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Response of materials under dynamic loading

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Abstract

The dynamic response of materials is important to the performance and failure resistance of many engineered systems. Experimental methods are now available to measure this response over a wide range of strain rates and temperatures. Application of these methods to the investigation of the dynamic response of metals, glasses and ceramics has led to important advances in understanding, while bringing into focus important issues that remain to be resolved. In this paper, a number of these issues are discussed and potentially fruitful research directions are suggested.

1. Introduction

High rate loading occurs in a wide range of technologically important applications including such obvious examples as armor penetration, crash worthiness of vehicles, high speed machining, rock blasting, and high speed forming. It occurs as well in all dynamic failure of structural components resulting from such wide ranging conditions as impact of a gravel stone on a windshield, impact of dust particles with aerospace vehicles, and ingestion of geese by jet engines. In addition, high rate loading occurs repeatedly in many aspects of everyday life including the impact of golf balls, the suppression of vibrations, and the operation of high speed, electro-mechanical actuators.

When the loading rate is high, the mechanical response of a material is generally different than it is at a lower loading rate. Such rate dependence is observed for nearly all materials including metals, ceramics, glasses and polymers. Over the past fifty years, improved understanding of the dynamic response of materials has been developed through major advances in the capability for high rate experiments. Two of the principal advances in experimental capability are the development of the so-called split Hopkinson bar or Kolsky bar and the development of the plate impact experiment. Each approach provides rich and varied configurations that have been adapted to the investigation of a wide

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range of dynamical behavior of materials. Initially the Kolsky bar was used to study the dynamic plastic response of metals under uniaxial compression at strain rates from $10^2\,s^{-1}$ to $10^3\,s^{-1}$. Subsequently it has been extended to study plastic flow in torsion and in tension as well as dynamic fracture. More recently it has been adapted to obtain higher strain rates ($>10^4\,s^{-1}$), operate at elevated temperatures, and be applicable to a wider class of materials including such brittle materials as ceramics and such soft materials as polymers. The plate impact experiment was developed originally to provide data on the equation-of-state or pressure-volume-temperature relationship for a material. Later this experiment was adapted to provide information on the shearing resistance of materials, at small plastic strains, as well as their tensile strength. More recently this experiment has been adapted to provide combined pressure and shear loading in order to study the dynamic shearing resistance of materials at large shearing deformations at the very high strain rates of $10^5\,s^{-1}$ to nearly $10^7\,s^{-1}$. Pressure-shear plate impact experiments have been extended to elevated temperatures and to the study of dynamic friction.

With this experimental capability it is now possible to study the dynamic inelastic response of materials over the range of strain rates from, say, $10^2\,s^{-1}$ to $5 \times 10^6\,s^{-1}$ — a range that covers the strain rate regimes of importance in most dynamic applications. At this time, research is directed not so much toward extending the range of strain rates, but toward the development of mechanism-based models to describe the observed response. One cannot simply test all materials over a wide range of strain rates and expect to develop an understanding that allows the response over loading regimes of specific applications to be predicted. Instead, mathematical models must be developed and calibrated with experimental data to provide a useful predictive tool.

2. Metals

For metals one of the principal directions in which research is needed is in the measurement, understanding and modeling of the dynamic response at elevated temperatures. Elevated temperatures arise naturally in high rate deformation through the heating associated with plastic working and the lack of sufficient time for the heat to be conducted away. In other critical applications, such as high speed machining, the elevated temperatures arise because of global heating of the region of interest as well as the localized heating associated with shear bands. Dynamic fracture also leads to substantial heating in the region near the tip of the crack.

Understanding of the dynamic plastic response of metals at high temperatures is of fundamental importance for the understanding of the rate-controlling mechanisms of plastic flow as well as of technological importance for applications in which high temperatures occur. The fundamental question regarding the rate controlling mechanisms is over the competition between the intrinsic resistance of the lattice to the motion of dislocations and the extrinsic resistance that such defects as vacancies, interstitials, dislocations, and inclusions provide to the motion of dislocations. The outcome of this competition is strongly affected by temperature because the intrinsic resistance of the lattice, due to the viscosity associated with thermal phonons, tends to increase with increasing temperature whereas the extrinsic resistance associated with defects tends to decrease with increasing temperature. For defects with long-range stress fields the effects of temperature on the extrinsic resistance can become quite weak. The outcome of the competition is also strongly affected by the crystal structure: bcc metals tend to have greater intrinsic resistance, yet a reduction in flow strength with increasing temperature, due to large short-range energy barriers for dislocations to overcome; fcc metals tend to show a strong increase in flow stress with increasing strain rate at elevated temperatures.

Thermal softening, or the reduction of flow stress with increasing temperature, is a destabilizing feature of the plastic flow of metals that results in the localization of shear strain into shear bands at high strain rates. For this reason, the dependence of the flow stress on temperature at high strain rates

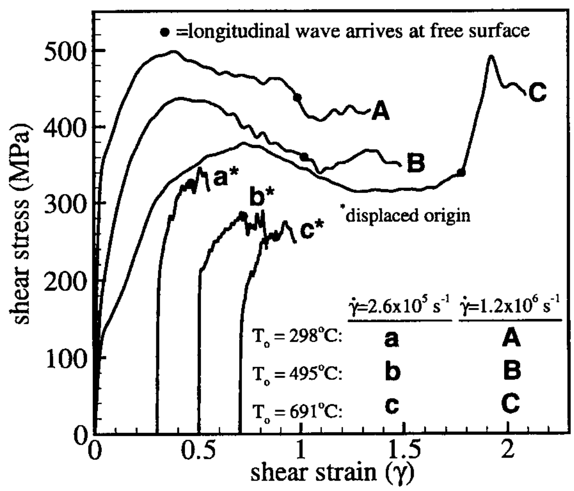

Fig. 1. Dependence of the shearing resistance of OFHC copper on strain rate and temperature at very high strain rates. Data are from pressure-shear plate impact experiments in which copper foils are sandwiched between two hard plates. Temperatures shown are the temperatures prior to loading; the heat of plastic working is calculated to increase the temperature by as much as 150°C at plastic shear strains of approximately unity. Segments of the curves beyond the time shown for the arrival of a reflected longitudinal wave could include spurious effects due to the plates not having perfectly parallel surfaces. (From Frutschy and Clifton, 1998b.)

and elevated temperatures is extremely important in understanding thermoplastic instabilities of plastic flow in such applications as high speed machining, perforation of well casings, and armor penetration. Largely because of the difficulty of acquiring data on the response of metals in the regime of high strain rates and high temperatures, constitutive models used to describe the shearing resistance of metals in this regime have been highly simplified empirical models. Commonly used models (e.g. those used in the comparisons by Frutschy and Clifton, 1998a) involve multiplying the expression for the flow stress at room temperature by a temperature-dependent factor that reduces to zero as the temperature approaches the melting temperature of the metal. Recent experiments at elevated temperatures and high strain rates have shown the inadequacy of such models. In particular, Kanel et al. (1996) have shown, by means of plate-impact experiments at elevated temperatures, that the dynamic yield stress for aluminum and for magnesium remains high at temperatures approaching the zero-pressure melting point and that the yield stress actually increases with increasing temperature in this high temperature regime. In related work, Frutschy and Clifton (1998a, 1998b) have shown, by means of pressure-shear plate-impact experiments on thin samples of high-purity copper sandwiched between hard plates, that the flow stress at large strains, and for strain rates of the order of $10^6 s^{-1}$ (See Fig. 1), remains high at temperatures up to 85% of the zero-pressure melting point. Such behavior is not in agreement with predictions based on current constitutive models. Further investigation of the plastic flow of metals at high strain rates and high temperatures should address both the development of a broader base of experimental data and the development of mechanism-based models that describe the behavior in this regime. Above melt, at large hydrostatic pressure and high shear rates, fcc metals are reported to flow as Newtonian liquids (Mineev and Mineev, 1997). For applications involving very high strain rates, such as armor penetration and explosive welding, material models are needed that describe the dependence of the flow stress on temperature, strain rate, and pressure through the transition from the solid to the liquid state. For fcc metals with low melting points, experiments to study this transition appear to be possible with current capability.

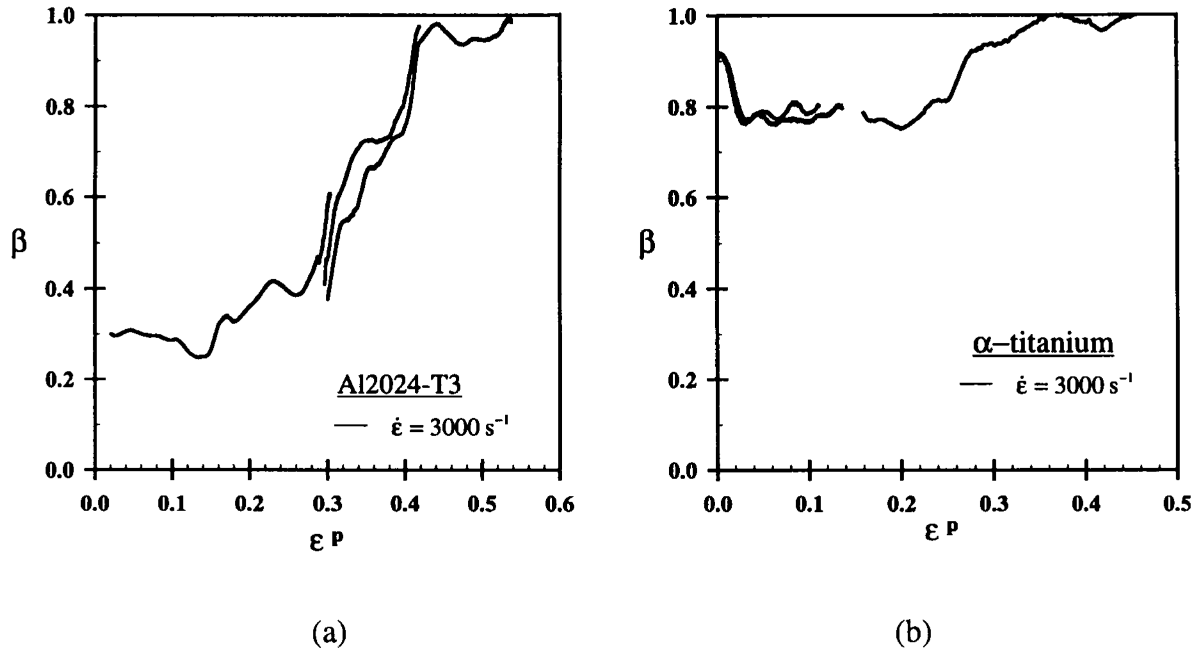

Fig. 2. Fraction (β) of plastic work converted to heat: (a) Al2024-T3; (b) α-titanium. Data are obtained from Kolsky pressure bar experiments in which the deformation-induced temperature rise was measured using a high-speed HgCdTe photoconductive detector. (From Hodowany et al., 1998.)

Experimental effort on the dynamic plastic response of metals should include investigation of the fraction of plastic work converted to heat — the so-called β factor — because this factor plays an important role in the prediction of thermoplastic instabilities. Until recently it was widely accepted, based on quasi-static experiments conducted more than 60 years ago, that $\beta \cong 0.9$ was a reasonable approximation. However, recent experiments have shown that β generally increases with accumulated plastic strain, reaching values of 0.9 and higher only at relatively large strains as shown in Fig. 2. Using too large a value of β (e.g. $\beta = 0.9$) for moderate strains can lead to predictions of too great an increase in temperature. Then, because of thermal softening, the simulations tend to underestimate the shearing resistance and predict premature shearing instability. If the thermal softening at high temperatures and high strain rates is also over-estimated, as previously mentioned, then the under-prediction of shearing resistance at high strain rates and high temperatures would be compounded. Therefore, research is needed that clarifies both the heat dissipated during plastic deformation (Rosakis et al., 1998) and the effect of the corresponding temperature increase on the shearing resistance of the material.

Computer simulation of dislocation motion is making possible numerical experiments that cannot be done as physical experiments. These numerical experiments hold great promise for systematic examination of the kinetics of various dislocation mechanisms. Inclusion of the effects of temperature appears to be particularly important. For individual dislocations there is need for a systematic study of the temperature dependence of dislocation motion through an otherwise-perfect lattice. Gumbsch and Gao (1999) have shown the possibility of such studies through their simulation of the motion of an edge dislocation in a pre-strained thin strip at low temperature. Studies like theirs should be extended to include the effects of the interaction of dislocations with various point defects: vacancies, interstitials, and substitutional atoms; additional studies should examine the kinetics of the interaction of dislocations with other dislocations especially at various dislocation junctions.

The dynamic fracture of metals occurs through a combination of ductile rupture and cleavage. For a given metal the competition between these two mechanisms depends on the temperature and the rate of loading. As the temperature is decreased, or the rate of loading is increased, there is less opportunity for relaxation of the stresses at the crack tip. As a result, the stresses are larger near the crack tip and failure by cleavage becomes more prevalent. Such a ductile-brittle transition is extremely important for applications because the transition is accompanied by a marked decrease in fracture resistance.

While important contributions have been made recently to the understanding of the ductile-brittle transition the transition is not fully understood even for quasi-static loading conditions; for dynamic loading conditions the understanding is relatively primitive. This situation needs to be addressed as the possibility of dynamic loading causing a change to a more brittle failure mode is potentially hazardous.

One regime of dynamic plastic response that is particularly difficult to address experimentally, but that is important in dynamic fracture, is that of high plastic strain rate at small plastic strains. This regime appears to be important in high-speed crack propagation where the crack is propagating fast enough that relatively little plastic deformation occurs in front of the crack before the crack arrives. In this case the stress may be high because of the crack-tip singularity, but the plastic strain rate may be small because the density of mobile dislocations in front of the tip may be small. Experimentally, the conditions of small plastic strain and high plastic strain rate are normally not achievable in Kolsky-bar type experiments because, during the time required to produce nominally homogeneous states of stress through the thickness of the sample, the accumulated plastic strain becomes substantial. A more useful experiment for this regime is the so-called precursor decay experiment of plate impact studies. In these experiments the decay of the elastic precursor (i.e. the leading wave, propagating at the elastic longitudinal wave speed) is related to the plastic strain rate at the wavefront where the plastic strain is nominally zero. The general impression from experiments is that the precursor decay under these conditions is greater than predicted by extrapolating usual plastic flow relations to very small strains. Therefore, it appears that current models tend to underestimate the flow stress at small strains and high strain rates.

Early time records for pressure-shear impact experiments provide another opportunity to examine plastic response in this regime. Numerical simulation of the experiments is required, but the wave profile for the first arrival of the shear wave at the rear surface of the sample is a sensitive measure of the shearing resistance at high shear rates and small shear strains.

Investigations of the high strain rate response of metals have tended to be directed toward pure metals and common structural alloys because of their evident importance in familiar applications. However, research is clearly needed on newly developed metals and intermetallics for which little or no data are available on their response at high strain rates. For example, only pilot studies have been conducted on zirconium-based bulk amorphous metals and results suggest the likelihood of pronounced loss of shearing resistance during impact due to local heating that caused the temperature of the region of contact to reach or approach the glass transition temperature (Cline et al., 1997). As another example, only pilot studies have been conducted on shape memory alloys which undergo martensitic phase transformations when subjected to impact loading at even modest velocities. Such studies on Cu–Al–Ni single crystals by Escobar and Clifton (1995) have determined the resolved shear stress required for the transformation and an indication of the velocity of propagation of the phase boundary. To be able to describe the kinetics of the transformation for the design of actuators, or for improved material processing in which phase transformations occur, it is necessary to develop physical understanding and suitable mathematical models for relating a driving force to the velocity of the propagating phase boundary. Stress wave loading of polycrystalline shape memory alloys (e.g. NiTi) poses additional theoretical and computational challenges that need to be overcome before the kinetics of devices made from these materials can be simulated.

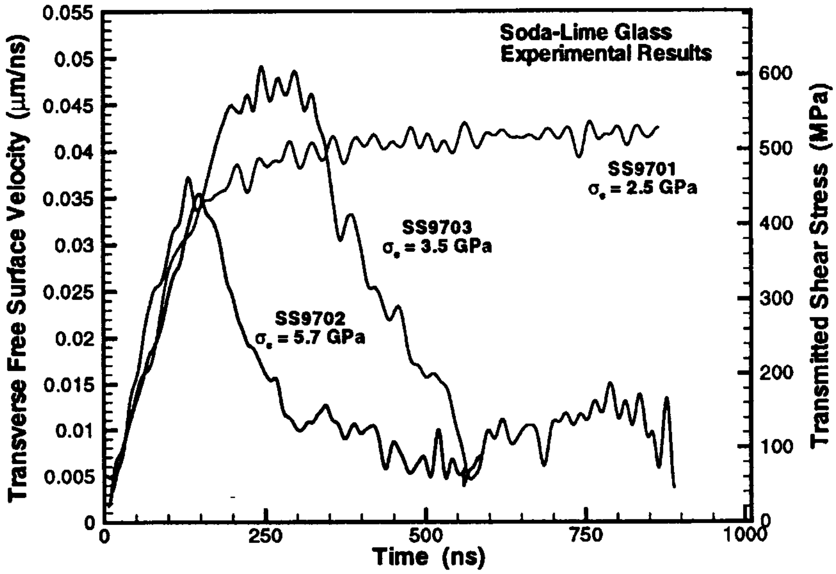

Fig. 3. Transverse velocity profiles at the rear surface of a target assembly in which a 5 μm thick soda-lime glass was sandwiched between two hard plates and subjected to pressure-shear impact with a skew angle of 22°.

3. Glasses

The dynamic response of glasses at high strain rates is important both for the impact response of glasses and that of ceramics in which a glass has been added as a sintering aid. In the latter case, the glass forms a thin layer between ceramic grains — a layer that may begin to deform plastically while the neighboring grains remain elastic, and that softens further due to the deformation-induced temperature increase. Interest in the shearing resistance of glasses is often regarded to be of secondary importance since in most applications the mechanical failure of glasses is by tensile fracture. However, in impact scenarios the stress state immediately ahead of an advancing projectile has large hydrostatic pressures before unloading waves arrive from free boundaries; consequently, tensile fracture is suppressed and the resistance of the glass to flow under large shear stresses plays a major role in the penetration resistance of the glass or sintered ceramic.

The shearing resistance of a common soda-lime glass at strain rates of $10^6\ s^{-1}$ and pressures of a few GPa is shown in Fig. 3 (Sundaram and Clifton, 1997). At the lowest impact velocity the shearing resistance is high (≈ 500 MPa) and remains high while substantial shearing deformation occurs. As the impact velocity is increased, with corresponding increase in pressure and shearing rate, the shearing resistance is shown to change character. After an initial rise the shear stress falls sharply before stabilizing at a much lower value (≈ 100 MPa). Sundaram and Clifton (1997) noted that the shear stress begins to fall at a shear strain of approximately 2.0 in both of the latter cases and reaches a value of only 0.5 in Shot SS9701 which exhibited no fall in the shear stress. They suggested that the sharp loss in shearing resistance may be due to a bond-switching mechanism suggested by Myuller (1960) which involves a switching of adjacent covalent Si–O bonds at large shear strains. Whether or not this mechanism is the correct one needs to be examined further by additional experiments and by molecular dynamics simulations.

The loss in shearing resistance shown in Fig. 3 has been suggested by Clifton et al. (1997) as being responsible for so-called *failure wave* phenomena that have been observed in plate impact experiments conducted on several glasses by a number of investigators (Brar et al., 1991; Kanel et al., 1992; Raiser et al., 1994; Bourne et al., 1998). These phenomena include the propagation of a wave across which the longitudinal stress does not change measurably, but the transverse normal stress increases substantially — thereby reducing the shear stress. The *failure wave*, which occurs at impact velocities above a threshold value, gets its name from the observation that the measured spall strength of the material behind the failure wave is effectively zero. The speed of a *failure wave,* being less than an elastic shear wave speed, does not correspond to any elastic wave speed for the glass. The experiments by Clifton et al. (1997) were pressure-shear plate impact experiments that included cases in which a glass plate impacted a hard, steel plate so that the stress state behind the failure wave could be inferred from the measured free surface velocity of the steel plate. From these experiments it is evident that the glass in the 'failed' region has undergone a strong loss in shearing resistance; however, the longitudinal normal stress in this region is essentially the same as expected for elastic response during both loading and unloading, except for the final stages of unloading. These results suggest that *failure waves* are related to a loss in shearing strength and that the loss in spall strength may be due to cracking that occurs as the compressive longitudinal normal stress approaches zero when the unloading wave from the rear surface arrives. Bourne et al. (1998) found that, for a target assembly consisting of two soda-lime glass plates, a second failure wave is generated at the interface between the two plates when the leading longitudinal wave arrives. This observation suggests that surface effects play a role in the nucleation of failure waves.

Further research is required to examine the validity and generality of these various observations as well as to develop a satisfactory, physically-based constitutive model for simulating the principal features of the measured wave profiles.

Such research can be expected to add not only to the understanding of the dynamic response of glasses, but also to the understanding of the dynamic response of bulk amorphous metals and metallic glasses. Additionally, shock loading at high impact velocities has been shown to lead to crystalline to amorphous phase transitions in many materials (e.g. Sikka and Gupta, 1997). Thus, improved understanding of the dynamic response of amorphous materials would have far-reaching implications for high velocity impact applications ranging from the design of armor, to the design of space vehicles against the impact of small particles, to the understanding of the impact of meteors on the earth and other planets.

4. Ceramics

Because of their high hardness and light weight, ceramics are attractive materials for both military and civilian armor. Their hardness coupled with their high strength at elevated temperatures make them attractive for such applications as machine tools, abrasives, and engine components. Whether for armor, or for other applications in which the primary loading is quasi-static, a critical consideration in the decision to use a ceramic in a given application is its resistance to impact — either as a primary design objective or as a requirement to prevent damage due to handling accidents.

Understanding the dynamic response of ceramics involves elements of the considerations that have been mentioned previously for metals and glasses. When confining stresses are sufficiently large, as can occur initially in front of a long rod penetrator impacting ceramic armor, then the ceramic grains can deform plastically by the motion of dislocations — as for metals. When confining stresses are small, as in most applications where free surfaces are present, then polycrystalline ceramics tend to fail by the nucleation and growth of microcracks along grain boundaries. Even for the strong confinement that occurs under the plane wave loading conditions of plate impact, the dynamic response of polycrystalline ceramics is generally understood to involve failure along grain boundaries (e.g. Grady, 1998). Such failure can be attributed to the heterogeneity of the microstructure, the relatively high shearing resistance of the grains, and the weakness of the interfaces due to the presence of impurities and voids, or a glass layer. At moderate confining stresses, as in Kolsky pressure bar experiments with lateral confinement (Chen and Ravichandran, 1996), the inelastic dynamic response in longitudinal compression appears to be largely due to splitting along the loading axis. An elementary theory for this failure mechanism has been provided by Bhattacharya et al. (1998). Without confining stresses, failure of ceramics under dynamic loading usually occurs by tensile fracture along grain boundaries as reflected waves result in tensile loading and the interfaces are generally weak in tension.

To advance the usefulness of ceramics in applications where the possibility of dynamic loading must be considered it is necessary to develop a theoretical framework that accounts for deformation and failure under a variety of microstructurally-based mechanisms. This framework should account for rate dependence (e.g. Espinosa, 1995) as the propagation of microcracks, the shearing of glass layers, and the motion of dislocations are all rate dependent on the time scales of importance in impact. A means for including heterogeneity is required to allow the possibility of failure by splitting under fully compressive stress states. Similarly the role of voids in the failure of heterogeneous brittle solids needs to be addressed for general multi-axial states of stress.

Dynamic failure by the growth and coalescence of grain-boundary microcracks involves the cooperative interaction of propagating cracks. Insight into such processes is required from the perspective of stochastic mechanics and from computer simulations of the debonding of assemblages of grains using realistic cohesive models for the grain boundaries. Critical elements in the development of a physically-based model of the dynamic deformation and failure of ceramics should be addressed by experiments designed specifically to examine the validity of particular assumptions or to measure particular material properties. Model systems may be suitable for some of these experiments in order to facilitate the direct comparison of theory and experiment.

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