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Effect of notch toughness and hardness on sliding wear of $Cu_{50}Hf_{41.5}Al_{8.5}$ bulk metallic glass

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Dry sliding wear experiments with a $Cu_{50}Hf_{41.5}Al_{8.5}$ bulk metallic glass were conducted following different annealing times of the glass at 510 °C. For as-cast and structurally relaxed samples and for the initial devitrification stage, wear is hardness controlled and improves with annealing. Upon longer annealing at 510 °C, the decrease in toughness increasingly controls the wear behavior. For prolonged annealing times, wear deteriorates and is toughness controlled. The wear stages can be quantified based on the theory of brittle wear.

Keywords: Amorphous alloys; Hardness; Toughness; Wear; Suction casting

Much controversy surrounds the question whether annealing improves or deteriorates the wear behavior of metallic glasses. Recent wear studies of Cu–Zr–Ti bulk metallic glasses (BMGs), for example, demonstrated a decrease in the wear resistance with annealing [1]. Such deterioration was attributed to the decline in fracture toughness following the annealing treatments. Sliding wear studies with $Cu_{50}Hf_{41.5}Al_{8.5}$ BMGs then demonstrated improvements in the wear resistance with annealing [2]. The discrepancies in these findings could reflect a dependence of wear on the competing effects of fracture toughness decrease and hardness increase during annealing of metallic glasses. Hardness improvements commonly indicate improvements in the wear behavior that at the same time can deteriorate with lower fracture toughness. Fracture toughness, however, is an important factor mostly for brittle wear when the fracture toughness values are less than about 10–20 MPa $m^{1/2}$ [3]. For tough materials, hardness is the main material parameter that controls wear. It might be expected then that the role of fracture toughness for wear of metallic glasses depends on the absolute values of the fracture toughness and the load. Lawn and Evans and, independently, Hagan developed an analytical relation between the critical load to nucleate [4], or propagate [5] subsurface flaws, the hardness, and fracture toughness of materials:

$$P^{*} = L \cdot \left[\frac{K_{IC}}{H}\right]^{3} K_{IC} \tag{1}$$

In Eq. (1) P^* is the critical load, K_{IC} is the mode I fracture toughness and H is the material hardness. The parameter L denotes a dimensionless parameter with a value of approximately 2×10^5 according to Ref. [3] and 2.2×10^4 according to Ref. [5]. Eq. (1) was developed strictly only for crystalline materials, but Hagan argued that the equation was valid also for glasses that reveal inhomogeneous flow. At room temperature, Cu-based glasses with glass transition temperatures well above 500 °C deform inhomogeneously and thus Eq. (1) should be valid for wear studies on metallic glasses.

Eq. (1) offers the intriguing possibility that the wear mechanism changes with annealing of metallic glasses. The mechanism would change when the critical load crossed the actual load during the sliding wear process. This prediction is tested with the current study. For each annealing condition, a set of hardness and notch toughness values is obtained and therefore also a critical load. The critical load is then compared with the applied load; significant changes in the wear behavior that could signal a change in mechanism are compared with the difference between the applied and critical loads.

$Cu_{50}Hf_{41.5}Al_{8.5}$ bulk metallic glass (BMG) was selected, mainly due to ease of synthesis of glassy 3 mm diameter rods and control of nanocrystal development. At an annealing temperature of 510 °C nanocrystals were identified in transmission electron microscopy

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(TEM) analysis after annealing for at least 30 min Ref. [2]. The detailed procedure for the preparation of the alloy, the annealing procedure and the test of amorphous-crystalline nature are reported elsewhere [2,6]. Microhardness measurements were performed using a Vickers diamond indenter with an applied load of 1 kg and a dwell time of 15 s. Linear reciprocating wear experiments were conducted on as-cast, annealed (at 510 °C for 15, 30, 60, 150, 300 and 450 min) and crystallized samples (750 °C for 300 min). A Si_3N_4 sphere of 3.2 mm in diameter was used as a counter-material, sliding on the metallic glass disk at a velocity of 1.2 mm s^{-1} and a load of 1.3 N. All experiments were performed at room temperature and in air. The wear track was analyzed using a Zygo optical interferometer to calculate the wear volume loss for different sliding distances. These results were used to calculate the dimensionless wear coefficient (Q) according to

$$Q = \frac{V_W \cdot H}{S \cdot N} \tag{2}$$

where V_W is the wear volume loss for sliding distance S under the applied normal load N for that material having a hardness H. Notch toughness experiments were conducted on $25 \times 2 \times 2$ mm samples in a three-point bending mode with an Instron 5869 mechanical test instrument under displacement control of 0.1 mm min^{-1}. Notches with a radius of about 100 μm were placed using a slow speed diamond saw. Between three and ten experiments were conducted for as-cast samples and for each annealing condition to gage the scatter in the data. A JEOL 6335 field emission scanning electron microscope operated at 10 kV was used to examine the fracture surfaces. The fracture energy (G_c) was calculated according to

$$G_c = K_c^2(1 - \nu^2)/E \tag{3}$$

where E is the Young's modulus and ν is the Poisson's ratio. The elastic properties were measured using a Magnaflux Quasar resonant ultrasound spectrometer following the procedures described in Ref. [7]. X-ray diffraction patterns, continuous heating differential scanning calorimetry curves and TEM micrographs for the as-cast and annealed samples are reported in Ref. [2].

The changes in hardness and notch toughness with annealing at 510 °C that are depicted in Figure 1(a) reflect two distinct regimes. The hardness increases by 3% for the first 30 min of annealing and by a total of 19% after 300 min of annealing, while the notch toughness decreases by 72% from a value of 74 MPa $m^{1/2}$ for the as-cast sample to about 21 MPa $m^{1/2}$ after 300 min of annealing. The value of 74 ± 14 MPa $m^{1/2}$ is slightly higher than the toughness values reported so far for Cu-based BMGs of 65 ± 10 MPa $m^{1/2}$ for a $Cu_{49}Hf_{42}Al_9$ BMG [8] and 67.6 MPa $m^{1/2}$ for a $Cu_{60}Zr_{20}Hf_{10}Ti_{10}$ BMG [9]. Data is included in Figure 1(a) for the completely crystallized alloy. Fully crystallized samples were too brittle for notches to be machined and therefore the notch toughness could not be determined. From the hardness and notch toughness data, the critical load can be determined according to Eq. (1) and the reported values for L. With a value for L of 2×10^5 that is taken from the literature [10], a contour map can be developed that indicates the critical

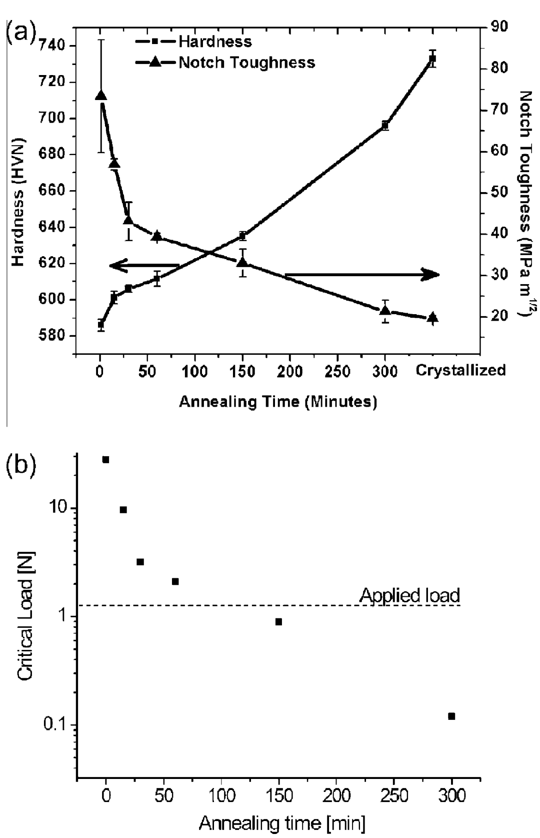

Figure 1. (a) Hardness and fracture toughness of the $Cu_{50}Hf_{41.5}Al_{8.5}$ BMG as a function of annealing time at 510 °C. (b) Critical load for brittle fracture as a function of annealing time.

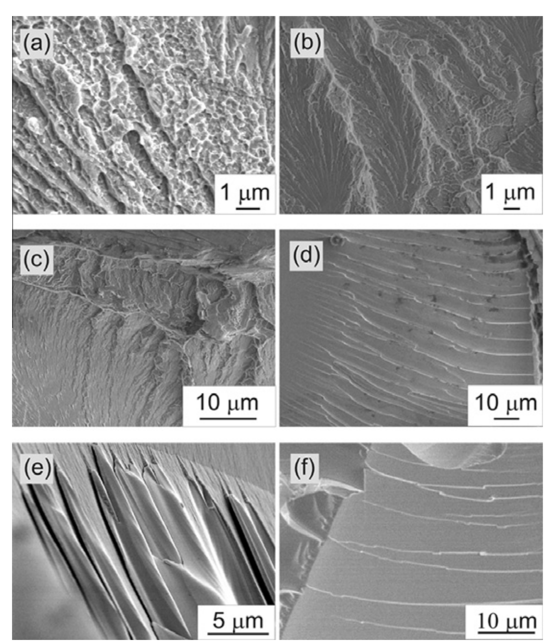

Figure 2. SEM fracture surfaces: (a) as-cast sample, with dimples and a chevron pattern morphology; (b) after 15 min; (c) after 30 min: vein patterns; (d) after 60 min: main and side branch veins are visible; (e) after 150 min; and (f) after 300 min annealing: mist zones with river vein patterned shear bands. All images were taken at the surface centers of the fractured samples.