Characterization of shocked beryllium

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Abstract. While numerous studies have investigated the low-strain-rate constitutive response of beryllium, the combined influence of high strain rate and temperature on the mechanical behavior and microstructure of beryllium has received limited attention over the last 40 years. In the current work, high strain rate tests were conducted using both explosive drive and a gas gun to accelerate the material. Prior studies have focused on tensile loading behavior, or limited conditions of dynamic strain rate and/or temperature. Two constitutive strength (plasticity) models, the Preston-Tonks-Wallace (PTW) and Mechanical Threshold Stress (MTS) models, were calibrated using common quasi-static and Hopkinson bar data. However, simulations with the two models give noticeably different results when compared with the measured experimental wave profiles. The experimental results indicate that, even if fractured by the initial shock loading, the Be remains sufficiently intact to support a shear stress following partial release and subsequent shock re-loading. Additional “arrested” drive shots were designed and tested to minimize the reflected tensile pulse in the sample. These tests were done to both validate the model and to put large shock induced compressive loads into the beryllium sample.

1 Introduction

Although beryllium has many unique properties, there is surprisingly little recent literature on its dynamic behavior [1–5]. The limited published work on the dynamic and shock loading behavior drives the need for work in this area. The strength and damage behavior of Be under dynamic loading conditions using various drive conditions are presented, including “arrested” shock loading experiments and characterization. This work addresses the shock peak stress for damage and spall to initiate, the shock strength of Be, and the ability for a damaged ensemble of fragments to support a shear stress when re-shocked [4–6]. These experiments also show the behavior of Be under different loading conditions. The triangular shaped wave, produced by explosive drive, interacts with the free surface and produces a region of tension due to the interaction of the free surface rarefaction with the Taylor wave. This is essentially the classic picture of one-dimensional explosively driven spall. Conversely, plate impact spall experiments imparting a square wave pulse have longer time for damage kinetics to proceed. Producing more uniform stresses and strain rates, rather than a distribution of quantities (explosive drive), defining the environment in which the spall proceeds.

The “arrested”, explosively driven shock experiment was first conceptually designed to provide insight into the ability of the equation of state (EOS), plasticity models, and damage models to capture the high-rate deformation characteristics of Be and to assess the probability of material failure under characteristic states of loading. The final design was determined to minimize the tensile stresses generated during standard shock experiments. The focus

of these experiments was to determine if Be fails in shock compression and if so what were the mechanisms causing failure. Post test characterization of the Be samples in the “arrested” shock experiments was performed to evaluate the microstructure and failure behaviors seen in these experiments.

2 Background Be investigations

Initial plate impact experiments on Be show a spall strength for sintered material to range from ~0.78 GPa to ~1.3 GPa [7]. Plate impact experiments on S200F Be were performed on a 40-mm gun in the transmission geometry (i.e., allowing the shock waves to propagate through the sample to an observable interface). This geometry can be used to study the strength, fracture behavior (in compression or tension (spall)), and Equation-of-State (EOS) [4,5]. These experiments showed a brittle spall behavior with spall strengths in the range of 0.8-0.9 GPa. The results of these plate impact experiments were compared to simulations using both an Arbitrary Lagrangian Eulerian (ALE) and a Lagrangian analysis to interrogate the performance of the EOS and strength models. The solutions that matched the experiment most accurately were used in computational simulations that guided the design of two additional types of experiments. The first experiment was a shock-release-reschok plate-impact experiment. Details of the experiment are described elsewhere [4]. Simulations using a Lagrangian analysis for flyer plate velocities were successful at predicting the velocity and pressure response of the impact. The second experiment considered a Be sample that was subjected to an explosively-driven shock-wave

Fig. 1. Summary plot of explosively driven spall strength measurements. The curve is added for an aid to the eye.

configured in such a way as to minimize tensile loading. The results of these simulations will be described in more detail later in this paper.

Explosively driven shock experiments were also conducted to look for differences in behavior based on kinetics. The spall strength data obtained in this study via basic analysis is summarized in Fig. 1. The observation of spall on the von Neumann spike indicates that Be fails under very short-lived loading (i.e., very narrow triangle waves). It is worthy to note that the values obtained here are similar to those found for plate impact studies, 0.78 GPa to 1.3 GPa for plate impact versus 1.04 GPa to 1.83 GPa for explosively driven. It appears that the spall strength of beryllium does not appear to be dependent on either strain rate or the shape of the incident wave implying that the kinetics of the failure mechanism is rapid.

Finally, to summarize the results of the plate impact, shock/partial release/re-shock, and explosively driven shock experiments, the following conclusions can be drawn. First, the equation-of-state (EOS) parameters used for model simulations lead to accurate predictions of the overall experimental behavior. Second, the strength characteristics of the components of the impactor (Ta, 6061-T6 Al, and Cu) do not have a significant impact on the shape of the wave profile determined experimentally over the timescale of interest, i.e. until the arrival of the bulk re-loading wave. Third, the Be remains sufficiently intact following shock loading and partial release to support a shear stress following partial release and subsequent shock re-loading of the material. This leads to the conclusion that the material still exhibits strength through the Hugoniot elastic limit (HEL) at pressure. Finally, the velocity records reveal various features that the computational models cannot reproduce at this time.

3 Experimental configuration

In this study tests were performed on S-200F beryllium from Brush Wellman, Inc.. The total impurity level for this grade is typically below 1% and the grain size is about 11 μm [8]. Specimens were machined such that the samples were loaded along the hot-pressing axis. The crystallographic texture in such samples is modest at 2 or less multiples of random distribution [9].

Based on computational simulations using the ALE and Lagrangian analyses of experiments described above an iterative simulation process was undertaken to develop

Fig. 2. Shock recovery assembly for the arrested beryllium experiments.

the test assembly and predict the deformation during the experiment. The purpose of the simulations was to design an experiment that would provide the appropriate stress levels and cause a compressive shear failure rather than a tensile failure. The simulations were also used to determine the offset of the Be sample from the high explosive (HE) that would provide the appropriate pressure levels. The shock loading of the assembly was modeled using a plane-wave generator with a line detonation and a program burn model for PBX 9501.

The final design of the test assembly (Fig. 2) consists of Al cylinders surrounding the test sample and the HE drive assembly. The inner cylinder is comprised of upper and lower halves (gray and purple). The height of the upper piece was chosen to be 35 mm, 45 mm, or 55 mm and the height of the lower piece is adjusted to maintain the overall height of this inner cylinder at 100 mm. The test sample, either Mg or Be, was 5 mm thick by 25 mm in diameter, placing the leading edge of the sample at 30 mm, 40, mm and 50 mm from the explosive charge. The blue and red layers designed as radial and axial momentum traps. In the experimental configuration, the HE assembly, a polyethylene sleeve surrounding a detonator, plane wave lens (TNT) and a disk of PBX-9501 produced the explosive drive (shock/Taylor wave) to mimic the plane wave generator used for the Lagrangian simulations.

4 Simulation, recovery, and results

Lagrangian simulations of the final design analyzed the sample offset, the effectiveness of the momentum traps, and stress levels within the sample (Fig. 3). It was determined that the design would provide relevant information regarding the deformation of Be under the conditions of interest and that the axial and radial momentum traps were sufficient to prevent the Be sample from being subjected to large tensile stress.

4.1 High explosive drive validation

Drive validation experiments were done using AZ31B Mg to verify that the Be would be contained within the Al assembly and to measure the shock pressure and particle

Fig. 3. Simulations of the final Design of the HE driven experiments with a 30 mm Be offset to explore the stress state in the Be sample a) $t = 10 \mu s$, b) $t = 50 \mu s$.

Fig. 4. Magnesium “test” experiment showing stress concentrations (arrows) and non-uniform loading (curvature).

velocity. The shock Hugoniot calculated by Marsh [3] $u_s = 4.57 + 1.21 u_p$ mm/ μs with $\rho = 1.776$ g/cm³, for this Mg gives a peak shock pressure of 15.5 GPa. Based on the analysis of arrival times using Photon Doppler Velocimetry (PDV) probes, the samples were flat to within 45 nm, however, the potential of a slight tilt at impact was present and contact between the Be sample and the Al assembly was not intimate. Post-test evaluations showed that the design would contain the Be samples if they failed (Fig. 4). Impact of the magnesium with the LiF window resulted in a shock propagated into the LiF and a reflected shock back into the Mg. The Hugoniot match at a particle velocity of 1.11 mm/ μs and a pressure of 19.6 GPa. From this time on, the interferometer records the velocity of the Mg / LiF interface and a correction associated with the window must be applied. Using the suggested correction [10] just after impact, a particle velocity of 1.094 mm/ μs was calculated. This is in adequate agreement with the particle velocity calculated on the basis of the Hugoniot data for LiF and AZ31B magnesium. Details of the analysis for particle velocity and shock pressure are presented elsewhere [11].

4.2 Sample recovery

The recovered inner cylinder, plug, and Mg sample assembly is shown in Fig. 4. Note that the magnesium did not deform uniformly and that there are regions of stress concentrations (red arrows). However, the final geometry of the plug is remarkably similar to the results of the simulation at $t = 50 \mu s$ (above). It is expected that the Be

Fig. 5. a) Idealized schematic of crack orientation, b) Test sample #5 located at 30 mm from HE, and c) Test sample #7 located at 50 mm from HE. Missing pieces have been used for other characterization Discs are 25 mm in diameter.

samples will experience the same loading that caused this deformation in the Mg.

Visual observation of the Be samples after they were recovered from the aluminum fixture indicates that the cracks were nominally vertical, parallel to the direction of loading (Fig. 5). Radial cracks are seen to propagate outward but they do not appear to be continuous from the center of the disc to the edge. In fact, there seems to be a ring of damage nominally at half the radius that divides the center and outer damage zones. This ring may be a reflection of the stress concentrations observed in the Mg disc. Statistically it is likely that there is a physical reason for this damage because it is present in all of the Be tests. The Be discs shown in Fig. 5 have pieces missing that were used for other characterization.

4.3 Fractography

An investigation of the fracture morphology by Scanning Electron Microscopy (SEM) was performed to address the dominant fracture mechanisms. As the recovery assembly was specifically designed to minimize rarefaction waves from loading the sample to induce tensile failure, a focus of the fractography was to understand how and where nucleation of damage occurred. It is worth noting that no cracks were observed in the plane of the sample as would be expected for sample undergoing a simple spall loading such as reported by Bat'kov [12] and Golubev [13]. In all images the general appearance of the fracture surfaces is consistent with that of cleavage in a characteristically brittle material. At the macroscopic level, the surfaces are flat with sharp edges and uniform metallic luster, free of any shear lips that would be suggestive of ductile plastic deformation. At the microscopic level the failure is dominantly transgranular. Figure 6 shows the dominant surface morphology is hackle markings, which can be accurately used to determine the crack propagation path. There are a limited number of intersection scarps that are likely more indicative of the intersection of multiple initiated cracks in the specimens. Rib marks, are present but there are no clear Wallner lines, perhaps adding additional support to the observation that the samples were all very effectively momentum trapped and hence there were no inwardly reflected waves for the crack fronts to interact with. Consistent with literature on quasistatic crack growth in beryllium associated with either failure of a tensile specimen [14,15] or crack propagation of a fracture specimen [16–18], the hackle marking can be

Fig. 6. Representative fracture plane propagating from near the center of the sample on the left to the outer edge on the right (as illustrated in the inset (a)). (b) Composite of multiple SEM images of the fracture surface. (c) Map of the key features with crack direction as inferred from the hackle marking (in red) initiating at the top and propagating from the center towards the outer edge. This crack path is consistent with the rib marks (in green). Due to the multiple initiation points, the cracks come together at the two dominant intersection scarps (in blue) (sample 7–50 mm from HE).

Fig. 7. Fracture surface of a circular crack in sample #3 (40 mm from HE) ((a) SEM image and (b) fracture map, inset (c) illustrates surfaces being viewed). The crack clearly propagated from the top to the bottom. The shock is propagation from the top to the bottom.

traced back to the origin of crack initiation. Moreover, consistent with this literature, there is no gross transition in fracture mechanism from the area of the initiation and regions of fast crack propagation (such as the mirror to hackle transition typically observed in brittle glasses).

In addition to the sample having been successfully momentum trapped with no formation of a spall plane, a few gross observations can be made from the fracture surfaces (see Figs. 6 to 8). First, the cracks always initiate on the top surface of the sample, consistent with the hypothesis that the sample experienced some nominal bending that puts the top of the specimen in tension. Second, once initiated, the cracks consistently propagate from the center of the sample towards the outer edge. This crack propagation path directly contradicts the proposed failure mechanism observed in Taylor impact experiments where unconfined compression at the impact surface leads to a large expansion and resultant large tensile hoop stress. In the Taylor impact experiment this can lead to radial cracks propagating inward from the outer edge.

Fig. 8. Fracture surface of a radial crack in sample #7 (50 mm from HE) ((a) SEM image and (b) fracture map, inset (c) illustrates surfaces being viewed). The crack clearly propagated from the top to the bottom and from the intersection with a circular crack to the outer edge. The shock is propagation from the top to the bottom.

Unlike the idealized schematic of the crack orientation shown in Fig. 5, the actual crack patterns are not uniformly spaced and the circular cracks do not always form a complete ring. While it is clear that cracks initiate at the top surface and propagate outward consistent with the sample experiencing bending, the crack paths are clearly stochastic. Under quasi-static loading, beryllium follows linear elastic fracture mechanics (LEFM) with a relatively low fracture toughness (K_{IC} of 5 to 20 MPa m^{1/2} [16–19]) consistent with the observed brittle failure mechanisms. On fracture surfaces of the circular cracks in the current work (see for example Fig. 7), the cracks clearly propagate from the top to the bottom with no strong propagation component in the circumferential direction. Therefore, it is likely that the circular cracks grow independent of the radial cracks. On fracture surfaces of the radial cracks (see for example Fig. 8), the cracks clearly propagated from the top to the bottom and from the intersection with a circular crack to the outer edge. Therefore it is likely that the radial cracks either are initiated from the radial cracks or grow with a strong interaction with the radial cracks.

4.4 Metallography

Metallographic characterization shows that deformation twins are found in all samples. There also appears to be evidence of void nucleation at grain boundaries. Similar nucleation, growth and linkage may be the initiation of the cracks that can be seen in all samples. It is likely that the cracks start from void linkage creating a crack initiation site. The image below is representative of the feature generally seen in all samples (Fig. 9).

4.5 Electron backscatter diffraction (EBSD)

Samples were prepared and characterized using the following process. Portions of the bulk samples were cut into cross sectional wedges and fixed in metallographic mounts. These samples were ground and polished to 1 μm with a final polish of colloidal silica. Electron Backscatter Diffraction (EBSD) was performed using an FEI XL30 ESEM. Microscope conditions include a beam voltage of 20 kV, spot size of 5, aperture size of 100 μm, and a working distance of 16 mm. Patterns were acquired using a camera binning of 2×2, a pattern binning of 160, theta of 1, a capture rate of 2.4 frames per second with background

Fig. 9. Metallographic image of the Beryllium samples taken by grinding the fracture surface off, polishing the sample, and imaging the surfaces. These images were all taken using an optical microscope. Sample 5 on right (30 mm from HE) and sample 7 on left (50 mm from HE).

Table 1. Twin Area Fraction (Fraction of area twinned).

	Internal	edge
Sample #5	0.129	0.136
Sample #7	0.080	0.076

subtraction and normalization, and a 0.5 μm step size. The samples were analyzed at both the center of the sample and near the edge (approximately 200 μm from the “top” surface) of the sample.

Processing of the EBSD data was performed using OIM Analysis 6.1. The data was cropped to dismiss out of focus data and was processed by cleaning 10% of data points with the lowest confidence index (CI), where the CI is a measure of how well the diffraction patterns match various solutions for a given crystal structure. Twins were also recognized in the software using default values. Only HCP tensile twins of the character 94.8 degrees @ $\langle -12 - 10 \rangle$ with 5 degrees of tolerance were included in calculations. Also, edge data was removed from twin calculations.

Inverse pole figure (IPF) maps were calculated for both the plane normal of the sample cross-section, as well as for the shocked direction. Qualitative analysis of the plane normal IPF maps shows a random texture for both Samples 5 and 7 with slightly more twins observed in sample 5 (30 mm from HE) as compared with sample 7 (50 mm). IPF maps with reference to the shocked direction show a stronger $\langle 0001 \rangle$ texture in Sample 5. (Fig. 10). Similar maps were generated for the edge texture but only the results of the twin fraction are shown here (Table 1). Sample 5 again shows a $\langle 0001 \rangle$ texture in the shocked direction, where the Sample 7 exhibits a random distribution of orientations. In both cases, Sample 5 exhibited more twinning.

Using the parameters described previously, twin statistics were obtained; these can be found in Table 4.1. Sample 5 shows a higher twin area fraction throughout the sample area viewed. Twinning is the primary damage mechanism in Be, and indicates that Sample 5 was damaged more than Sample 7.

To determine relative amounts of residual strain in the materials, Kernel Average Misorientation (KAM) maps were calculated with a maximum misorientation of 5

Fig. 10. Inverse pole figures (Shocked Direction) for the “center” texture. Grains in Sample #5 show a stronger $\langle 0001 \rangle$ texture in the shocked direction than are seen in Sample 7 indicating a larger fraction of twins. Note: All scans were performed with a step size of .5 micron.

Fig. 11. Kernel Average Misorientations (2nd neighbor, perimeter) Be Sample #5 exhibits more misorientation, a sign of greater strain present within the sample. Note: Min and Max in legend are in degrees. Maximum misorientation = 5 degrees. Step size is 0.5 μm .

degrees. This calculation shows the degree of misorientation of data points throughout the grain with respect to the average grain orientation. KAM maps were calculated for both the nearest neighbor and for the 2nd nearest neighbor. Although both sets of maps show that sample 5 exhibits more residual strain than Sample 7 it is much more apparent in the 2nd nearest neighbor plots (Fig. 11).

5 Conclusions

Arrested Beryllium (Be) experiments were focused on determining if Be fails in compression and to observe

post-shock microstructure. The final design of the experiment was determined using the results from constitutive and spallation studies referenced in this paper. Post mortem characterization of the Be used in the “arrested” shock experiments was done to evaluate the microstructure and failure behaviors seen in these experiments.

Characterization of the arrested beryllium (and magnesium) samples using metallographic techniques shows several interesting characteristics. First, a visual characterization of the Mg sample shows regions of stress concentration that are likely reproduced in the Be sample. The inner stress concentration is likely to be the source of the load that created the circumferential crack in the Be samples. There were also radial cracks present in the samples that may have been caused by bending moments (see curved surface of the Mg sample as reference). These stresses likely generated enough tensile stress to crack the samples. General results of the fractography showed that there were no definitive sources of failure in the Be discs. Typically, fracture propagated from the side of the sample closest to the explosives, the side with the highest tensile bending stress. There was evidence that fracture may have propagated from the circumferential crack outward and downward radially, but that was not true in all cases. The EBSD results were the most informative of the characterization techniques used. EBSD provides information regarding texture, residual strain, and twinning. There was clear evidence of grain rotation as evidenced by the inverse pole figures.

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