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TITLE SHOCK EXPERIMENTS IN METALS AND CERAMICS

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SHOCK EXPERIMENTS IN METALS AND CERAMICS

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Shock recovery and spallation experiments, in which material structure / property effects are systematically varied and characterized quantitatively, offer two important experimental techniques to probe the physical mechanisms controlling shock processes and dynamic fracture. This paper highlights the current state of knowledge and principal challenges of the structure / property effects of shock-wave deformation on metals and ceramics. Recent shock-recovery and spallation experimental results on post-mortem material properties and fracture behavior in metals and ceramics are reviewed. Finally, the influence of shock-wave deformation on several intermetallics and a recent experiment examining the Bauschinger effect in Al-4% Cu during shock loading are presented.

1. INTRODUCTION

Since the late 1950's shock recovery and spallation experiments have been used to probe the structure / property effects and physical mechanisms controlling the deformation processes and dynamic fracture of materials subjected to impulse or shock loading. Several review papers have summarized in-depth the structure / property response of numerous metals, alloys, ceramics, and non-metallic materials to shock-wave deformation [1-9] and spallation [10]. Overall, the extreme conditions of strain rate and temperature imposed by a shock are known to induce a high density of defects in crystalline materials which in turn affect the post-shock mechanical properties. Accommodation of the applied high-rate-uniaxial-strain pulse during the shock is observed to cause the generation of dislocations, deformation twins, point defects, stacking faults, or in some materials may activate a pressure-induced phase transition in response to the shock. As a result of this high defect density, most metals and alloys are observed to exhibit a greater degree of hardening, measured via post-shock hardness or stress-strain response, due to shock loading than quasi-static deformation to the same total strain, particularly if the metal passes through a polymorphic phase transition, such as the $\alpha-\varepsilon$ in iron or the $\alpha-\omega$ in titanium [1-6,8].

The exact type and arrangement of these defects, whether dislocation cells, planar faults, twins, or phase products in the post-shock substructure is dependent on both "chemically" and microstructurally controlled components within the material as well as the externally imposed variables such as the temperature and shock loading parameters. Chemical components include the influence of material chemistry (due to the alloying and interstitial content) on the starting crystal structure, stacking fault energy, phase constitution (influencing the precipitation of second phases), and phase stability of the material undergoing shock deformation or dynamic fracture. Independent from the chemistry of the starting material variables such as the grain size, second phase size and distribution, and crystallographic texture can be altered through the manipulation of heat-treatment. More recent developments in the growing area of composite materials can also introduce chemically inert dispersoids, such as Al_2O_3 or SiC in aluminum alloys for example, which can significantly influence the structure property response of the material.

Since the last Explomet conference held in 1985[11], numerous studies have investigated the influence of a variety of microstructural and externally applied variables on the "real-time" and post-shock structure / property responses of a broad class of materials. Many of these studies have been reported in recent conference proceedings including: the two American Physical Society conferences on condensed matter held in 1987 [12] and 1989 [13], IMPACT '87 [14], the Oxford conference on materials at high rate [15], DYMAT 88 [16], and a conference on dynamic compaction technology[17]. In addition, two recent books edited by Blazynski [18] and Murr [19] summarize some additional shock studies. Scientists interested in examining the breath of the rapidly evolving shock-wave field, in particular as related to materials, are advised to examine these references.

The intent of this review is to highlight some of the recent experimental results and research trends of shock effects on materials with particular emphasis on the shock deformation behavior of conventional metals, alloys, and ceramic materials. Due to space limitations only studies on prior-produced materials will be reviewed. The areas of shock-wave synthesis and dynamic compaction of materials will not be included. Even upon a cursory review of the literature over the past five years, the stronger than usual influence of funding priorities from the government and military sector, both in the USA and abroad, on shock-wave and hyper-velocity research are readily apparent. The largest focusing elements were the formation in 1983 of the Strategic Defense Initiative and more important for materials research in particular was the report on armor/antiarmor by the Defense Science Board in 1985 [20].

The general area of armor/antiarmor, funded heavily by several DARPA and DoD initiatives to universities, federal labs, and the private sector, started a broad range of studies on materials research in the areas of penetration failure mechanics, constitutive behavior, and spallation particularly in the area of ceramics. This research emphasis has in turn created a flurry of activity in the materials/physics communities in the areas of shear band formation, microstructural effects on spallation, shock-wave damage mechanisms in ceramics, and constitutive data on armor/antiarmor materials, to name just a few. The review of recent results will be divided into sections covering metallic and ceramic materials. Finally, some recent experimental results will be presented on two emerging materials shock-wave research areas; the effects of laser-induced shocks on materials, the influence of shock-wave deformation on several intermetallic materials, and then recent experimental results demonstrating how the strain-path change inherent during a shock can lead to a Bauschinger effect in some materials.

II. METAL AND ALLOYS STUDIES

Numerous aspects of the deformation and spall behavior of metals and alloys subjected to shock loading have been examined experimentally and correlated with materials models in the past 5 years[21-43]. Consistent with previous work the increased strain rate in the shock increases the propensity for deformation twinning in shock-loaded metals[38,39] including in Al-4.8 Mg[21] and 6061-T6 Al[22]. Microstructure is still observed to significantly affect the shock hardening and deformation response of materials[21-23,41]; the shock hardening of solution or underaged JBK-75 stainless steel and 6061 Al alloys exceed that of the peak aged materials[23]. Several studies have also examined the kinetics of the martensite transformation induced by controlled tensile pulses produced by shock loading[29,30]. The effect of dynamic prestraining[34] and repeated shock loading[26] on subsequent mechanical properties have shown that prior dynamic loading reduces subsequent strain hardening during shock or quasi-static reloading. High temperature preheating prior to shock loading was found to enhance the shock hardening and impact strength of steel [28]. In addition to the broad range of experimental studies, Mogilevsky [27,31,32] and Follansbee[33] have conducted modeling studies considering the nature of defect generation during shock loading and compared their predictions with experimental results.

Finally, lasers represent a new method for assessing the shock-wave and dynamic fracture of materials. Lasers offer the possibility of producing shock pressures in targets of 1 to more than 10^2 GPa with pulse durations related to the laser pulse length, usually in the nanosecond range. The limitations of the laser shock pulse are principally its diameter owing to the small diameters of typical lasers and total pulse duration. The laser pulses however can be equivalent to the impacts produced by microparticles at high velocity and are therefore of great interest to studies such as micrometeroite impacts in outerspace. Several recent studies have assessed the influence of laser shocks on the deformation substructure of metals[44,45]. These studies have shown that the substructure evolution in iron and stainless steel targets are very similar to those obtained by conventional shock methods.

III. CERAMIC STUDIES

The study of shock-wave and dynamic fracture effects on brittle solids, in particular ceramics, represents one of the largest growth areas for materials science in the past five years[46-62]. Much of this interest has been catalyzed by a need for data concerning the dynamic deformation and fracture behavior of monolithic ceramics and cermets for potential armor applications. Investigations of the Hugoniot Elastic Limit (HEL) of alumina found that the HEL in Al_2O_3 is not strain rate dependent[46] and that it is higher than the static yield strength [47]. The question of whether plastic flow and microcracking are activated above or below the HEL of a ceramic has been the subject of numerous studies yet the results do not define a self consistent pattern.

Studies have shown that while alumina retains its shear strength to 15 GPa[49] in some cases alumina is seen to start cracking below the HEL[50,58,60,62] while in another case no microcracking was observed, but only pore closure, up to twice the HEL [51,56,59,61]. In addition to the post-shock sample analysis for cracks, in one study the observation of a decaying stress pulse was attributed to the attenuation of the incident compressive pulse due to compressive damage[60]. Figure 1 conversely shows a spall trace of symmetrically shocked TiB_2 below the HEL displaying a stable shock pulse. Analysis of shock recovered polyphase alumina showed that one reason for the cracking below the HEL was due to the impedance mismatch and residual stresses associated with the glasssy phase[53,57]. Studies of recovered alumina[52] and an Al-B_4C cermet [54] both observed dislocation activation above the HEL and in the cermet none below the HEL. Several studies have also measured the spall strengths of TiB_2[55] and Al_2O_3[56]; the spall strength of TiB_2 decreasing to zero above the HEL. Given the difficulties in recovering ceramic samples post-shock[9], resolution of the pressure and material dependencies on the deformation and microcracking in shock-loaded ceramics will be difficult to conclusively settle.

IV. RECENT SHOCK STUDIES

A. *INTERMETALLICS*

Intermetallics and some composites are receiving increasing attention due to their high specific strengths, stiffnesses, and potential high temperature properties. Within the last few years high-strain-rate and shock-loading experiments have begun to probe the dynamic deformation response of a range of intermetallic and composite materials. To assess the substructure evolution and mechanical response of Ni_3Al as a function of peak pressure, specimens were "soft" shock recovered and mechanical testing and TEM samples were sectioned from the recovered disks. Figure 2 shows the reload compressive stress-strain response of the Ni_3Al following shock loading as compared to the annealed starting material. The shock-loaded stress strain curves are plotted offset starting at the approximate total transient shock strains [calculated as $4/3 \ln(V/V_0)$ where V and V_0 are the final and initial volumes] for the two shocks. The peak shock pressures are approximated using the EOS of pure nickel in the absence of an EOS for Ni_3Al. The reload yield strength of Ni_3Al increases from 250 MPa to 750 and 1250 MPa following the ~ 14 and 23.5 GPa shocks, respectively. As in the case of most metals and alloys the effective hardening in the shock-loaded Ni_3Al exceeds that quasi-statically obtained when deformed to roughly the equivalent strain[4].

The substructure evolution in the shock-loaded Ni_3Al is observed to depend on the peak shock pressure consistent with the reload data. Increasing the peak pressure is observed to significantly increase the density of stacking faults and deformation twins. Figure 3 shows the substructure of the Ni_3Al shock-loaded at 996 m/sec (~ 23.5 GPa) consisting of a high density of dislocations on octahedral planes, planar stacking faults, and deformation twins. The substructure evolution in shock-loaded TiAl and Ti_3Al have also been investigated. Similar to Ni_3Al the substructure of both Ti-aluminides is dependent on the peak shock pressure. While the TiAl readily deforms via deformation twinning when subjected to shock loading, the Ti_3Al exhibits a substructure consisting of solely coarse planar slip. Figure 4 shows the high density of deformation twins formed in TiAl shock loaded to 7 GPa. Finally for intermetallics, as mentioned earlier, lasers offer an alternative method for applying shocks to materials. Figure 5 shows the high density of dislocations formed in Ti_3Al due to a laser-driven miniature flyer plate impacted at 4.3 km/sec [63]. The deformation mechanisms activated, while formed by a very short pulse duration due to the ~10μm thick flyer plate, are similar to that observed in Ti_3Al conventionally shock loaded.

B. *BAUSCHINGER EFFECT DURING SHOCK LOADING*

Due to the intrinsic nature of the shock process the structure/property response of a material is a result of the total shock excursion comprised of the compressive loading regime occurring at a very high (~10^5 - 10^8 s^{-1}) strain rate (shock rise), a time of reasonable stable stress (pulse duration), and finally a tensile release of the applied compressive load returning the sample to ambient pressure at a lower strain rate. Collectively the loading sequence in a shock amounts to a single cycle stress/strain path change excursion with elastic and plastic deformation operative in two directions. In this regard the shock process may be compared to a single high-amplitude "fatigue-type" cycle with a dwell time representing the pulse duration [25].

Some materials after reversing the direction of stressing quasi-statically, exhibit an offset yield upon stress reversal due to directional kinematic and isotropic hardening[64,65]. In most instances, given the existence of a back stress acting in the matrix due to the presence of the unrelaxed plastic strain in the particle vicinity, yielding in the reverse direction occurs at a reduced stress level compared with the forward flow stress level. To directly assess if the strain path reversal inherent to the shock contains a bauschinger effect for a shock-loaded two-phase material, two samples of an Al-4wt.% Cu alloy exhibiting a well known bauschinger effect as a function of microstructure were shock loaded to 5 GPa and "soft" recovered in the same shock assembly to assure identical shock-loading conditions.

Figure 6 shows the stress-strain response of the starting microstructures and quasi-statically-reloaded shock-loaded samples. The reload shock-loaded sample curves starting points are shifted to a strain of 7.5% equal to the calculated total shock transient strain. While the as-heat-treated yield strengths of the starting microstructures are the same, the reload behavior of the shock-loaded samples is quite different between the solution-treated and θ' aged microstructures. While the solutionized sample shock hardens above the quasi-static response to equivalent strain levels the aged sample flow behavior is considerably below that of the unshocked aged sample response. This data is consistent with the operation of a bauschinger effect, i.e. reverse strain cycle, occuring during the shock process in the two-phase Al-Cu alloy studied. This effect offers a consistent explanation of the lack of significant shock hardening and/or softening compared to an equivalent quasi-static strain level in 2-phase materials as a whole. Finally, while the bauschinger effect will be manifested in two-phase materials, the proven existence of the strain reversal contribution occuring during the shock process supports the previous model including reversibility as the controlling variable of pulse duration effects[25].

V. SUMMARY

Shock recovery and spallation experiments, in which material structure / property effects are systematically varied and characterized quantitatively, offer two important experimental techniques to probe the physical mechanisms controlling shock processes and dynamic fracture. In conventional metals, alloys, and also in cermets and ceramics experimental studies continue to find that microstructure significantly affects the shock hardening and deformation response. In ceramic materials, the question of when plastic flow and microcracking are activated, in particular above or below the HEL, of a ceramic has been the subject of numerous studies yet the results do not define a self-consistent pattern. A shock loading study on Ni_3Al has shown that similar to most disordered metals and alloys the effective hardening in shock-loaded Ni_3Al exceeds that quasi-statically obtained when deformed to roughly the equivalent strain. Finally, experiments on Al-4wt.% Cu as a function of heat-treatment have demonstrated that the shock process behaves as a bauschinger effect test to materials sensitive to strain-path reversals.

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FIGURE CAPTIONS

FIG. 1 *Stress vs. time spall trace measured in the plexiglas backing for a symmetric* TiB_2 *test with an impact velocity of 596 m/sec (15.2 GPa). Spall strength = 0.63 GPa.*

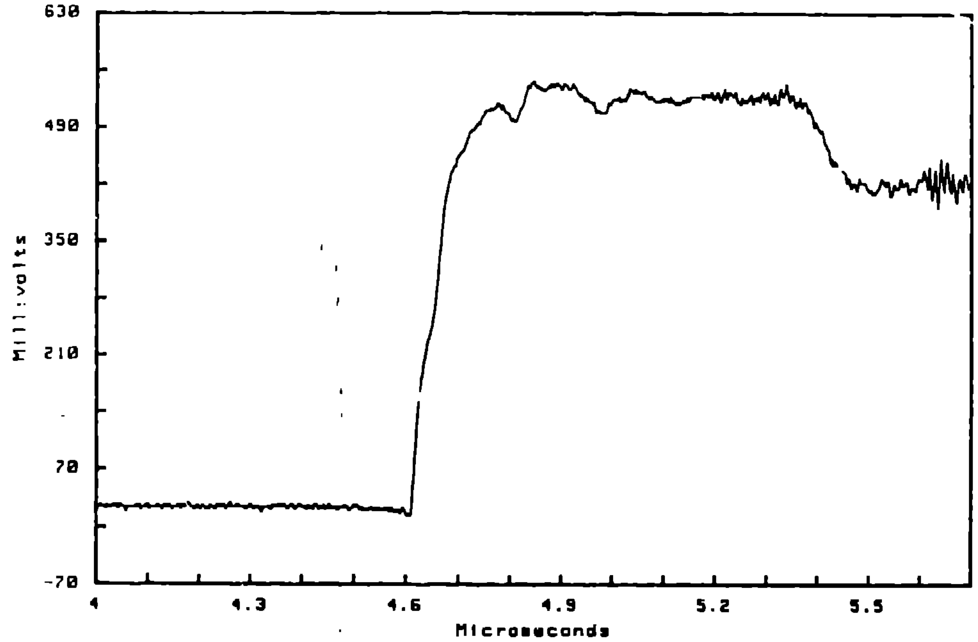
FIG. 2 *Stress-strain behavior of shock-loaded* Ni_3Al *as a function of peak pressure.*

FIG. 3 *Brightfield electron micrograph of dislocations, stacking faults, and deformation twins in* Ni_3Al *shock loaded at 996 m/sec (~23.5 GPa).*

FIG. 4 *Brightfield electron micrograph of deformation twins in TiAl containing* Ti_3Al *shock loaded at 580 m/sec (~7 GPa).*

FIG. 5 *Brightfield electron micrograph of dislocation debris in* Ti_3Al *shock loaded by a laser-driven flyer plate accelerated to 4.3 km/sec.*

FIG. 6 *Stress-strain response of shock-loaded Al-4wt.% Cu as a function of heat-treatment illustrating the Bauschinger effect inherent to the shock process between the solution-treated and* Θ' *conditions.*



630
490
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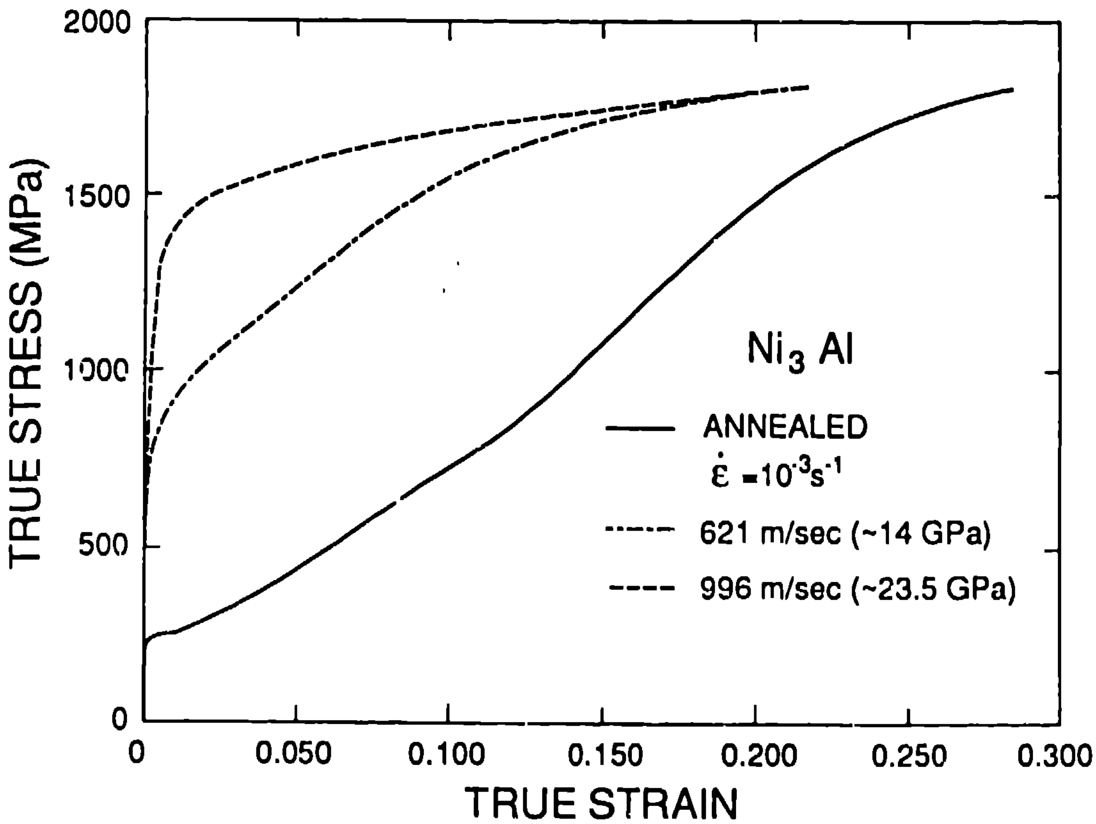
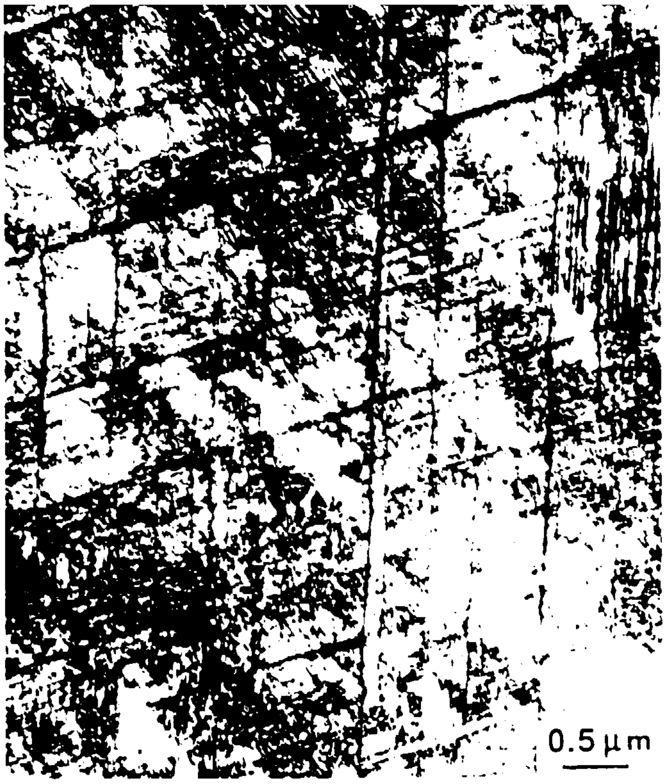
70
-70
Millivolts
4
4.3
4.6
4.9
5.2
5.5
Microseconds

Fig. 1



2000
1500
1000
500
0
TRUE STRESS (MPa)
0
0.050
0.100
0.150
0.200
0.250
0.300
TRUE STRAIN
Ni_3 Al
ANNEALED
$\dot{\varepsilon} = 10^{-3} s^{-1}$
621 m/sec (~14 GPa)
996 m/sec (~23.5 GPa)

Fig. 2


0.5 μm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0.2 μm