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Deformation-induced spatiotemporal fluctuation, evolution and localization of strain fields in a bulk metallic glass

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Deformation behavior and local strain evolutions upon loading and unloading of a bulk metallic glass (BMG) were systematically investigated by *in situ* digital image correlation (DIC). Distinct fluctuations and irreversible local strains were observed before the onset of macroscopic yielding. Statistical analysis shows that these fluctuations might be related to intrinsic structural heterogeneities, and that the evolution history and characteristics of local strain fields play an important role in the subsequent initiation of shear bands. Effects of sample size, pre-strain, and loading conditions were systematically analyzed in terms of the probability distributions of the resulting local strain fields. It is found that a higher degree of local shear strain heterogeneity corresponds to a more ductile stress-strain curve. Implications of these findings are discussed for the design of new materials.

Keywords: digital image correlation (DIC); strain heterogeneity; bulk metallic glass; inhomogeneous deformation

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1. Introduction

Bulk metallic glasses (BMGs) have been shown to have superior mechanical properties such as strength and ductility under geometrically constrained conditions, but are oftentimes brittle in unconstrained conditions such as uniaxial tension tests (Zhang et al., 2003; Schuh et al., 2007; Yang et al., 2010; Wang et al., 2012). The precursor of brittle failure is strain localization in thin bands or planes, which are referred to as shear bands. Despite many models and significant advances in recent decades (Spaepen, 1977; Argon, 1979; Johnson et al., 2005), there is still a lack of understanding of the structure-property relationship that controls the initiation of shear bands and their evolution into fracture failure. From the atomistic point of view, the shear bands are believed to originate at local structural heterogeneities such as free volume regions, shear transformation zones, or some sorts of soft regions (Spaepen, 1977; Argon, 1979; Maloney et al., 2004; Wang et al., 2012; Zhao et al., 2013). These are intrinsic factors that govern the deformation behavior. Macroscopically, the applied strains are accommodated by the localized deformation in shear bands (Lewandowski et al., 2006; Jiang et al., 2007; Fornell et al., 2009), so the ductile-versus-brittle behavior has good correlation with the strain partitioning in shear bands. That is, the promotion of shear band multiplication or prevention of shear band propagation tend to reduce the strain in individual shear bands and thus prevent catastrophic failure (Chen et al., 2013); these are the extrinsic design principles utilized in composites and coating approaches (Zhou et al., 2013). A connection between the above atomistic and macroscopic processes will clearly broaden our design capabilities in order to achieve better ductility and toughness (Schuh et al., 2007; Barrat et al., 2007; Homer et al., 2009). The critical information needed for this is

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5 the ability to visualize and quantify the full extent of shear banding, from initiation to
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7 propagation, to final fracture, across multiple length scales.
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10 The shear banding process has been found to be jerky in time and inhomogeneous in space
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12 (Jiang et al., 2008; Wright et al., 2013; Wu et al., 2011), which makes it difficult to fully capture
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14 its initiation, propagation, interaction and especially the evolution of local strain fields. Among
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16 the many methods applied in BMG studies, transmission electron microscopy (TEM) can probe
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18 structural information of the local deformed region, but the reduction of the sampled volume to
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20 nanometer scale and the exposure to electron beam might lead to artifacts that complicate the
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22 understanding of shear banding behavior. Scanning electron microscopy (SEM) can observe the
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24 evolution of shear bands patterns, but cannot easily capture the initiation of shear bands (Li et al.,
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26 2003). Statistical analysis of the serrated flow, observed on stress-strain curves, can shed light on
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28 the collective shear banding dynamics, but the evolution of individual shear bands cannot be
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30 uniquely obtained (Song et al., 2008; Sun et al., 2010). Infrared thermography can measure the
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32 temperature field that reveals the shear bands where heating occurs due to the severe plastic
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34 deformation in the shear bands (Yang et al., 2004), but this method is limited by its poor spatial
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36 and temporal resolutions. The above limitations make it difficult to achieve an *in situ*
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38 measurement of the entire strain field and its evolution.
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51 The digital image correlation (DIC) technique is a visualization method that tracks surface
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53 features on the test sample. Sequential digital images of sample surfaces are recorded during
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55 deformation and these images are post-processed to obtain strain fields based on a
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57 pattern-matching algorithm (Chu et al., 1985; Sutton et al., 2009). Because it provides a full-field
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strain map, the DIC method is ideal for *in situ* investigation of localized deformation behavior in BMGs (Zhang et al., 2009; Joo et al., 2013). For example, Joo et al. (2013) measured the shear band arrangements under different strain rates using DIC, and analyzed the dependence of deformation mode and fracture mechanism on strain rate. On the other hand, the local strain fields and their dynamic evolution during loading/unloading, which is the most significant advantage of the DIC method especially at the initial stage of shear band formation, have not been fully addressed in the literature. Therefore, in this work, from *in situ* DIC characterization of local strain field evolution and shear banding processes in a model BMG, we will compare the DIC strain measurements under different loading circumstances (such as tension versus compression, or tests on samples with different sizes), to explore the connections between the local deformation behavior and the macroscopic stress-strain curves in BMGs and to attempt to reveal the mechanisms that govern ductile versus brittle behavior in BMGs.

2. Experiments and Methods

2.1 Sample preparation

The nominal composition of BMG samples used in this work is $\text{Zr}_{52.5}\text{Ti}_5\text{Cu}_{17.9}\text{Ni}_{14.6}\text{Al}_{10}$ (at%), denoted as BAM11 (Loffler et al., 2000). Master alloys were prepared by arc melting the pure elements (>99.99%) under a purified argon atmosphere. Each button was turned over and re-melted five times to maximize chemical homogeneity. Subsequently, cylindrical rods with diameters of 7 mm were obtained by casting the molten BMG into water-cooled copper molds under helium atmosphere. Compression specimens with an aspect ratio (height/diameter) of 2

and diameters of 2, 3, or 5 mm (denoted as D2mm, D3mm, and D5mm in these figures) were cut from the center of the as-cast rods. Stress relaxation of different-sized specimens was conducted by loading the specimens to a specific strain and then recording the stress variation while keeping the strain constant. Tensile specimens with gauge dimension of 10 mm × 1.5 mm × 1 mm were cut from the as-cast 10-mm rods. Pre-strained specimens were prepared by first compressing a disk-shaped sample with diameter of 7 mm and aspect ratio of 0.8 to a thickness reduction of 50%. Details of the pre-straining process can be found in Bei et al. (2006). Then samples with dimensions of 2 mm × 4 mm (diameter × height) were taken along the radial direction perpendicular to the compression direction. The amorphous structure was confirmed by x-ray diffraction.

2.2 *In situ mechanical tests with DIC*

The test specimens were ultrasonically cleaned in acetone, coated with a thin layer of white spray paint, and then a black toner powder was sprayed on to form a speckle pattern suitable for DIC measurements (e.g., Fig. 1a).

Compression/tension tests were conducted at an engineering strain rate of $5 \times 10^{-4} \text{ s}^{-1}$ on an Instron 5881 machine. A high-resolution camera, model GRAS-50S5M-C, was used to record the speckle images during deformation at a speed of 5 images per second. The images were then analyzed by the Vic-2D 2009 software, which provided strain values. Since in-plane displacement fields are recorded, we have strains in longitudinal and transversal directions being ϵ_{yy} and ϵ_{xx} , respectively. Shear strain ϵ_{xy} in the DIC results is half of the engineering shear strain. The quality of the speckles, subset size and step size during calculation can influence the

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5 resolution of the strain map. A subset of 39 and step size of 2 pixels were used for our strain
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8 calculations, leading to a spatial resolution of 6 μm for the strain measurements.
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10 11 12 13 **3. Results** 14

15 16 *3.1 Strain field mapping during shear band formation* 17

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19 The stress versus time curve under compression of a BAM11 specimen with a diameter of
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21 3 mm in Fig. 1b clearly reveals the serrations (i.e., stress drops) near the macroscopic yield
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23 point and beyond. Throughout this paper, positive values are used for the compressive stress
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25 and strain for clarity purpose. The strain fields corresponding to the different deformed states
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27 labeled by the numbers in Fig. 1b are given in Fig. 1c. Before the macroscopic yield point,
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29 small serrations (e.g., 1 to 2) result from the formation of tiny shear bands. At Point 5, the
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31 engineering stress reaches the maximum, and the transition from Points 5 to 6 corresponds to
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33 the appearance of new shear bands on the surface, as well as to the increase of strain magnitude
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35 in the pre-existing shear bands that are present at Point 5. It should be noted that the observed
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37 shear bands are the intersection of the “shear plane” and the front surface of the specimen, so
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39 various inclined angles can be found in the images. For each shear band that appears in these
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41 optical micrographs, there is a corresponding stress drop on the stress-strain curve. While some
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43 shear bands may propagate gradually, others can burst across the whole sample and
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45 intermittently slip with further loading (Song et al., 2008; Sun et al., 2010; Chen et al., 2009).
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57 Representative strain maps of ϵ_{xx} , ϵ_{yy} , ϵ_{xy} , and ϵ_{Mises} , after the formation of many shear
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59 bands (corresponding to the point labeled “+” in Fig. 1b), are shown in Fig. 2. The signs of
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these strains in the shear bands vary, e.g., ε_{yy} is negative while ε_{xx} is positive for the loading condition in Fig. 2. To distinguish between local strains inside and outside the shear bands, representative locations in a shear band (“S” in Fig. 2b) and in the adjacent matrix (“M” in Fig. 2b) were monitored. The evolution of local strains upon loading is shown in Figs. 2e and 2f. It can be seen that the local longitudinal strains in the shear band and away from it increase at nearly the same rates until the macroscopic strain (obtained as part of the DIC measurement) reaches around 2% (which corresponds to the macroscopic yield point on the load-displacement curves). Afterwards, the local longitudinal strain within the shear band dramatically increases while, at locations away from the shear band, it keeps almost unchanged, indicating that the shear deformation was mostly localized in the shear band.

It can be concluded that the plastic deformation after shear band formation is both spatially and temporally heterogeneous. That is, the overall strain fields are localized spatially in narrow bands, and these bands exhibit intermittent “stick-slip” behavior on the stress versus time plots. Once a major shear band forms, there is little that can prevent or delay its propagation, except using extrinsic methods such as geometric constraints, second phases, and low-aspect-ratio samples, among many others. If the intrinsic ductility needs to be improved, a question that naturally arises is how does a shear band develop starting from the early stage of heterogeneous strain fluctuations?

3.2 Histograms of the local shear strain fluctuations

To evaluate the evolution of local strain fluctuation during deformation, statistical analysis of the local strain fields at different deformation stages was performed for different-sized

BAM11 specimens, and the sample with diameter of 2 mm was taken as the representative example in this section. It should be noted that the surface of the cylindrical sample was curved in the x-direction, and the displacement vector will also have a component in the z-direction. The displacement along z direction can be taken as an out-of-plane motion from the xy plane and may affect the value of shear strain on xy plane. To reduce the effects from the displacement along z direction, only the central part of the sample was analyzed for the quantitative statistical analysis, where the arc angle in the x-z plane is restricted within $\pm 15^\circ$. Based on previous investigations [Sutton et al, 2008; Haddadi and Belhabib, 2008; Lava et al, 2011], the effects of z-direction displacement on the strain components in xy plane can be analyzed as in the Appendix and was estimated to be around 10^{-5} in the current investigation, which is consistent with the prediction from previous works and one order of magnitude less than the shear strains in our current statistic analysis. Results off the central part provide qualitative measure of the strain distributions. From the strain plots in Figs. 1 and 2, irreversible local deformation and large fluctuation of local strain are found even in what may be considered the macroscopically elastic region of the stress-strain curve and before any shear band becomes visible, which is consistent with previous studies (Park et al., 2009; Ye et al., 2010; Homer, 2014). Thus a state corresponding to a stress below the global yield stress ($\sim 85\%$ of σ_y) was chosen (Point A in Fig. 3a), for a comparison to the global yield point (Point B in Fig. 3a) and to the plastic deformation stage near fracture (Point C in Fig. 3a). The probability of finding a certain value of local shear strain can be obtained from statistical analysis, and can be described by a Gaussian distribution (Wu et al., 2009):

$$p(\epsilon) = p_0 + p_1 \exp \left[-\frac{(\epsilon - \bar{\epsilon})^2}{2\sigma^2} \right] \quad (1)$$

where p_0 and p_1 are fitting parameters, $\bar{\epsilon}$ is the average shear strain, and σ is the standard deviation. A larger σ corresponds to a larger fraction of local sites having strains different from the average value, implying a larger degree of strain heterogeneity. Fitting parameters to the Gaussian distribution are listed in Table 1. It can be seen that the peak height decreases but the width slightly increases (i.e., σ increases from 0.121 to 0.162) when the stress increases from $0.85\sigma_y$ to σ_y (i.e., Points A to B in Fig. 3a), and this trend continues from Points B to C. Additionally, several distinct discontinuities at strain magnitudes up to 2% have been observed (Fig. 3b), which correspond to the formation of shear bands. Therefore, it can be seen that the global deformation was actually a manifestation of the fluctuation and evolution of local strain heterogeneity. The applied strain is not uniformly accommodated in the entire solid, but by a statistical distribution of local strains that may extend to values that are far different from the applied strain (or the mean strain).

3.3 Sample size effects in metallic glasses

As shown in Fig. 3a, the decrease of the specimen size leads to an increase of the ductility as observed in Bei et al. (2010) and Wu et al. (2010). Samples with diameter of 2mm were usually used as the representative of small-sized BMGs when studying mechanical properties [Joo et al., 2013, Wu et al. 2010]. Herein, samples with diameters of 2mm and two larger-sized samples (D=3, 5mm) were used to study size effects of BMGs. Probability distribution curves are shown in Figs. 3b-3d for samples with diameter of 2mm, 3mm, and 5mm, respectively. All these curves display the same features of evolution, i.e., the increase of width and the decrease

of height with respect to the applied strain. However, the width increases (i.e., the increase of σ) from Point A to Point B for the 3-mm and 5-mm samples are much smaller than that of the 2-mm sample (as shown in Table 1), indicating a lower degree of strain heterogeneity (i.e., a relatively smaller fraction of local sites that deviate significantly from the mean value) during this stage. This trend can be seen more clearly from a comparison of the distribution curves near fracture of the three different-sized samples in Fig. 3e. These results suggest that a larger σ , i.e., a larger degree of strain heterogeneity, corresponds to a higher ductility.

3.4 Pre-strain effects on local strain heterogeneity

According to the above correlation between the strain heterogeneity and the degree of ductility, it is anticipated that a processing that promotes structural heterogeneity may result in an improved degree of ductility. This hypothesis is examined in the context of pre-strain effects (He et al., 2008; Wu et al., 2010b). The compressive stress-strain curves of the pre-strained and as-cast samples are shown in Fig. 4a. It can be seen that pre-straining dramatically increases the ductility before failure. The comparison of local shear strain distribution curves for the as-cast and pre-strained samples at different stages (marked as A, B and C in Fig. 4a) are given in Figs. 4b-4d. At Point A, not only does the pre-strained sample have a wider distribution of local strains (as listed in Table 1), there are also a few distinct points with strains larger than 2% (i.e., multiple shear bands induced by pre-strains). At Points B and C in Figs. 4c and 4d, the pre-strained sample has several stand-alone points with strains up to 30%. It is anticipated that these shear bands will not lead to fracture because an increased number of shear bands will reduce the strain on each shear band and thus delay fracture, or equivalently speaking, the

increased strain heterogeneity accommodates the applied strain more easily.

3.5 Tension-compression asymmetry in metallic glasses

Tension-compression asymmetry has been a long-standing topic in the study of deformation behavior of metallic glasses (Zhang et al., 2003; Schuh et al., 2007). Our experimental results in Fig. 5a show that the specimen under tension fractures at around 1690 MPa, which is 93% of the compressive yield strength of about 1810 MPa. The fluctuations and evolutions of local strain fields are given in Figs. 5b and 5c at different stages. At Point A, the specimen under tension has nearly the same probability distribution curve as that of the compressive specimen. However, the tensile sample shows much narrower distribution curve near failure (Fig. 5c), indicating not only a lower degree of local strain heterogeneity but also almost no change of this local strain heterogeneity from Points A to B. Again, these results show the correspondence between the evolution of the local strain heterogeneity and the ductile/brittle behavior of the mechanical tests.

4. Discussion

The structure of BMGs is spatially heterogeneous and consists of atomic structural heterogeneities, or loosely bonded and tightly bonded regions that respond differently to the external stress (Wagner et al., 2011; Liu et al., 2011; Li et al., 2013). The evolution of these atomic heterogeneities has been visualized as the reshuffling of atomic clusters that need to overcome a certain energy barrier. On the other hand, DIC results in Figs. 3-5 show a relationship between the degree of local strain heterogeneity and the macroscopic ductility. It is

naturally interesting to investigate whether there is a relationship between the local strain heterogeneity and the atomic structural heterogeneity.

The statistical nature of atomic structural heterogeneity dictates that the energy barrier has a probability distribution (Wu et al., 2009; Rodney et al., 2009). In a recently developed thermal activation model (Wu et al., 2009), BMGs are regarded as an ensemble of shear transformation zone (STZ) embryos with a finite activation energy v that has a probability distribution of $p(v)$. The activation rate is given by, $\Gamma(v) = \gamma \exp(-v/k_B T)$, where γ is a pre-factor and is independent of energy and temperature T , and k_B is the Boltzmann constant. For a stress relaxation experiment, the normalized stress evolves as

$$\bar{\sigma}(t) = 1 - \int_0^\infty p(v)[1 - \exp(-\Gamma(v)t)]dv \quad (2)$$

where $\bar{\sigma}(t) = \frac{\sigma(t)}{\sigma_0}$, and σ_0 is the stress at the beginning of the relaxation. Details in the derivation of the above equation can be found in the previous work (Wu et al., 2009). By fitting stress relaxation results to Eq. (2), the distribution of activation energy, $p(v)$, can be obtained, and thus we obtain a knowledge of atomic structural heterogeneity.

Stress relaxation experiments at the stress of Point B for the 2-mm and 3-mm samples were conducted, and the evolution curves are given in Fig. 6a. The corresponding probability distribution curves of the STZ activation energy are calculated by Eq. (2) and given in Fig. 6b. The activation energy of the smaller sample exhibits a lower average value and a slightly narrower distribution, indicating that a larger fraction of local regions has a lower activation energy and can be more easily activated. In other words, in the smaller sample, there is a larger

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5 fraction of local regions where the local strains vary from the average local strain, leading to a
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7 larger degree of local strain heterogeneity. Results in Fig. 6 suggest that the local mechanical
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9 heterogeneity in DIC measurements is qualitatively in accord with the atomic structural
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11 heterogeneity of BMGs, The evolution of local mechanical heterogeneity during deformation
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13 can be taken as a manifest of the evolution of local structural heterogeneity, e.g., flow defects
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18 [Flores et al, 2007]. However, the DIC strain measurements do not have the resolution to
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21 capture the evolution of atomic or sub-micron strain fields.
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24 Although it is tempting to relate the local strain heterogeneity *intrinsically* to the atomic
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26 structural heterogeneity, the subsequent evolution of DIC-measured strain fields depends on a
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28 number of both *intrinsic* and *extrinsic* factors. During the plastic deformation near and after
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30 Point B in Figs. 3-5, the shear bands initiate, propagate, and eventually lead to fracture.
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32 Intrinsically, if the starting local-strain heterogeneity has a broad probability distribution (due to,
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34 say, pre-strains, alloying, or other means), the strain fields will tend to evolve into multiple
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36 shear bands, and fracture is less likely to occur in the early stage of the shear banding process.
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38 For example, as shown in Fig. 4c, even if there are some stand-alone points at high strains (that
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40 correspond to shear bands), the pre-strained sample still deforms without failing prematurely.
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42 Extrinsically, external geometric constraints altering mean or local stress field [Flores and
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44 Dauskardt, 2001] (such as low aspect-ratio samples under compression, the introduction of
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46 precipitates or second phases, etc.) will prevent the accumulation of a large strain on each shear
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48 band, so as to delay fracture. From the above discussion, it is obvious that the local strain
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65 heterogeneity can be used as a measure of the ability of BMGs to deform in a ductile or brittle

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5 manner. While the initiation of this fluctuation might depend on the atomic structure of the
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8 material, the subsequent evolution can be manipulated by both intrinsic and extrinsic means.
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10 11 12 13 **5. Conclusions**

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15 Using *in situ* digital image correlation, local strain fields in BMGs and their evolution have
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17 been investigated. Quantitative analysis of DIC-measured strain fields provides a unique
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19 approach for a better understanding of the heterogeneous deformation in metallic glasses. The
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21 local strain field is found to be both temporally and spatially inhomogeneous. Significant strain
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23 fluctuations and irreversible local strains are observed even before macroscopic yielding occurs,
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25 although eventually the total applied strain is mainly localized in one or a few shear bands.
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27 Statistical analysis of the spatial fluctuations of local shear strains has shown the roles played
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29 by the local strain heterogeneity and its evolution. A wider distribution of the local strain
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31 distribution, i.e., a larger degree of strain heterogeneity, corresponds to a larger degree of plastic
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33 deformation. Any factors that may affect the initiation and evolution of local strain
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35 heterogeneity, *intrinsically* or *extrinsically*, such as sample size, pre-strains, loading conditions,
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37 can therefore affect the degree of ductility in these BMGs.
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56 Energy Sciences, Materials Sciences and Engineering Division.
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Appendix: Effects of curved sample surface on strain measurements

When using DIC technique to measure the surface strains of a curved sample surface, there are two possible sources that affect the accuracy of strain measurements. First, strain values calculated from the recorded digital images are components in the xy plane, while the true surface strains are in the tangent plane that makes an angle of α with the x-axis, as shown in Fig. 7. A straightforward tensor transformation suggests that difference is about $\cos^2 \alpha$, which is very small since our data processing in Figs. 3-5 were restricted to the center line of the sample. Second, the out-of-plane displacement (i.e., z-direction displacement) may also affect the measured strains. Since the analysis of this source is nontrivial, presented below is an adapted discussion based several previous investigations [Sutton et al, 2008; Haddadi and Belhabib, 2008; Lava et al, 2011].

As shown in Fig. 7, when a displacement of $(\Delta X, \Delta Y, \Delta Z)$ moves the point P to P*, the measured strain components in the xy plane may be affected by ΔZ , since this displacement component cannot be captured by the camera. In the xz plane as given in Fig. 7(c), a geometric relationship gives $h/2 = (1 - \cos \alpha) R/2$, where α is inclined angle, R is the radius of the sample. Following Sutton et al. [2008], in a standard lens system, Z is the object distance, and (X, Y) the object in-plane coordinates. Denote L as the image distance, and $(x_s(Z), y_s(Z))$ as the in-plane image coordinates. The in-plane displacement fields due to ΔZ are given by

$$U(\Delta Z) = x_s(Z + \Delta Z) - x_s(Z) \approx -\frac{L}{Z} X \left(-\frac{\Delta Z}{Z} \right) = -\frac{\Delta Z}{Z} x_s,$$

$$V(\Delta Z) = y_s(Z + \Delta Z) - y_s(Z) \approx -\frac{L}{Z} Y\left(-\frac{\Delta Z}{Z}\right) = -\frac{\Delta Z}{Z} y_s, \quad (3)$$

leading to strain components of about $-\Delta Z/Z$. In our current setup, the object distance Z is 300mm, the maximum out-of-plane translation, ΔZ , can be estimated as $h/2=(1-\cos\alpha)\times R/2$. The statistic analysis was focused on the central part of the sample and $\alpha \approx 15^\circ$, so that the maximum shear strain caused by the displacement along z direction can be estimated as $5.7 \times 10^{-5} - 1.4 \times 10^{-4}$ when the sample diameter varies from 2mm to 5mm. These estimates are consistent with the observations in previous works [Sutton et al, 2008; Haddadi and Belhabib, 2008; Lava et al, 2011].

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Table Captions

Table 1 Fitting parameters for the Gaussian distribution function of local shear strains for different cases, including the changes of specimen sizes, pre-strain history, and loading conditions.

cases parameters	2mm			3mm			5mm		Pre-strained 2mm			Tension	
	A	B	C	A	B	C	A	B	A	B	C	A	B
p_0	0.167	0.149	0.116	0.118	0.113	0.154	-0.049	0.186	0.126	0.128	0.130	0.159	0.128
$\bar{\varepsilon}$	-0.001	-0.022	-0.011	-0.095	-0.089	-0.083	-0.059	-0.017	-0.098	-0.110	-0.177	-0.042	-0.052
	0.121	0.162	0.247	0.145	0.148	0.169	0.128	0.133	0.226	0.305	0.343	0.112	0.119
p_1	2.67	2.01	1.28	2.47	2.42	1.85	2.84	2.50	1.447	0.95	0.69	2.94	2.87

Figure Captions

Fig. 1 (a) Optical image of a representative BMG specimen with surface speckles for *in situ* DIC measurements. (b) A portion of the compressive stress versus time curve of the sample with a diameter of 3mm. (c) Strain fields of ε_{yy} at different times from DIC measurements. Image numbers correspond to those marked in (b).

Fig. 2 Strain contour plots near fracture (labeled by “+” in Fig. 1b): (a) ε_{xx} , (b) ε_{yy} , (c) ε_{xy} , and (d) Mises strain. The evolution of local longitudinal strain (e) and shear strain (f) inside and outside the shear band, as marked by locations “S” and “M” in (b).

Fig. 3 (a) Engineering stress-strain curves of BMG samples of the same aspect ratio but with different diameters. (b)-(d) Probability distributions of local shear strains are plotted at different stages for BMG samples with diameter of 2 mm in (b), 3 mm in (c), and 5 mm in (d), respectively. (e) Comparisons near the final failure points of these three samples.

Fig. 4 (a) Engineering stress-strain curves of the pre-strained (thickness reduction of 50%) and as-cast BMG samples. (b)-(d) Comparisons of the probability distributions of local shear strains at different stages as marked in (a).

Fig. 5 (a) Engineering stress-strain curves of compression and tension tests. The compressive stress is given in positive values for comparison purpose. (b) and (c) Comparisons of the probability distributions of local shear strains at different stages.

Fig. 6 (a) Stress relaxation responses (measurements given by solid curves and simulations by circles) for BMG samples with diameters of 2 mm and 3 mm at the corresponding yield points (i.e., Points B in Fig. 4a). (b) The corresponding probability distributions of

activation energy for these two cases.

Fig. 7 (a) Schematic illustration of a displacement on the curved sample surface and its components in the x, y, z directions, (b) schematic illustration of the selection of the investigated central region on the curved sample surface, and (c) the geometric relationship in the xz plane.

Fig. 1

(c)

Fig. 2

Fig. 3

Fig. 4

(a)

(b)

(c)

(d)

Fig. 5

(a)

(b)

(c)

Fig. 6

Fig. 7