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DISCUSSION OF

PRESSURE-SHEAR IMPACT AND THE DYNAMIC VISCOPLASTIC RESPONSE OF METALS

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Specific comments on paper

The oblique parallel-plate impact experiments conducted by Clifton and co-workers over the past years have yielded relevant information on the propagation of plastic waves in metals; they have made possible the determination of the plastic response of metals at strain rates not attainable by the split Hopkinson (Kolsky) bar ($\sim 10^4$ s^{-1}) and below those in shock-wave deformation (10^8 s^{-1}). The results obtained from these experiments should be of importance as input to computer codes simulating dynamic deformation. On the more fundamental side, the mechanical response of metals at these high strain rates can be well characterized and the link-up between macro and micromechanics performed by Clifton and co-workers reveals, quantitatively, the behavior of dislocations when the applied strain rate is in the range 10^5–10^7 s^{-1}. The oblique parallel-plate impact experiments are a spin-off (or a development) of normal parallel-plate experiments which had as a primary objective the determination of the equation of state of materials. These equations of state were required in the hydrodynamic treatment of shock waves and are an important component in 'hydrocodes'.

However, a great portion of the plastic deformation taking place in armor penetration and impact, and warhead expansion and fragmentation occurs under non-uniaxial strain conditions required for shock loading. Thus, the propagation of non-shock disturbances is of great importance. Fig. 1, from earlier work of Clifton and co-workers (1976), shows a typical plot of the transverse displacement at the back of an 6061-T6 aluminum alloy target as a function of time. At time 2.05 μs, transverse displacement is initiated. This displacement (perpendicular to the direction of propagation of the disturbance) varies linearly with time and defines the characteristics of propagation of the plastic shear wave. For this particular case, the velocities of propagation of the pressure front and shear front were, respectively, 5.24 mm/μs and 3.10 mm/μs. The earlier work of Clifton and co-workers (shown in Fig. 1) was followed by experiments aimed at establishing fundamental parameters of dislocation motion. From these fundamental parameters (such as the viscous drag coefficient B) and the consideration of the Taylor–Bishop–Hill minimum requirements for five independent slip systems, the dynamic shear stress–shear strain curves can be correctly ob-

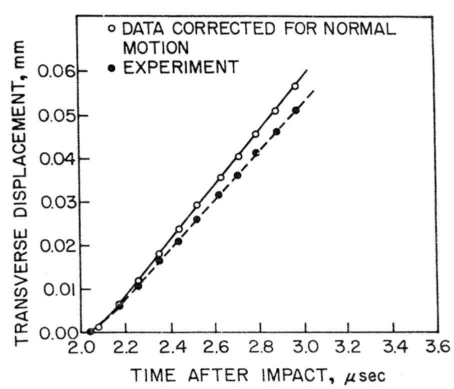

Fig. 1. Propagation of a plastic shear wave; from (Abou-Sayed and Clifton, 1976).

tained independently. This work led naturally to the analysis of the mechanisms responsible for opposing dislocation motion at the high imposed strain rates. Different materials exhibited different responses. Klopp, Clifton and Shawky found that for relativity pure FCC metals the response was different than for BCC (α) iron; a high-strength steel, on the other hand, exhibited only a very weak strain-rate dependence up to 10^5 s^{-1}. Their interpretation of these different responses is somewhat arbitrary. They consider only two rate controlling mechanisms: thermal activation and viscous drag. The term viscous drag is vague; it is composed of electron viscosity, phonon drag, and other mechanisms. The overcoming of barriers by thermal activation is an interesting concept at low strain rates, but this discusser would like to introduce an additional consideration. The thermal activation theory is based on the attempt frequency for overcoming barriers. At these high imposed strain rates the vibrational frequency of dislocations and their velocity of motion could produce interferences which would substantially modify the configuration. Taking, as a first approximation, the vibrational frequency of a dislocation as 10^{11} s^{-1} and the amplitude of vibration equal to 30 Å, one could define an average velocity of motion of the dislocation, in the unstressed state, of:

$$v = A\nu = 30 \times 10^{-10} \times 10^{11} = 300 \text{ m/s}.$$

At these velocities one would expect interferences between thermal activation and motion of a kind not considered in the classical treatments.

An important aspect of dislocation dynamics not considered by Clifton and co-workers is the relativistic effect. Eshelby (1949) originated and Weertman and Weertman (1980) developed the relativistic theory of dislocation dynamics, analogous to the relativity theory of Einstein (the velocity of light is replaced by the velocity of shear waves). According to this treatment the self-energy of a dislocation approaches infinity as the velocities of elastic shear and longitudinal waves are approached. The effect of this increased energy is to require higher stresses for the motion of the dislocation. Relativistic effects might play an important role in the strain-rate dependence of flow

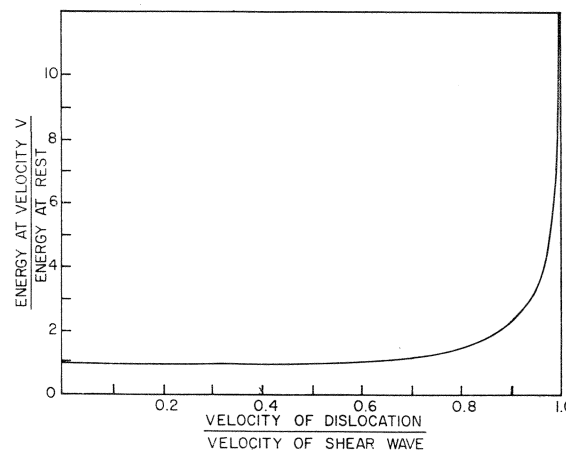

Fig. 2. Variation of dislocation self-energy with velocity; from (Meyers and Chawla, 1984).

stress observed by Clifton and co-workers and should be considered in the overall model. Figure 2 shows the energy of a dislocation line as a function of its velocity. It can be clearly seen that the energy tends to infinity as the velocity approaches that of the elastic shear waves.

2. General comments

The work conducted by Clifton and co-workers addresses one specific aspect of a broad spectrum of inter-related phenomena which comprise different dynamic deformation systems. An example of such a dynamic deformation system is shown in Fig. 3. It shows a target impacted by a projectile. Somewhat arbitrarily, the different phenomena of importance were identified in Fig. 3 (Meyers and Aimone, 1983). Shock-wave strengthening, shear and longitudinal plastic wave propagation, adiabatic shear band formation, dynamic fracture, spalling are phenomena involved in the overall process. For all these phenomena, the micromechanical aspects are of great importance and theory has failed, to the present moment, to have a really predictive capability. In effect, dislocation theory explains why different metals deform in different ways, but has often failed to predict responses under different imposed constraints.

The discusser believes that careful microstruct-

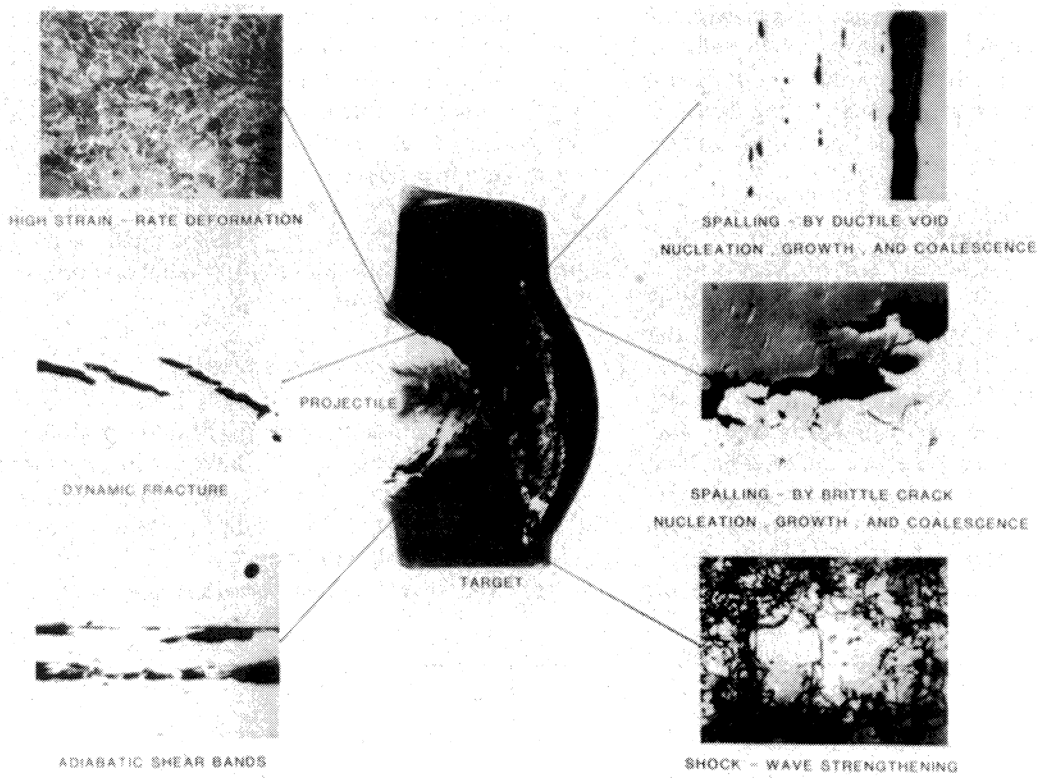

Fig. 3. Dynamic deformation system; from (Meyers and Aimone, 1983).

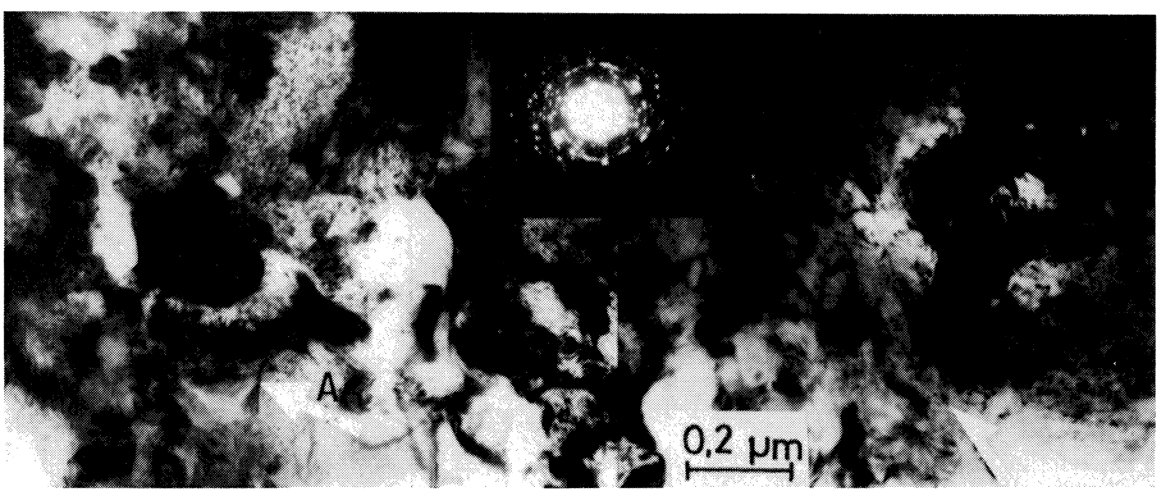

Fig. 4. Adiabatic shear band in titanium; shear band region is microcrystalline; high-voltage transmission electron microscopy; from (Meyers and Pak, 1984).

ural observations under experimental conditions established by theoreticians (in order to validate and expand their mechanistic understanding) are needed and will be of great benefit in the development of constitutive equations. Some ideas will be reviewed below.

2.1. Adiabatic shear bands

Recent observations by Meyers and Pak (1984) at the high voltage electron microscope of the National Center for Electron Microscopy (Lawrence Berkeley Laboratory) revealed the detailed structure of the adiabatic shear bands and showed them to be composed of micro-crystalline grains (Fig. 4). This is indicative of high temperatures in the shear band area. The stress–strain response of the material inside the band area is much better represented by an adiabatic than by an isothermal curve, leading to softening. This shear instability could be incorporated into models, if they had a strain-rate dependence (required for heat-transfer purposes). Figure 5 shows both an isothermal and an adiabatic curve. The adiabatic curve (produced by heat evolution due to plastic deformation) shows the onset of softening at a plastic strain of γ_{ins}. This should have a profound effect on the plastic deformation response of the material.

2.2. Shock-wave propagation in rocks

Models describing the propagation of high-amplitude disturbances in rock have to incorporate the time-dependency of the fragmentation process. Recent experiments by Aimone, Meyers and Mojtabai (1984) showed that:

(a) The mean fragment size decreases with in-

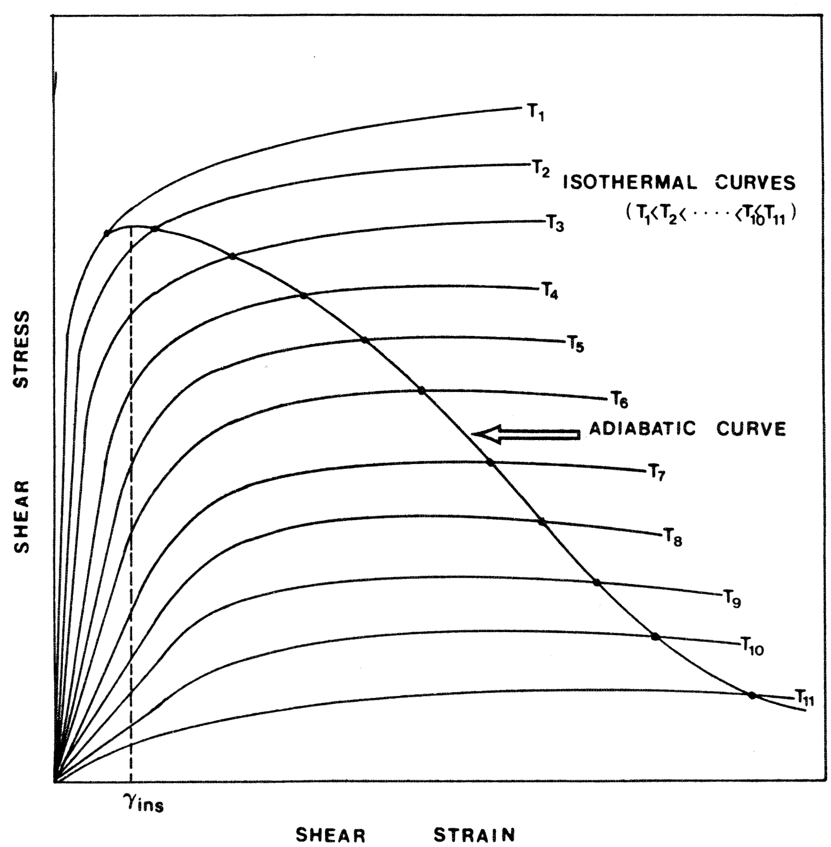

Fig. 5. Isothermal and adiabatic stress–strain curves for a material.

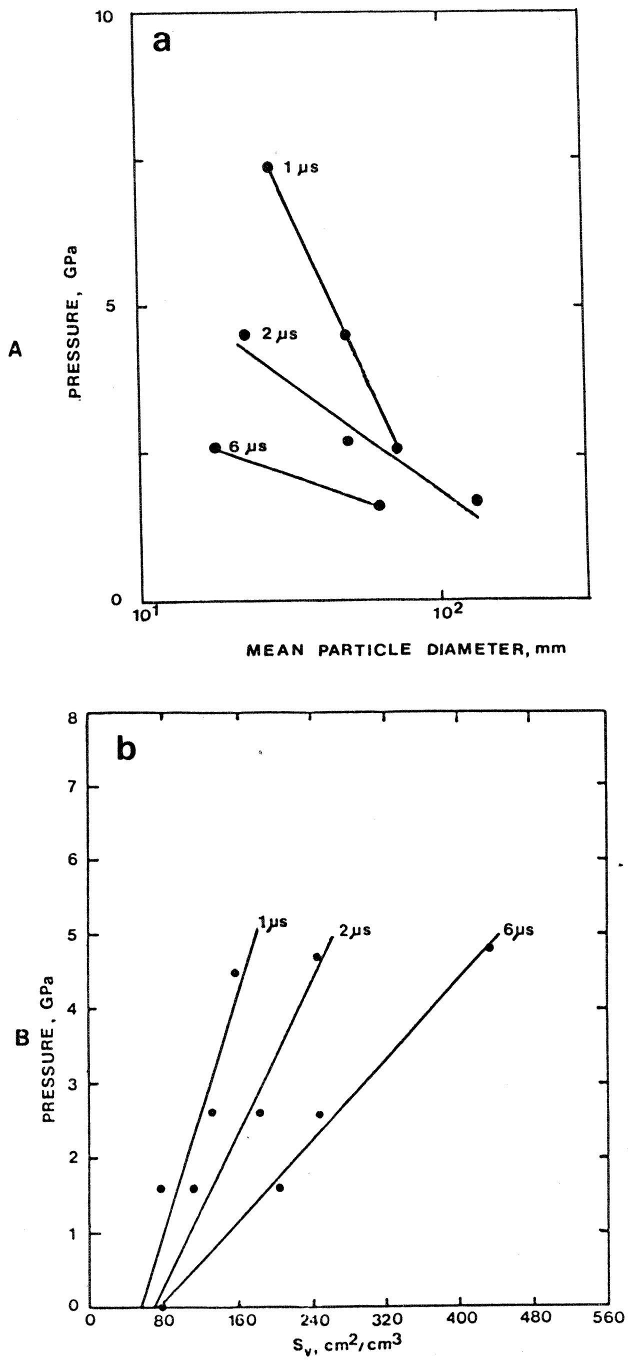

Fig. 6. (a) Mean particle diameter versus pressure for three nominal pulse durations in shock-loaded quartz monzonite (From Aimone, Meyers, and Mojtabai, 1984). (b) Crack density (surface area per unit volume) versus pressure for three nominal pulse durations in shock-loaded quartz monzonite. From (Mojtabai, 1984).

creasing pulse duration, at constant pressure, and with increasing pressure, at constant pulse duration. This is shown in Fig. 6(a) for quartz monzonite copper porphyry subjected to different shock-wave pulses.

(b) The internal damage generated in rock by passage of a shock wave is much larger than the damage simply observed by the fragments. The total surface area of cracks generated internally (within the fragments) is higher, by at least one order of magnitude, than the surface area of the fragments. Thus, a great portion of the damage remains hidden inside the fragments. Fig. 6(b) shows the effect of pulse duration on the formation of cracks. For the specific case shown in Fig. 6(b) the damage S_v (surface area of cracks per unit volume) can be described as:

$$S_v = 9.1\, t_p P + 5.3\, P^2 - 11.2\, P - 3.8\, t_p^2 + 34.9\, t_p + 38$$

where t_p is the pulse duration and P is the pressure.

This time-dependent failure of rocks is consistent with the concepts of nucleation, growth, and coalescence. If a realistic description of the passage of a shock wave through a rock is attempted, it has to incorporate these effects. It is important to assess the energy stored in a material, as defects, due to the passage of a stress wave. The energy dissipated by the wave can be determined from instrumented experiments, as shown in Fig. 7. A stress pulse, with an initial pressure of 2.5 GPa and pulse duration of 0.5 μs, loses momentum as it travels through the material. This can be clearly seen by comparing the pulses passing through gages 1, 2, and 3. This energy loss is equal to the energy stored in the material. The various energy-storage mechanisms are:

(a) Generation of cracks
(b) Pore collapse
(c) Phase transformations and interfacial defects
(e) Frictional processes such as sliding of crack surfaces and dislocations motion.

2.3. Spalling (Meyers and Aimone, 1983)

Spalling is a material failure produced by the action of tensile stresses developed in the interior

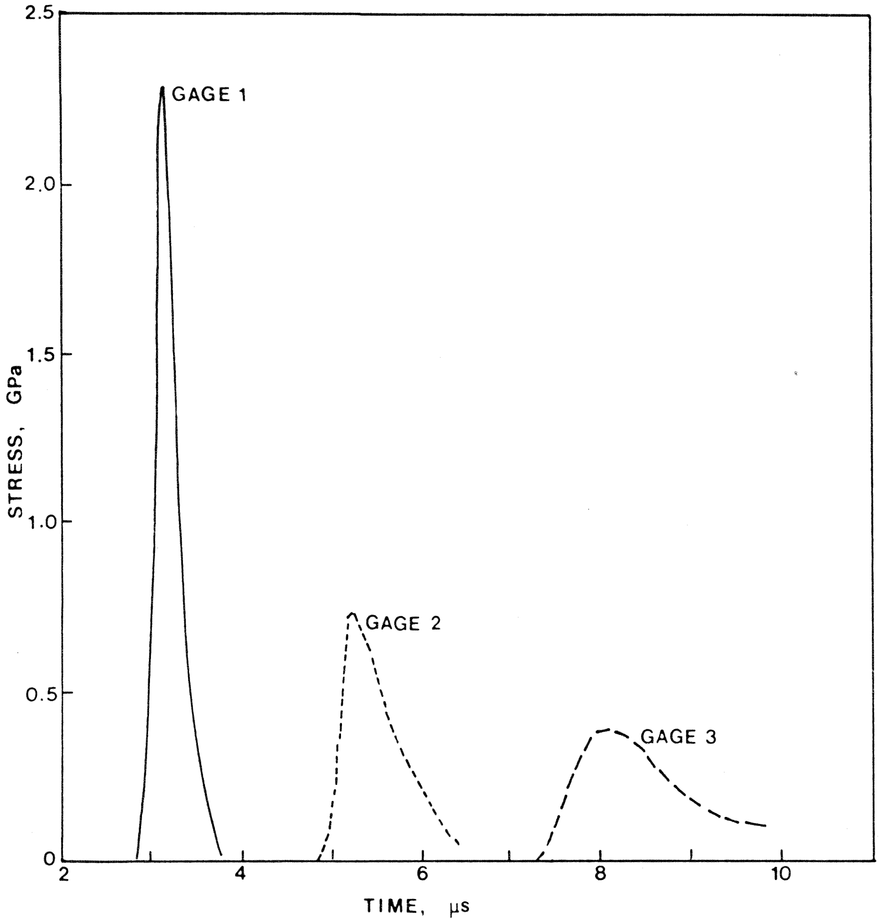

Fig. 7. Attenuation of shock wave in quartz monzonite; from (Lee, 1984).

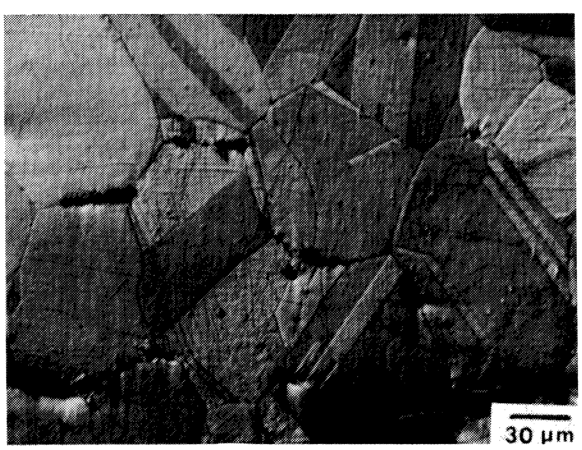

Fig. 8. Voids forming in copper due to tensile pulse produced by reflection of shick waves at a free surface; from (Christy, 1984).

of a body when a shock wave interacts with the reflected wave (tensile) produced at a surface. Spalling is an important component of dynamic failure. On the micromechanical level, spalling is known to result from the nucleation, growth, and coalescence of voids (or cracks). Figure 8 shows a typical incipient spall formation in copper; total separation of the material will occur when the voids have coalesced. Fig. 9 shows the rationale developed at SRI-International (Curran, Seaman, and Shockey, 1974) for the understanding of spalling. As the tensile stress rises, voids nucleate. With void growth, the specific volume increases. After some growth, coalescence takes place. The final step in the process is fragmentation. Curran, Seaman, and Shockey (1974) have developed equa-

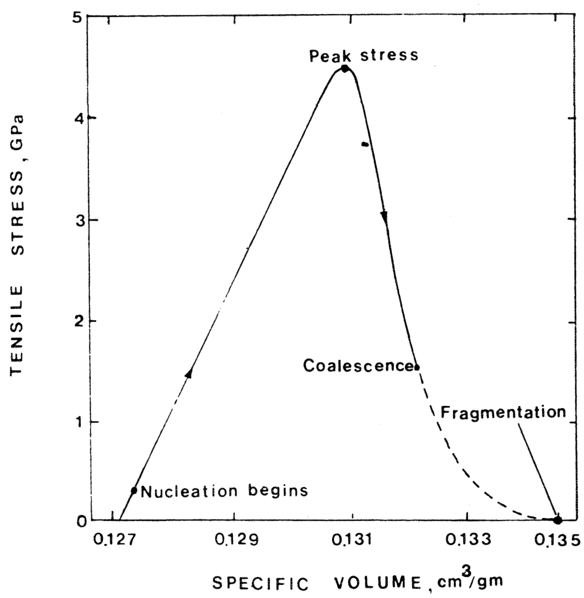

Fig. 9. Stress–volume path for Armco iron loaded to fragmentation at constant strain rate. The dashed line marks the region in which fragmentation progresses by microcrack coalescence. From (Curran, Seaman, and Shockey, 1977).

tions that describe the processes of nucleation ($\dot{N}$) and growth ($\dot{R}$) rates

$$\dot{N} = \dot{N}_0 \exp[(\sigma - \sigma_{n0})/\sigma_1],$$

$$\dot{R} = \left(\frac{\sigma - \sigma_{g0}}{4\eta}\right) R$$

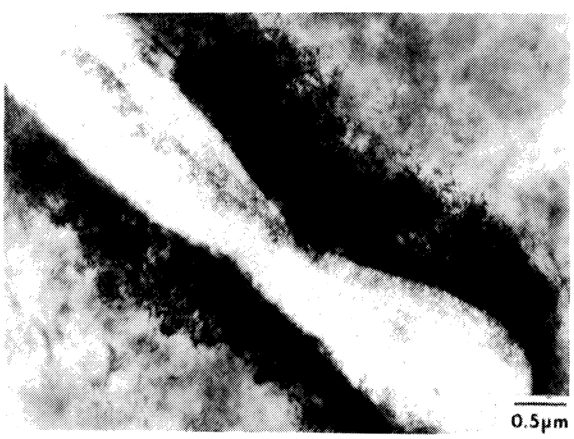

Fig. 10. Peanut-shaped void viewed by high-voltage transmission electron microscopy; from (Meyers and Pak, 1984).

where $\dot{N}_0$ is the threshold nucleation rate, σ_{n0} is the tensile threshold stress, σ_1 is the stress sensitivity for nucleation, σ_{g0} is the threshold stress for growth, and η is the viscosity of the medium. These equations are important in the formulation of computer codes describing fragmentation, such as SRI's NAG-FRAG. What is lacking is a micromechanical (in terms of dislocation movement) model that explains the growth rate, $\dot{R}$; one has to first establish the plastic deformation field around a growing void. Figure 10 shows the first observation of a void by transmission electron microscopy (Christy, 1984). This required the high-voltage transmission microscope of the National Center for Electron Microscopy. One can clearly see a dark region around the peanut-shaped void. This is the result of plastic deformation around the void. These observations will eventually lead to the formulation of a growth-rate model.

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