Mg-based bulk metallic glass composites with plasticity and gigapascal strength

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Abstract

The lightweight Mg65Cu7.5Ni7.5Zn5Ag5Y10 bulk metallic glasses (BMG) always fail before yielding, with irreproducible strengths <650 MPa. Adding particles of TiB2 as the reinforcing phase, which can be uniformly distributed in the matrix without adversely affecting its glass-forming ability upon copper mold casting, the BMG composites can sustain a plastic strain to failure of 2–3% at a compressive fracture strength as high as 1.3 GPa. The strengthening/toughening effects of the second phase particles are attributed to their ability to serve as obstacles to the propagation of run-away shear bands and micro-cracks, as well as to the generation of multiple shear bands due to the large elastic mismatch between the TiB2 and the matrix. The angle between the compressive shear plane and loading axis is $\frac{\pi}{4}$, consistent with the Mohr–Coulomb criterion. The very high specific strength of $3.5 \times 10^5$ N m kg$^{-1}$ exceeds by far those of commercial Mg or Al alloys and even Zr-based BMGs.

Keywords: Metallic glasses; Mechanical properties; Magnesium; Composites

1. Introduction

Over the past decade, bulk metallic glasses (BMGs) have attracted considerable attention [1,2]. This interest in BMGs has been particularly motivated by their potential application as new high-strength structural materials. However, all BMGs face a challenging problem in that they plastically deform by the formation of highly localized shear bands, making them susceptible to catastrophic failure without much macroscopic plasticity [3–8]. To solve this problem, BMG composites (mostly with Zr-based BMG as the matrix) with ductile metal or refractory ceramic particles as reinforcements were developed to hinder the propagation of run-away shear bands and encourage the formation of multiple shear bands [9–15]. Such composites exhibit enhanced compressive, and occasionally even tensile, plastic strains and significantly improved impact toughness compared to monolithic BMGs.

For Mg alloys, the BMG route is certainly an attractive way to overcome one of their major drawbacks, i.e., their relatively low strength compared with common engineering materials such as steels and aluminum alloys. By reaching the amorphous state, one gets rid of dislocations, the carrier for yielding in crystalline alloys. This results in high strength in the low-density Mg alloys, i.e., an impressive specific strength. This quality, together with their low cost and ability to be recycled, mates Mg alloys should be highly desirable for applications. Indeed, since the first Mg-based BMG (Mg65Cu25Y10, in atomic percentage) was fabricated [16], extensive efforts have been devoted worldwide to
locating Mg-based BMG formers in several multi-component systems [17–23]. Recently, fully glassy rods close to one centimeter in diameter have been fabricated by using copper mold casting [20,22]. These Mg-based BMGs exhibit high compressive fracture strength up to 800 MPa [19,22,23]. The specific strength, defined by the ratio of fracture strength $\sigma_f$ to density $\rho$, is about $3 \times 10^{4}$ N m kg$^{-1}$, almost doubling the highest value for conventional Mg-based crystalline alloys. But here again, as is the case for other BMGs, the Mg-based BMGs suffer from low ductility problems. In fact, they have been found to be the most brittle in comparison with other BMGs and always fail in the elastic regime with no observable plasticity. Although our recent work [24] showed that macroscopic plasticity could be achieved through the formation of in situ Mg-BMG composites containing iron as second phase dispersions, such metallic reinforcements have limitations, including corrosion problems, increased density and the need for immiscibility with the matrix.

The present work explores the design, fabrication, and properties of Mg-BMG composites formed by adding ceramic reinforcements. The $\text{Mg}_{65}\text{Cu}_{7.5}\text{Ni}_{7.5}\text{Zn}_{15}\text{Ag}_{5}\text{Y}_{10}$ BMG-forming alloy [20] was chosen as the matrix material. TiB$_2$ particles were employed as the second phase, based on our preliminary work [25]. TiB$_2$ has several desirable properties. It is highly stable, with a melting point of 3173 K, and is insoluble with the matrix alloy. As such, Mg-based BMG composites with TiB$_2$ particulates at several volume fractions can be fabricated easily using melt casting. No reaction with the matrix or corrosion problems are expected. The ceramic addition also has a relatively low density ($4.5 \times 10^{-3}$ kg m$^{-3}$) not very different from the matrix. In terms of mechanical properties, the choice of a ceramic such as TiB$_2$ has advantages as well. The reinforcing agent has a strength (9–10 GPa, estimated based on $\sigma_f = 1/3H_v$) much higher than the matrix. Meanwhile, the very large difference between the elastic modulus of TiB$_2$ (529 GPa) and that of the matrix Mg–BMG alloy (50–70 GPa) causes a large elastic strain mismatch upon loading, generating stress concentrations that would make it favorable for shear bands to initiate at the particle/matrix interfaces. Such controlled formation of multiple shear bands is expected to enhance the overall plasticity. A systematic account of the microstructures, thermal stability and mechanical behavior in uniaxial compression of these composites is presented in this paper.

2. Experimental

Elemental pieces (>99.9% purity) were used as the starting materials. Cu–Ni–Ag–Y ingots as an intermediate alloy were prepared by arc melting under a Ti-gettered argon atmosphere in a water-cooled copper crucible. This alloy was then melted with Mg and Zn pieces by induction melting under inert atmosphere to obtain a master alloy with the desired composition. Subsequently, TiB$_2$ particles were added into the master alloy, which was induction melted under a purified inert atmosphere. The volume fraction, $V_f$, of the added TiB$_2$ particles varied from 10% to 30%. The average particle size was 10 $\mu$m and the oxygen content was 0.1 wt%. The composite re-melted in a quartz tube was then injected in a purified inert atmosphere into the copper mold to obtain rod samples 4 mm in diameter and 50 mm in length, or strip samples 4 mm in thickness, 8 mm in width and 30 mm in length.

The cross-sectional surfaces of the as-cast rods were examined using X-ray diffraction (XRD) in a Rigaku D/Max 2400 diffractometer with monochromated Cu $K\alpha$ radiation. The thermal stability of the glass and composites was investigated using a Perkin–Elmer differential scanning calorimeter (DSC-7) under flowing purified argon. The samples were contained in graphite pans. A heating rate of 20 K/min was employed. A second run under identical conditions was used to determine the baseline after each run. The as-cast rods were also examined in a Cambridge S360 scanning electron microscope (SEM) with energy dispersive X-ray (EDX) analysis.

The compression test samples were either cylindrical rods 8 mm in height and 4 mm in diameter, cut from the as-cast rods using a diamond saw, or rectangular columns 6 mm in height and 3 mm in edge length for the square-shaped cross-section. The rectangular samples, which facilitate the observation of shear bands on the surfaces, were cut from the as-cast strips by electrical discharge machining into actual test sizes. The side surfaces were mechanically polished with diamond paste to a mirror finish to minimize surface effects. The loading surfaces were polished to be parallel to an accuracy of less than 10 $\mu$m. Room-temperature compression tests for at least five samples of each composition were carried out using a strain rate of $1 \times 10^{-4}$ s$^{-1}$. The strain was determined from the platen displacement after correction for machine compliance. The morphology of the fracture surface was observed with SEM. Vickers microhardness $H_v$ was measured using a load of 200 g, averaging over $>20$ measurements. The density of the as-cast samples was determined using the Archimedes’ method by weighing in toluene and in air.

3. Results

3.1. Characterization of TiB$_2$-containing composites

Fig. 1(a) shows XRD patterns taken from the cross-sectional surfaces of the as-cast rods of the $\text{Mg}_{65}\text{Cu}_{7.5}\text{-}$
Ni$_{7.5}$Zn$_{5}$Ag$_{5}$Y$_{10}$ monolithic (single-phase) glass and the composites containing TiB$_2$ particles with different V$_f$.

For the composites, the XRD patterns show diffraction peaks from TiB$_2$ superimposed on the broad diffuse scattering maxima from the amorphous matrix. The introduced TiB$_2$ particles have the expected crystal structure and lattice parameters, apparently exerting no negative influence on the glass-forming ability of the matrix alloy.

DSC scans of the monolithic glass and TiB$_2$-containing composites are shown in Fig. 1 (b). In all the traces, an endothermic signal associated with glass transition and several exothermic events due to crystallization are observed. The crystallization of the monolithic glass, or the matrix glass in the composites, is a complex process that is completed through at least four steps. For all the samples, the glass transition temperature, $T_g$, the onset temperature of primary crystallization, $T_{x1}$, and the width of supercooled liquid region, defined as $\Delta T_x$ ($\Delta T_x = T_{x1} - T_g$), are listed in Table 1, together with the heat of crystallization obtained by integrating the area under the exothermic peaks, $\Delta H_x$. It is noticed that the TiB$_2$-containing composites exhibit identical thermal properties as the monolithic glass, within experimental error. This suggests that the matrix in the composites is nearly the same as the monolithic glass. In other words, there is no obvious chemical reaction between the TiB$_2$ particles and the matrix alloy to change the glass-forming ability of the latter (and the thermal properties of the glass formed). This is different from the cases of adding SiC into Cu$_{47}$Ti$_{34}$Zr$_{11}$Ni$_8$ [26] or Zr$_{57}$Nb$_{5}$Al$_{10}$Cu$_{15.4}$Ni$_{12.6}$ BMGs [11], where considerable silicon or carbon atoms were dissolved into the matrix.

Fig. 2 (a)–(d) shows the SEM micrographs of the polished cross-sectional surfaces of the as-cast Mg$_{65}$Cu$_{7.5}$Ni$_{7.5}$Zn$_{5}$Ag$_{5}$Y$_{10}$ monolithic glass and the composites with different V$_f$. For the monolithic glass, no crystalline phase can be observed except for occasional Y$_2$O$_3$ inclusions. An example is given in Fig. 2 (a). In the composites, the TiB$_2$ particles are uniformly dispersed throughout the glassy matrix, and the interparticle spacing between the particles decreases with increasing V$_f$, as seen in Fig. 2 (b)–(d). The average chemical composition of the matrix, measured using EDX, is close to the nominal composition, Mg$_{65}$Cu$_{7.5}$Ni$_{7.5}$Zn$_{5}$Ag$_{5}$Y$_{10}$. There is no evidence of any thick reaction layer (micrometers) at the interface. Moreover, there is little if any observable porosity in the SEM micrographs.

3.2. Compressive properties of TiB$_2$-containing composites

Fig. 3 shows the uniaxial compressive stress–strain curves for the Mg$_{65}$Cu$_{7.5}$Ni$_{7.5}$Zn$_{5}$Ag$_{5}$Y$_{10}$ glass and the composites. The slope of the linear part of the curves provides the elastic modulus, $E$, for each material. The obtained mechanical properties are listed in Table 2. As seen in Fig. 3 (a), the monolithic glass exhibits elastic behavior only, failing without any macroscopic yielding and plastic strain, i.e., $\varepsilon_p = 0$. The $\sigma_t$ varies in the range of 490–650 MPa for the five samples tested, comparable
with those of Mg-based BMGs at other compositions [19,22,23]. The irreproducible test results are a sign that the strength is controlled by flaws and premature brittle failure before reaching the elastic limit, and the material behaves more or less like a ceramic. The cracks are possibly initiated by minor flaws, pre-existing in the as-cast sample or formed in incipient local shear bands upon local yielding. There is apparently no mechanism that prevents the initiation and fast propagation of major cracks. Upon failure, the sample shatters into pieces or is crushed into fragments.

For the BMG material to be useful, it must show macroscopic plasticity with clear yielding. This was achieved by adding TiB₂ second phase particles. With 10 vol% hard particles, the composite shows a slightly higher elastic strain limit and an $\sigma_f$ of about 990 MPa. This strength is a factor of 1.6 higher than the monolithic glass. Yielding is barely visible. The $\sigma_f$ continued to increase at $V_f = 20$ vol%. Meanwhile, a clear yield point is observed. The ultimate fracture strength and plastic strain $\varepsilon_p$ to failure of the composites with $V_f = 20$ and 30% reached 1200 MPa and 3.2% and 1330 MPa and 1.8%, respectively. Among all the samples, the composite at $V_f = 20\%$ exhibits the largest $\varepsilon_p$, while the highest $\sigma_f$ was attained in the composite at $V_f = 30\%$. Note that in these composites, the strength is now reproducible and consistent for the several samples tested, Table 2. The strength level of well above 1 GPa is impressive and attractive for applications, because it exceeds the strength typical of commercial Ti alloys and most steels, let alone Al and Mg alloys.

Fig. 2. SEM backscattered electron images taken from the cross-sectional surface of the as-cast rods: (a) Mg₆₅Cu₇.₅Ni₇.₅Zn₅Ag₅Y₁₀ metallic glass, and the composites with (b) 10%, (c) 20% and (d) 30% TiB₂ particles.

Fig. 3. (a) Compressive true stress–strain curves of as-cast rods loaded to failure for the monolithic Mg₆₅Cu₇.₅Ni₇.₅Zn₅Ag₅Y₁₀ metallic glass, in comparison with (b) composites with different volume fraction of TiB₂ particles. The curves are shifted relative to each other for clarity.
In addition, serrated plastic flow was observed in the compression stress–strain curves of the composites with \( V_f = 20\% \) and \( V_f = 30\% \), as seen in Fig. 3 (b). To our knowledge, such a serrated flow has never been observed in Mg-based BMGs before, but only in BMGs that exhibit plasticity, for example in Pd–Ni–P [27,28], Pd–Cu–Si [29] and Zr–Ti–Ni–Cu–Be BMGs [28,30]. The origin of serrated flow in metallic glasses is generally attributed to the formation of shear bands, e.g., nucleation of new shear bands rather than slip on existing ones [31]. Therefore, the appearance of serrated flow in the TiB\(_2\)-containing composites is indicative of the formation of multiple shear bands in plastic deformation. Microscopic examinations (next section) support this conclusion.

The \( \sigma_f \) and specific strength \( \sigma_f/\rho \), \( H_v \) and measured elastic modulus of the composites increased with increasing \( V_f \), as plotted in Fig. 4 (a)–(c). The correlation of \( \sigma_f \) with \( H_v \) is in close agreement with the Tabor relation of \( \sigma_f = 1/3 H_v \) for all the samples, monolithic or composite. Additionally, the variation of \( H_v \) with \( V_f \) significantly deviates from the rule of mixtures (RoM, see dashed line in Fig. 4 (b)). The highest \( \sigma_f/\rho \) of about \( 3.6 \times 10^5 \text{ N m kg}^{-1} \) is achieved at \( V_f = 30\% \). The modulus of the composites was calculated using the RoM, 

\[
E_c = fE_p + (1 - f)E_m, \tag{1}
\]

where \( E_c \), \( E_m \) and \( E_p \) are the modulus for the composite, matrix glass and the second phase particles, respectively, and \( f = V_f \), as well as the Halpin and Tsai equation (H–T Eq.) [32]:

\[
E_c = \frac{E_m (1 + \xi \eta f)}{(1 - \eta f)}, \tag{2}
\]

where \( \eta = \frac{E_m / E_p - 1}{E_m / E_m + \xi} \) and \( \xi \) is a factor to account for the influence of geometry of the reinforcing phase, taken as 2 here for spherical particles. The measured \( E_c \) in Fig. 4 (c) significantly deviates from the RoM prediction. It is apparently different from the case of tungsten or steel fiber reinforced Zr–Ti–Cu–Ni–Be BMG composites [33], where RoM seemed to be adequate. Furthermore, a better match is obtained when using the H–T Eq.

### 3.3. Observation of shear bands and fracture morphology

Due to its inherent brittleness, the Mg\(_{65}\)Cu\(_{17.5}\)Ni\(_{17.5}\)Zn\(_5\)Ag\(_5\)Y\(_{10}\) monolithic glass shattered into fragments easily during compressive testing. In contrast, for the TiB\(_2\)-containing composites, the lower portion of the fractured sample remained as a whole piece. This is a clear sign of ductility improvement. Fig. 5 (a) and (b) displays the SEM micrographs showing the morphology of the fractured samples of the composite with 20 vol\% TiB\(_2\), in the cylindrical and rectangular geometry, respectively. For all these fractured samples, the angle between the fracture surface (shear plane) and the compressive loading axis, \( \theta \), was in the range 40.2–42.2\(^\circ\), averaging to 41.3\(^\circ\).

Table 2

<table>
<thead>
<tr>
<th>Materials</th>
<th>( e_y (%) )</th>
<th>( e_p (%) )</th>
<th>( E ) (GPa)</th>
<th>( \sigma_f ) (MPa)</th>
<th>( \sigma_f/\rho ) ( \times 10^5 \text{ N m kg}^{-1} )</th>
<th>( H_v ) (GPa)</th>
</tr>
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<tr>
<td>Monolithic glass</td>
<td>1.70</td>
<td>0</td>
<td>39 ± 2</td>
<td>490–650</td>
<td>1.80</td>
<td>2.60 ± 0.05</td>
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<tr>
<td>+10% TiB(_2)</td>
<td>2.35</td>
<td>0</td>
<td>47 ± 4</td>
<td>992 ± 39</td>
<td>2.93</td>
<td>3.12 ± 0.09</td>
</tr>
<tr>
<td>+20% TiB(_2)</td>
<td>2.38</td>
<td>3.2</td>
<td>50 ± 3</td>
<td>1212 ± 16</td>
<td>3.42</td>
<td>3.43 ± 0.21</td>
</tr>
<tr>
<td>+30% TiB(_2)</td>
<td>2.18</td>
<td>1.8</td>
<td>61 ± 1</td>
<td>1336 ± 15</td>
<td>3.59</td>
<td>4.14 ± 0.34</td>
</tr>
</tbody>
</table>

Fig. 4. (a) Fracture strength \( \sigma_f \) and specific strength \( \sigma_f/\rho \), (b) microhardness and (c) elastic modulus of the composites vs. volume fraction of TiB\(_2\) particles.
Fig. 6 displays the SEM pictures of the surfaces of a deformed sample with \( V_f = 20\% \). The compressive load was applied up to a plastic strain of 2\%, followed by unloading without failure. As seen in Fig. 6(a), slip was observed on multiple and intersecting planes. Cracks have nucleated at some slip traces. Fig. 6(b) shows a view of shear steps on the lateral surfaces, as marked by arrows. Fig. 6(c) demonstrates shear bands running between the particles, suggesting that the TiB\(_2\) particles may initiate and/or stop shear bands. Note that with a large modulus mismatch between the particles and the matrix, a large difference in elastic strain is expected upon loading of the composite. This causes stress concentrations at the particle/matrix interfaces, encouraging the emission of shear bands when the softer matrix yields.

The distance traveled inside the matrix by each of the emitted local shear bands is limited due to the constraints imposed by the high population of particles. Also, it is noted that shear bands were created not only in one direction. Multiple nearly parallel shear bands oriented at an angle not far from 45\(^\circ\) to the loading direction are observed on the surfaces of the deformed samples. These concurrent shear events are presumably responsible for the macroscopic plastic deformation observed.

In addition to promoting multiple shear band formation, the TiB\(_2\) particles also play a role in arresting catastrophic shear failure. Both of these effects enhance the observable macroscopic plastic flow before failure. Fig. 7 shows a micrograph of a crack formed in the failed composite sample with \( V_f = 20\% \). The propagation direction of the crack is quite consistent with the shear direction. It is seen that, when encountering the TiB\(_2\) particles during propagation, a crack either cuts through the particle (marked with arrows A) or bypasses the particle around the edge (marked with arrow B). Evidently, the TiB\(_2\) particles block the crack propagation in the matrix. Meanwhile, the fact that the crack penetrates through the particles also reflects the strong bonding at the particle-matrix interfaces. Such blocking/bypassing actions are also expected for shear bands before cracking initiates inside the bands, as can also be found in Fig. 6(c). When a shear band is fully stopped at an obstacle or has to travel around the obstacle, additional stress build-up and energy would be required, making a runaway major shear band and hence catastrophic failure less likely.

SEM micrographs of the typical fracture surfaces of the Mg\(_{65}\)Cu\(_{7.5}\)Ni\(_{7.5}\)Zn\(_{7}\)Ag\(_{3}\)Y\(_{10}\) monolithic glass and the composites with different \( V_f \) are displayed in Fig. 8(a)–(d). For the monolithic glass, the fracture surface consists of two distinct zones, as seen in Fig. 8(a). One is smooth and featureless. The other shows a vein pattern...
Fig. 7. Crack propagation in a composite with 20% TiB₂ particles, observed using SEM.

Fig. 8. SEM micrographs of the fractured surfaces for (a) Mg₆₅Cu₇.₅Ni₇.₅Zn₅Ag₅Y₁₀ metallic glass, and composites with (b) 10%, (c) 20% and (d) 30% TiB₂ particles.
characteristic of BMGs. The smooth areas inside the vein zones may be the initiation regions, and the smoothness might be the result of shear deformation that occurs prior to fracture. The large smooth and flat areas suggest that shear propagated in an unimpeded fashion over rather large distances in the monolithic glass. In contrast, in the TiB₂-containing composites, the well-developed vein patterns are dominant, accompanied by a number of molten droplets due to adiabatic heating, as observed in Fig. 8(b)–(d). Furthermore, as the $V_t$ increased, the length scales of the vein patterns was gradually reduced, and the number of the molten droplets increased. At $V_t = 30\%$, the melting occurred over a large area and the vein patterns were covered by the molten zone, as seen in Fig. 8(d). These changes reflect the fact that increased TiB₂ population resulted in a significant increase of energy release during fracture. This is in agreement with the large increase of fracture strength and macroscopic plastic deformation, as seen in Fig. 3(b).

### 4. Discussion

The deformation and failure of metallic glasses, at a temperature below $T_g$, are inhomogeneous and operated by localized shear bands [27,29,34–36]. However, the origin of the reduction of viscosity and softening in the shear band has been controversial. Local adiabatic heating was proposed to lead to a temperature rise beyond $T_g$, or even the melting temperature, reducing the viscosity by several orders of magnitude [34]. Later models [36–38], however, assume accumulation of free volume or dilatation in the shear bands, leading to a local lowering of the viscosity. Recently, Wright et al. [30] estimated the temperature increase in a shear band to be $<102\,K$ due to local adiabatic heating for the Zr₄₀Ti₁₄⁻²Ni₁₀Cu₁₂Be₂₄ BMG. Significant heating was predicted only for the final failure event. This result supports the notion that shear localization is more likely to be caused by changes in viscosity associated with increased free volume in the shear bands, rather than by adiabatic thermal softening.

The serrated flows observed signals the formation of multiple shear bands. Cracking is unlikely to be the cause because the load drops in the serrated flow region are very small and typical of those known for shear band formation in BMGs. The localized shear band propagation did not immediately cause major cracking because it was restricted by the second phase particles.

According the von Mises yield criterion, slip should occur on the plane of maximum resolved shear stress, i.e., at $45^\circ$ from the loading axis for an isotropic material loaded uniaxially. As seen in Fig. 8, for our BMG-based composites the shear failure plane is oriented at an angle of $41^\circ$ but not $45^\circ$. Similar observations have been made before for a number of other metallic glasses loaded in uniaxial compression. The fact that BMGs and their composites share this common behavior suggests that once unstable fracture is initiated, the crack propagation proceeds in the same direction, although some cracks may be slowed down or even stopped by the presence of particles. Assuming the failure plane corresponds to the plane where shear banding occurred upon yielding, the fracture angle is representative of the shear band angle. The deviation from $45^\circ$ suggests that the von Mises criterion may not be applicable, similar to the cases of Pd- [39,40] and Zr-based BMGs [30,41,42]. Yielding dependent on the normal stress on the shear plane may be modeled using the Mohr–Coulomb criterion, which predicts that

$$\tau_c = k_0 - \alpha \sigma_n,$$

where $\tau_c$ is the shear stress on the slip plane at yielding, $k_0$ and $\alpha$ are constants and $\sigma_n$ is the stress component in the direction normal to the slip plane. The magnitude of $\alpha$ depends on both the size of the structural units involved and the strength of the bonding between them, whereas $k_0$ depends primarily on the bond strength [39]. Using the value of $\theta = 41.3^\circ$, $\alpha$ was estimated from $\alpha = \frac{\text{const}}{\text{an}^{0.20-0.25}}$ [43] to be 0.13. This value compares well with the value of 0.11 ± 0.05 for Pd₄₀Ni₄₀P₂₀ metallic glass [39] and of 0.105 for Zr₄₀Ti₁₄Ni₁₀Cu₁₂Be₂₄ [30].

The fracture surface morphology of vein patterns has been observed as typical for metallic glasses due to the fracture mechanism based on the Taylor instability [44,45]. Meanwhile, evidence for local melting, in the form of droplets, was also observed in many metallic glasses [28,44,46–49]. Note that it is not true that such features are necessarily associated with a temperature spike within a shear band during its propagation. As a matter of fact, the local melting may have occurred only during the final failure event since the magnitude of the load drop during failure is much larger than that during

![Fig. 9. Comparison of the specific fracture strength $\sigma_f/\rho$ of Mg₆₅Cu₁₇.₇–Ni₇.₅Zn₅Ag₀.₅Y₁₀ metallic glass composites with 30% TiB₂ particles with representative commercial alloys as well as Zr-based BMG.](image-url)
serrated flow [28,30]. Also note that the melting over a large area seen here and recently for a Pd–Ni–Cu–P BMG [50] is rare. The increase of melted area is likely due to the increased $V_f$ of TiB$_2$ and the associated large increase of energy release at the moment of fracture.

Finally, the Mg$_{65.5}$Cu$_{7.5}$Ni$_{7.5}$Zn$_{5}$Ag$_{5}$Y$_{10}$/TiB$_2$ metallic glass composite maintains a very high specific strength of $\sim 3.5 \times 10^5$ N m kg$^{-1}$. Fig. 9 shows a comparison of $\sigma_p / \rho$ ($V_f = 30\%$) with those of some commercial alloys, including the 316L stainless steel, cast magnesium alloy (AZ91C), high-strength steel (HT 100), aluminum alloy (A7075) and titanium alloy (Ti–6Al–4V), as well as the popular Zr-based Vitreloy representative of BMGs. The specific strength of the present composites is higher than all these alloys, at a level very few, if any, alloys based on an engineering metal are able to rival.

5. Conclusions

1. Mg–BMG based composites containing TiB$_2$ particles can be processed readily through copper mold casting. The second phase particles are uniformly dispersed in the glass matrix, without marked interfacial reactions during processing or chemical composition change in the matrix. The composite recipe and processing have no adverse effect on the glass forming ability of the Mg$_{65.5}$Cu$_{7.5}$Ni$_{7.5}$Zn$_{5}$Ag$_{5}$Y$_{10}$ alloy.

2. The strong TiB$_2$ particles not only reinforce but also promote plasticity through multiple shear band formation due to the large elastic mismatch with the matrix. In addition to controlling plastic instability (shear banding), the obstacles tackle crack propagation as well. Microscopic examinations confirmed the interaction of particles with shear bands and the arrest of catastrophic failure. As a result, the compressive fracture strength increased from an irreproducible 590 MPa for the Mg$_{65.5}$Cu$_{7.5}$Ni$_{7.5}$Zn$_{5}$Ag$_{5}$Y$_{10}$ monolithic glass to consistently above $\sim 1200$ MPa for the composites. These super strong Mg alloys also exhibit clear yielding and serrated flow to appreciable plastic strains, no longer failing in a manner similar to ceramics.

3. The angle between the compressive shear plane and loading axis was about 41°. This finding is consistent with that for a number of other BMGs and composites and supports the proposal that the von Mises criterion may not be applicable.

4. Our composite design takes advantage of the light weight of both the glassy matrix Mg alloy and the ceramic second phase. The specific strength of $\sim 3.5 \times 10^5$ N m kg$^{-1}$ is among the highest for all the existing alloys based on an engineering metal, including all commercial light-weight alloys based on Mg, Al, and Ti, as well as high-strength steels and Zr-based BMGs.

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