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THE ROLE OF INTERFACES ON DYNAMIC DAMAGE IN TWO PHASE METALS

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Abstract. For ductile metals, the process of dynamic fracture during shock loading is thought to occur through nucleation of voids, void growth, and then coalescence that leads to material failure. Particularly for high purity metals, it has been observed by numerous investigators that voids appear to heterogeneously nucleate at grain boundaries. However, for materials of engineering significance, those with inclusions, second phase particles, or chemical banding it is less clear what the role of grain boundaries versus other types of interfaces in the metal will be on nucleation of damage. To approach this problem in a step-wise fashion two materials have been investigated: high purity copper, and copper with 1 % lead. These materials have been shock loaded at 1.4 GPa and soft recovered. In-situ VISAR and post mortem metallography reveals significantly less damage in the metals with no lead. The role of lead at grain boundary triple points and its behavior during shock loading will be discussed.

Keywords: Grain Boundary, Dynamic damage, Alloy
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INTRODUCTION

Many industries require materials that can tolerate extensive damage during dynamic loading. In order to design materials with specific response to dynamic damage it is important to understand the effect of microstructure on its spall response [1-3]. Previous dynamic damage experiments on single-phase ductile metals have shown that during shock and spall experiments voids preferentially nucleate at grain boundaries. Specifically, most of the damage is focused on non-special boundaries whereas there is no void nucleation at special boundaries [4]. Addition of second phase particles that possess a big size mismatch with respect to the primary phase can lead to segregation of these second phase particles to the grain boundaries. However, while these second phase particles at the grain boundaries are expected to change defect nucleation, the mechanism and evolution of this damage process is not well understood. For example, an interface between the two phases may be more susceptible to void nucleation as compared to the single-phase boundaries. Hence, this study aims to characterize the role of grain boundaries in comparison to other types of interfaces in a metal, on damage nucleation during dynamic loading.

EXPERIMENTAL PROCEDURE

The composition of the Cu-Pb alloy used in this study is listed in Table 1. The pure Cu samples used in this study were prepared from fully annealed 99.99 % pure oxygen-free high-conductivity (OFHC) copper and had the same pedigree as samples used in a prior study [4-5]. Both the Cu and Cu-Pb samples had an average

TABLE 1. Composition of Cu-Pb alloy (wt%)

Cu	Ni	Pb	Sn	Zn	Fe
98.75	0.01	1.24	0.01	0.01	0.01

grain size of 60 μm following annealing under vacuum at 600 °C for 1 hour. The microstructure of the annealed specimens was examined with scanning electron microscopy (SEM) and electron back-scattered diffraction (EBSD) with a FEI Inspect and Phillips XL30 respectively. Standard metallographic techniques were used to prepare samples for inspection. For EBSD and SEM a final polish of 1 μm Al_2O_3 was followed by an electrochemical polish using a solution of two parts phosphoric acid to one part water at ~1.9 V for 30-40 sec. The samples were also lightly etched using a solution of $FeCl_3$ and HCl.

Figure 1 shows an SEM and EBSD image of the annealed pre-shock microstructure for Cu-Pb. Details about the microstructure of the Cu samples can be found in Ref [4].

Figure 1. The image on the left shows an EBSD image with grains colored by orientation. The black spots correspond to Pb. The image on the left is an SEM image showing Pb inclusions at the grain boundary.

Figure 1 shows that most of the lead is concentrated at grain boundaries and triple junctions. Specifically most of the Pb is concentrated at non-Σ3 boundaries. However, it is important to note that the lead concentration over large sample areas is not homogeneous. Figure 2 shows that over large areas there are select regions where lead preferentially clusters in vertical lines or stringers.

Incipient spall experiments were performed using a smooth bore 80 mm light gas gun with a quartz flyer plate, which was 3 mm and 2 mm in thickness for Cu-Pb and Cu, respectively. The details of these experiments are discussed elsewhere [4]. The recovered samples were prepared for inspection in a Zeiss optical microscope, SEM and EBSD using the techniques discussed above.

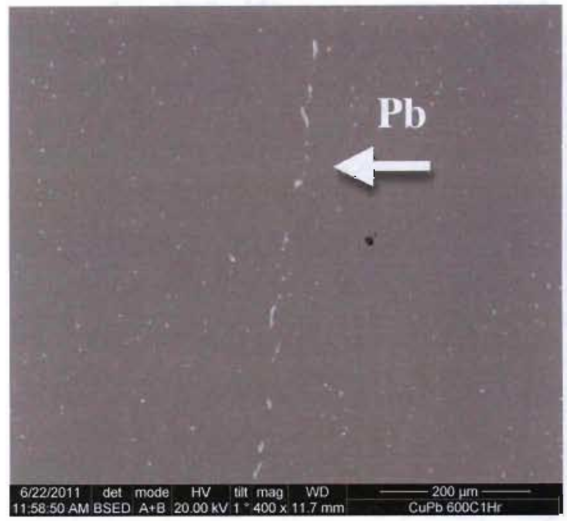

Figure 2. SEM Image showing a lead stringer in a Cu-Pb sample. Lead is colored white.

Quasi-static compressive tests performed for Cu and Cu-Pb yielded similar yield stress and work hardening response of the two materials as well as similar strain rate and temperature sensitivities. This is coincident with the similarities in their microstructures.

RESULTS AND DISCUSSION

Figure 3 shows the free surface velocity (FSV) histories for the incipient spall experiments on high purity Cu and Cu-Pb. The difference in free surface velocity from the peak state to the minima (as shown in region 1 in Fig. 3) is 38 m/s and 75 m/s in Cu-Pb and high purity Cu, respectively, showing that the spall strength is lower in Cu-Pb.

EBSD analysis of the spalled samples shows that majority of the voids nucleate at non-Σ3 boundaries in Cu-Pb. Similar analysis of the as-annealed Cu-Pb sample shows that majority of the Pb is also located at non-Σ3 boundaries. This is shown in Figure 4. Hence in conjunction with the

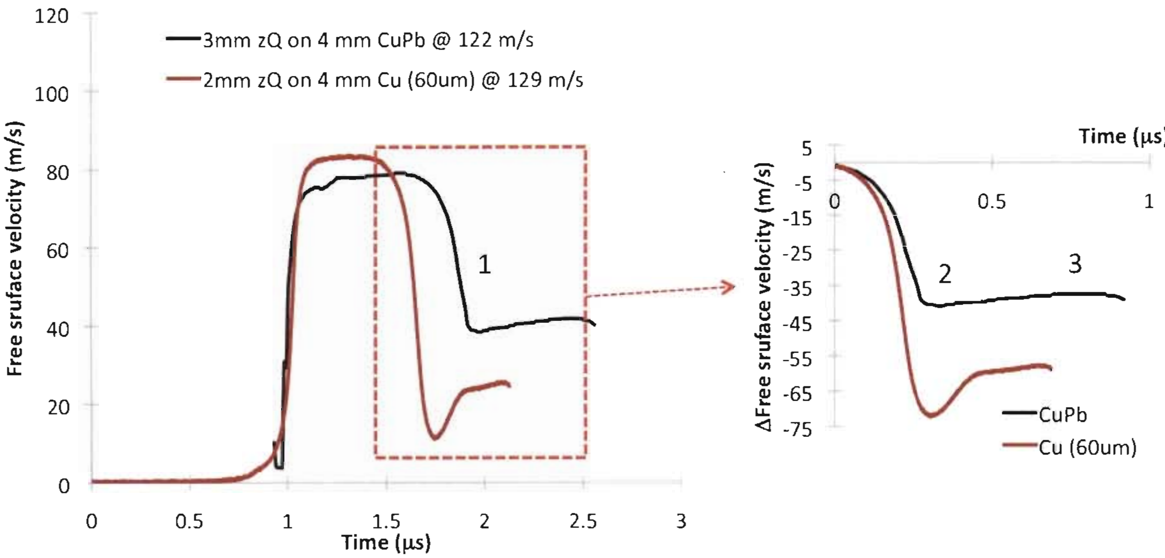

Figure 3. Free surface velocity histories for incipient spall experiments on Cu (red) and Cu-Pb (red).

EBSD data from the as-annealed undeformed sample it can be interpreted that majority of the voids are nucleating at grain boundaries with high concentrations of Pb.

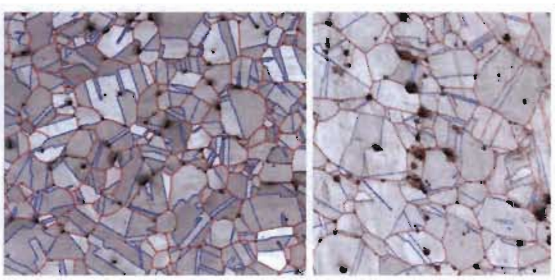

Figure 4. EBSD images of the as-annealed (left) and spalled (right) Cu-Pb sample. The grain boundaries are colored by rotation angle. Red and blue boundaries correspond to rotation angles between 2 to 58° and 59 to 61° respectively.

It is also interesting to note that differences are observed in the rate at which the velocity rises to the first peak beyond the minima (regions 2 and 3 in Fig. 3) between Cu and Cu-Pb. In order to make a comparison between Cu and Cu-Pb the times and velocities were shifted such that both the minima coincide with each other. Data from region 2 then suggests that high purity Cu specimen has more damage (higher spall peak) as compared to Cu-Pb and also that the rate at which this damage evolves is much faster in Cu as compared to Cu-Pb (region 3 in Fig. 3). This suggests that it is easier for voids to nucleate at the Cu-Pb interface as compared to Cu-Cu interface. However, once the voids nucleate they need plastic dissipation to grow. In Cu-Pb samples, since Pb is more ductile than Cu, it likely absorbs plastic work hence making it harder for voids to grow in the Cu-Pb sample. In the high purity Cu samples it is much easier for the voids to grow since it lacks a softer phase for plastic dissipation. Therefore void growth is the primary mechanism available for the material to dissipate plastic work.

Optical images in Fig. 5 also show that more damage is formed in Cu as compared to Cu-Pb if the vertical lines of damage in Cu-Pb are not taken Within the Cu-Pb specimen, vertical lines of damage exist due to the presence of lead stringers. Interestingly, the area fraction of voids for both the high purity Cu and Cu-Pb samples is similar, 0.25 % and 0.2%, respectively, when the lead stringers are not taken into account. However, the size of voids in Cu is larger as compared to Cu-Pb, which supports the plastic dissipation hypothesis mentioned above. Nevertheless, it is important to note that the pulse duration in the Cu sample is shorter as compared to the Cu-Pb sample. This is significant because it is well recognized even if it is not well understood, that for a given material, long pulse durations lead to enhanced damage [6]. Hence a direct comparison between this data might not be meaningful. However, the extrapolated data suggests that since Cu-Pb has the same amount of damage as high purity Cu even though the pulse duration was longer, that a shorter pulse duration test might have developed less damage. However, an impact experiment with the same pulse duration and compressive stresses needs to be conducted to test this hypothesis.

Figure 5. Optical images showing damage in high purity Cu and Cu-Pb after an incipient spall experiment. The top image corresponds to pure Cu whereas the bottom image is for Cu-Pb.

CONCLUSIONS

Plate impact experiments were conducted on Cu-Pb and were compared with high purity Cu to further elucidate the role of grain boundaries in comparison to other types of interfaces in a metal, on damage nucleation during dynamic loading.

Quasi-static compressive tests performed for Cu and Cu-Pb yielded similar yield stress and work hardening response. This is coincident with the similarities in their microstructures. The FSV data showed that Cu-Pb had lower spall strength in comparison to Cu. Metallographic and FSV data suggests that damage in Cu developed at a faster rate than in Cu-Pb.This could be because of lack of softer phase for plastic dissipation in high purity Cu. More impact experiments will be conducted to test this hypothesis.

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