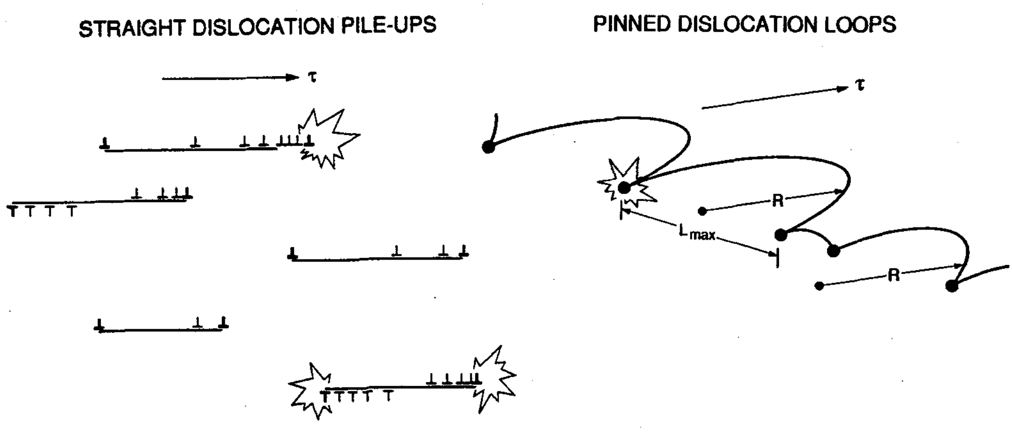

Fig. 2.- Sources of internal stress.

pinned dislocation loops as shown in figure 2. The shear stress in the shocked state is controlled by the <u>weakest</u> pinning points (as shown by the bursts in figure 2). Behind the shock front numerous stressed pile-ups and pinned loops remain. When the release wave arrives and the applied shear stress begins to decrease, internal stresses immediately drive the pinned and piled-up dislocations in the reverse direction, thus contributing to the anelastic strain rate $d\gamma/dt$. This is the source of anelastic release-wave behavior. For straight dislocation pile-ups

$$\dot{\gamma}_S = (b^2N/B)(\tau - \beta_S), \tag{5}$$

where b is the magnitude of the Burgers vector, N is the density of piled-up straight dislocations, B is the viscous drag coefficient, and β_S in the back stress provided by the pinning points.

For pinned dislocation loops

$$\dot{\gamma}_P = (b^2N/LB)(dA/dh)(\tau - \beta_P) \quad , \tag{6}$$

where β_P is the back stress due to dislocation curvature and line tension, L is the distance between pinning points, A is the area swept out by the bowed loop (as a function of the applied stress τ), and h is the maximum displacement of the bowed loop from its (straight) equilibrium position.

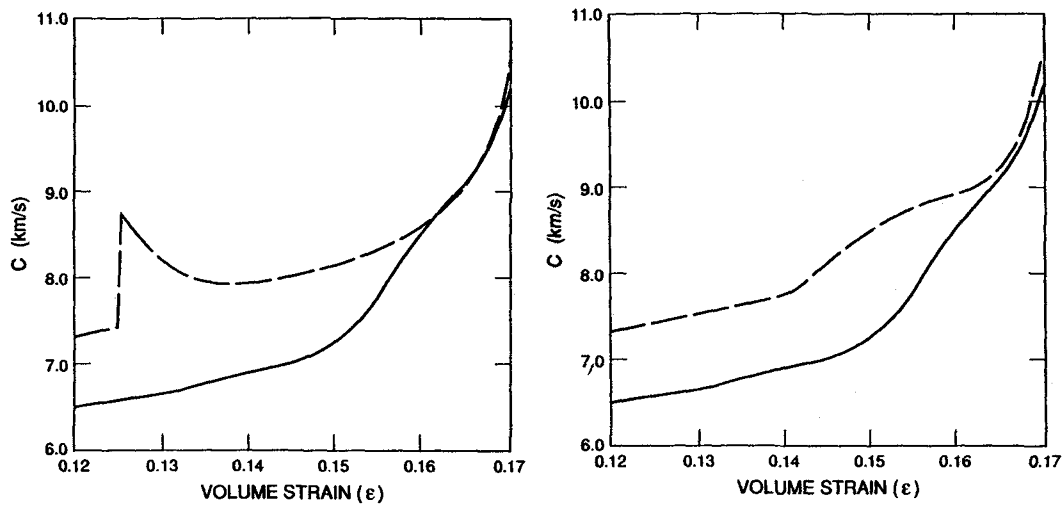

Fig. 3.- Calculated and measured Lagrangian sound speeds: pile-up (left); pinned loops (right).

Calculated wave speeds are shown (dashed line) in figure 3 for straight dislocation pile-ups and pinned loops in comparison with unloading wave data (solid line) for 6061-T6 aluminum /2/. The material parameters used in these calculations are $L/b = 500$, $N \simeq 10^{15}\ m^{-2}$, and $B \simeq 0.1$ Pa s. A constant value of $(1/\rho)d\rho/dt = -10^5\ s^{-1}$ is also assumed. Values of L/b and N are quite reasonable, but the value of B is approximately two orders of magnitude greater than for pure aluminum under uncompressed, room-temperature conditions /3/. When calculations are performed with $B \simeq 10^{-3}$ Pa s, it is found that there is a <u>very</u> rapid readjustment of internal stress such that C decreases rapidly without substantial change in ϵ. Figure 4 shows this effect in the vicinity of $\epsilon_0 = 0.17$ for unloading from a 20.7 GPa shock in 6061-T6 aluminum. The material parameters used in this calculation are $L/b = 1000$, $N \simeq 10^{14}\ m^{-2}$, and $B \simeq 10^{-3}$ Pa s for straight dislocation pile-ups. The initial value of C = 10.54 km/s at ϵ_0 = 0.17 corresponds to bulk and shear moduli in the shocked state (T = 590 K and σ = 20.7 GPa) of K = 160 GPa and G = 67 GPa /1/.

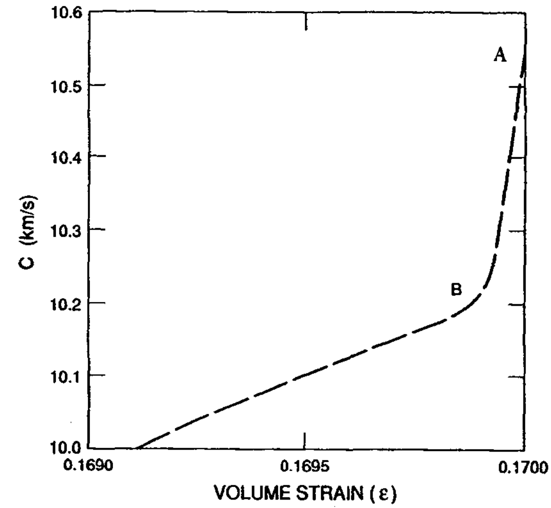

Fig. 4.- Lagrangian wave speed near ϵ_0 = 0.17 for B ≃ 10^{-3} Pa s.

The very rapid drop in C near ϵ = 0.17 would be difficult to observe experimentally. If such a phenomenon were to occur, the data would be interpreted as giving C ≃ 10.2 km/s: it would be unlikely that the very small drop in strain from 0.17 to ~ 0.1699 could be discerned in any wave profile measurement. This is the same as saying that the first true arrival (A) of the unloading wave would not be distinguishable. Consequently, the data would suggest point B as the first arrival, and a serious error would be made in determining the longitudinal modulus in the shocked state.

Physically this means that for B ≃ 10^{-3} Pa s straight dislocation pile-ups and pinned loops can adjust almost instantaneously to a reduction in applied stress, with the result that the longitudinal elastic modulus appears to measureably less than its true value. When the drag coefficient is large, the time scale for this readjustment is made longer and one can actually resolve the true first arrival.

A more formal analysis of this phenomenon can be performed as follows. The time rate of change of the back stress acting on straight dislocation pile-ups and pinned loops for $\beta L/Gb \ll 1$ ($\beta = \beta_s$ and β_p) is given by /1/

$$\dot{\beta}_s \cong (8/t_0)(\tau - \beta_s) \quad , \tag{7}$$

$$\dot{\beta}_p \cong [16\pi(1-\nu)/t_0](\tau - \beta_p) \quad , \tag{8}$$

where $t_0 = BL^2/ba$, $a = Gb/2\pi(1-\nu)$, ν = Poisson's ratio (here taken to be 1/3), and L is the linear spatial dimension of the pile-up or distance between pinning points for pinned dislocation segments.

Combination of eqs. (3), (5), and (7) then gives, for straight dislocation pile-ups,

$$\frac{d}{dt}(\tau - \beta_s) = G\dot{\rho}/\rho - (\tau - \beta_s)/t_c \quad , \tag{9}$$

$$t_c \equiv t_0/[8 + 2\pi(1-\nu)NL^2] \quad . \tag{10}$$

A solution to eq. (9) for $\tau - \beta_s = 0$ at $t = 0$ (the start of the unloading process at constant $d\ln\rho/dt \sim -10^5\ s^{-1}$) is

$$\tau - \beta_s = (G\dot{\rho}/\rho)t_c[1 - \exp(-t/t_c)] \quad . \tag{11}$$

Therefore, for small t_c, there is a rapid initial readjustment of $\tau - \beta_s$ which gives a sudden increase in $d\gamma_s/dt$, eq. (5), and a corresponding drop in C, eq. (4). Similar results are obtained for pinned loops /1/.

For $B \sim 10^{-3}$ Pa s, $L/b \sim 10^3$, and $b^2N \sim 10^{-5}$, we have $t_0 \sim 63$ ns and $t_c \sim 1.3$ ns. This very short characteristic time is responsible for the rapid drop in C near $\epsilon = 0.17$ as shown in figure 4. For $B \sim 10^{-1}$ Pa s, $L/b \sim 500$, and $b^2N \sim 10^{-4}$, we have $t_0 \sim 4000$ ns and $t_c \sim 400$ ns: the response time is greatly extended and a rapid decrease in C due to anelasticity does not occur.

4.- Discussion

The anelastic release-wave behavior in shock-compressed solids can be explained in terms of straight dislocation pile-ups and pinned dislocation loops which are created in the shock-loading process. As the applied shear stress is reduced, reverse plastic flow occurs immediately, thus reducing the longitudinal modulus and hence the unloading wave speed.

The calculational results shown in figure 3 suggest that pinned loops are responsible for anelastic release from the shock-compressed state of 20.7 GPa in 6061-T6 aluminum. The multivalued functional dependence shown at the left in figure 3 is an artifact resulting from the constant $(1/\rho)d\rho/dt$ assumption. Such behavior is physically impossible - we cannot have two different values of strain arrive at a measurement location at the same time (that is, with the same wave propagation speed). In the actual dynamic flow process, $(1/\rho)d\rho/dt$ would change as necessary to conserve mass and momentum, and this would have the effect of re-establishing monotonicity in the wave speed. This might be expected to produce considerable structure in the anelastic unloading wave. Nothing of this kind is seen in the original interface data /2/, so we become suspicious of the straight dislocation pile-up as the micromechanical basis for anelastic release in 6061-T6 aluminum.

Another important point is the significant difference between theory and experiment below a volume strain of 0.15: figure 3. This is the fully plastic ($\rho d\gamma/d\rho = 1$) unloading wave which occurs when $\tau \le -\tau_c$ $(=-3\tau_0)$ as determined by Asay and Chhabildas /2/: $\tau_0 = 0.2$ GPa at a shock pressure of 20.7 GPa. If a significant amount of work hardening were occurring, the dashed lines of figure 3 in the fully plastic region would be even higher. Because we have such good equation-of-state data for this material, this variation is difficult to understand. We expect it may have to do with some aspect of the data analysis in going from the original interfacial particle velocity measurement to the wave-speed representation of figure 3.

For viscous drag coefficients on the order of $B \sim 10^{-3}$ Pa s, the characteristic time for readjustment of the internal stress can be ~ 1 ns. This has the effect of reducing the effective longitudinal elastic modulus of unloading and gives rise to potential problems of interpretation in using unloading-wave data from which to infer elastic moduli in the shock-compressed state. This does not seem to be the case for 6061-T6 aluminum shocked to 20.7 GPa /2/, however, and the data suggest $B \sim 10^{-1}$ Pa s.

The fact that $B \sim 10^{-1}$ Pa s is approximately two orders of magnitude greater than that for pure aluminum under uncompressed, room-temperature conditions is an unusual and unexpected result. We can estimate the increase in drag coefficient from the results of Clifton, Gilat, and Li /4/, Leibfried /5/, and Hirth and Lothe /6/. Under the assumption that b^3 scales as ρ_0/ρ we find

$$B/B_0 = (T/T_0)(G_0/G)^{1/2}(\rho/\rho_0)^{7/6} \quad , \tag{12}$$

with $T_0 = 300$ K, $T = 590$ K, $G_0 = 27.6$ GPa, $G = 67.0$ GPa, and $\rho/\rho_0 = 0.83$, we obtain $B/B_0 = 1.57$. We should thus expect B to be increased somewhat in relation to the ambient value, but not by two orders of magnitude. This remains a subject of investigation.

The recognition that internal stresses can result in measureable reverse plastic deformation immediately upon release from the shock compressed state explains the long-standing observation that metals do not exhibit ideal elastic-plastic unloading. We are not only able to answer this question, but are able to use unloading-wave profiles to extract additional micromechanical information concerning conditions in the shocked state.

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