

Metallic glass–steel composite with improved compressive plasticity



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ABSTRACT

A $Zr_{52.5}Cu_{18}Ni_{14.5}Al_{10}Ti_5$ bulk metallic glass toughened with a commercially available spring-shaped steel wire has been produced by centrifugal casting. The addition of the steel spring significantly affects shear band nucleation and propagation through the blockage, deflection and multiplication of shear bands at the glass–spring interface. As a result of the more homogeneous distribution of the plastic strain, the room temperature plasticity increases from 0.9% for the monolithic glass to about 4% for the glass–spring composite. Given the low volume fraction of the spring used in the composite (4.2 vol.%), these results demonstrate the extreme effectiveness of the steel spring for improving the plasticity of the metallic glass.

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1. Introduction

Bulk metallic glasses (BMGs) have received much attention within the last decades due to their large elastic limit and high strength compared to their crystalline counterparts [1]. Despite these advantages, the limited room temperature plasticity of BMGs is still a major drawback which hinders the utilization of these materials in engineering applications. Plastic deformation of BMGs occurs through the formation of highly localized shear bands, which propagate quickly resulting in catastrophic failure soon after yielding [1].

A way to improve the plasticity of BMGs is the creation of bulk metallic glass composites (BMGCs), where the presence of a second phase in the amorphous matrix improves the plasticity via restricting shear bands propagation and through the generation of multiple shear bands [2–4]. BMGCs can be classified into two main categories according to the processing route used [5]: in situ and ex situ composites. In the ex situ composites, micro-/nano-sized particles, fibers or wires are added to the glassy matrix by using melt infiltration [6] or powder metallurgy [7,8], whereas in situ composites are produced directly during solidification through the appropriate choice of composition or cooling rate [5,9,10]. Alternatively, in situ composites can also be produced by controlled heat treatment of the monolithic glass to precipitate micro-/nano-sized crystalline phases from the amorphous matrix [5]. Although the in situ processing has the merit to simplify the process, the ex situ processing gives more freedom in tailoring

the microstructure (e.g. size and volume fraction of the second phase).

Zr-based BMGs are of significant interest as glassy matrices in BMGCs thanks to their excellent glass forming ability and wide supercooled liquid region [2,11–14]. The second phases in these composites are fibers, whiskers or particles discontinuously distributed within the glassy matrix and their amount is rather large, generally exceeding 10 vol.%. Examples are the ex situ Zr-based BMGCs with second phases such as steel [2], W [2,14], Ta, Nb and Mo [12].

The homogeneous distribution of the second phases has a decisive effect on the mechanical properties of the resulting composites [15,16]. This is particularly critical for composites with discontinuously distributed second phases, where particles clustering may occur, consequently reducing their effectiveness as toughening or strengthening agents [17]. Recently, Wang et al. [18] have overcome this drawback through the creation of composites consisting of a BMG matrix and an open-cell Cu foam, which acts as a continuous three-dimensional deformable network. This approach is very effective for combating catastrophic shear banding of BMGs, given the extremely low volume fraction of the toughening second phase (4.2 vol.%): the room temperature plasticity increases from 2.5% for the monolithic BMG to 5.6% for the BMG composite [18].

In this work, we further examine this approach by using a steel spring as continuous second phase with reduced volume fraction to produce plastically deformable Zr-based BMGCs. The spring shape was selected in order to analyze the effect of a second phase with a well-defined geometry on the shear band evolution and resulting mechanical properties.

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2. Experimental details

BMG composites consisting of a glassy matrix with nominal composition $\text{Zr}_{52.5}\text{Cu}_{18}\text{Ni}_{14.5}\text{Al}_{10}\text{Ti}_5$ (at.%) and a commercially available spring-shaped steel wire were produced by centrifugal copper mold casting in the form of cylindrical samples with 4 mm diameter and 48 mm length. For this (see schematic illustration in Fig. 1(a)), the $\text{Zr}_{52.5}\text{Cu}_{18}\text{Ni}_{14.5}\text{Al}_{10}\text{Ti}_5$ alloy (produced by arc melting in a titanium-gettered argon atmosphere) was cast into the cylindrical copper die containing the steel spring (outer diameter ~ 2.8 mm, inter-ring spacing ~ 1.4 mm and wire thickness ~ 300 μm). For comparison, the monolithic $\text{Zr}_{52.5}\text{Cu}_{18}\text{Ni}_{14.5}\text{Al}_{10}\text{Ti}_5$ BMG was also produced using the same casting parameters (ejection temperature 1573 K; argon overpressure 100 mbar) as used for the glass–spring composite. Cylindrical specimens with aspect ratio of 2 (8 mm length and 4 mm diameter) were prepared from the cast rods and mechanically tested at room temperature using an Instron 8562 testing facility under quasistatic compressive loading (strain rate $\sim 1 \times 10^{-4} \text{ s}^{-1}$). Both ends of the specimens were carefully polished to make them parallel to each other prior to the compression test. The compressive strain was measured directly on the specimens using a Fiedler laser-extensometer. To obtain the volume fraction of the steel spring in the composites, the density of the spring and BMGC were determined using the Archimedes principle, which gives a volume fraction of steel of 4.2%. The microstructure of the samples and their surface morphology after the mechanical tests were investigated by scanning electron microscopy (SEM) using a Gemini 1530 microscope coupled with energy-dispersive X-ray (EDX) analysis. The amorphous nature of the matrix in the specimens was verified by X-ray diffraction (XRD) using a Philips PW 1050 diffractometer (Co K α radiation).

3. Results and discussion

Fig. 1(b) shows the SEM micrograph of the longitudinal cross section of the $\text{Zr}_{52.5}\text{Cu}_{18}\text{Ni}_{14.5}\text{Al}_{10}\text{Ti}_5$ BMG toughened with the steel spring. The spring (dark grey circles in Fig. 1(b)) is embedded in the glassy matrix (light grey area in Fig. 1(b)) and spirals continuously

along the sample. Most of the glass–spring interfaces are continuous and free of porosity (Fig. 1(c)); however, imperfect interfaces and porosity (indicated by an arrow in Fig. 1(d)) can occasionally be observed. The good glass–spring interface can be ascribed to the low volume fraction and to the simple shape of the spring, through which the melt can flow easily and fill available gaps between the steel wires. In addition, the low surface roughness of the spring most likely prevents the turbulent flow of the melt at the interface, avoiding the formation of gaps, which may be quenched in the sample as a result of the fast cooling rate. Another important factor for affecting the glass–spring interface is the small difference of the coefficients of thermal expansion between steel and the $\text{Zr}_{52.5}\text{Cu}_{18}\text{Ni}_{14.5}\text{Al}_{10}\text{Ti}_5$ BMG ($4.06 \times 10^{-5} \text{ K}^{-1}$ [19] and $3.9 \times 10^{-5} \text{ K}^{-1}$ [20], respectively), which may prevent debonding at the interface during cooling.

EDX compositional analysis (Fig. 2(a)–(c)) for the two main elements Zr (red) and Fe (yellow) indicates that no visible inter-diffusion of these elements between the glassy matrix and the steel spring occurs during sample preparation. The absence of inter-diffusion is in contrast to the results reported by Wang et al. [18], who observed significant Cu diffusion from the Cu foam into the Ti-based BMG matrix. This contrasting behavior can be ascribed to the difference of cooling rate between centrifugal and suction casting [21] and to the resulting time spent by the alloy in the liquid state, where diffusion is faster. The absence of a reaction between glassy matrix and spring is confirmed by the XRD pattern of the composite (Fig. 2(d)): the pattern displays the broad diffuse maxima characteristic of the monolithic glass along with a sharp crystalline peak belonging to steel. No peaks due to any additional phases are detected.

The room temperature compressive stress–strain curves for the monolithic BMG and glass–spring composite are shown in Fig. 3(a). The monolithic glass exhibits an elastic regime of 1.9% before yielding, which occurs at about 1650 MPa. After yielding the stress slightly increases with increasing strain up to fracture, which takes place at 1700 MPa stress and 2.8% strain. This results in a plastic strain of 0.9%. The addition of the steel spring remarkably affects the mechanical properties of the material. Although the yield strength and the elastic limit (1450 MPa and 1.65%) are reduced

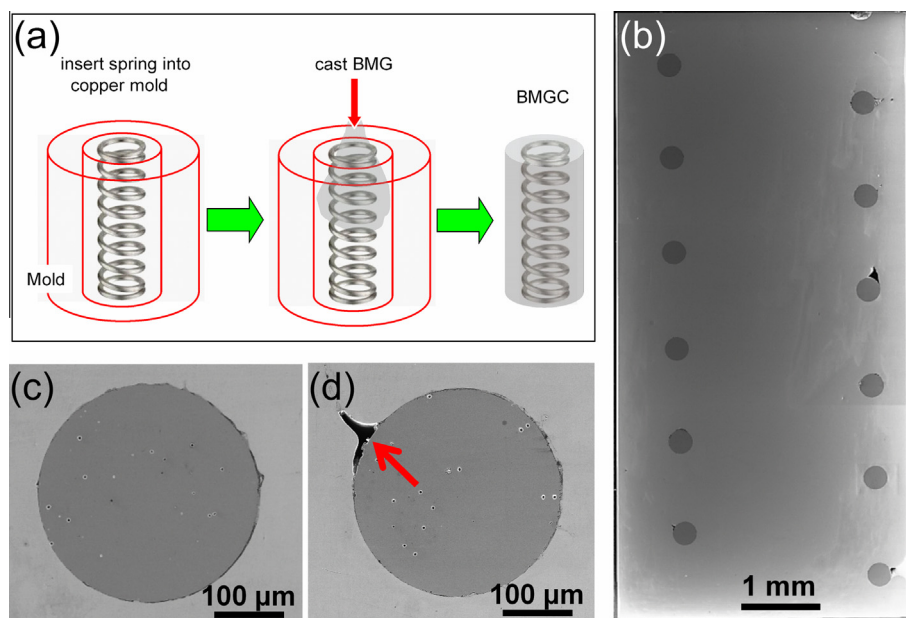


Fig. 1. (a) Schematic illustration of the preparation of the $\text{Zr}_{52.5}\text{Cu}_{18}\text{Ni}_{14.5}\text{Al}_{10}\text{Ti}_5$ glass–steel spring composite. SEM micrographs of (b) longitudinal cross-section of the glass–spring composite, (c) continuous, pore-free glass–spring interface and (d) interface with porosity (indicated by an arrow).

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