Deformation-induced martensitic transformation in Cu–Zr–(Al,Ti) bulk metallic glass composites

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Plastic deformation of Cu–Zr–(Al,Ti) bulk metallic glass (BMG) composites induces a martensitic phase transformation from the B2 to the B19' CuZr phase. Addition of Ti to binary Cu–Zr increases the temperature above which the B2 CuZr phase becomes stable. This affects the phase formation upon quenching in Cu–Zr–Ti BMG composites. The deformation-induced martensitic transformation is believed to cause the strong work hardening and to contribute to the large compressive deformability with plastic strains up to 15%.

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Binary Cu–Zr alloys have been known since the early 1980s and much research has been done on them [1–8]. They exhibit two interesting and unique characteristics: (i) even binary Cu–Zr melts solidify into bulk metallic glasses (BMGs) [4,5,8]; and (ii) the equiatomic CuZr intermetallic compound can undergo a martensitic transformation from a cubic primitive B2 to a monoclinic B19' phase similarly to the well-known transformation that occurs in the NiTi system [9–11]. As the transformation from austenite (B2) to martensite (B19') is reversible, a shape memory effect (SME) arises in the CuZr alloy [10,12]. The addition of third elements such as Ti or Al significantly improves the glass-forming ability [13–15] and allows casting of larger specimens of fully glassy material. Alloys on the Cu-rich side, e.g. Cu₆₄Zr₃₆ [3–5], show poor plasticity, whereas Cu₅₀Zr₅₀ or compositions in its vicinity show an appreciable plastic strain [4,8]. Once the alloy composition has been optimized the plastic strain can be increased by the introduction of second-phase particles [16]. Structural heterogeneities such as nano- and micrometer-sized crystals have been found to be beneficial for the enhancement of macroscopic deformability [17–19].

Thus, the combination of both the structural heterogeneities and the ability to undergo a diffusionless and hence very fast shear transformation promises interesting mechanical properties in Cu–Zr-based alloys. It is important to note here that the B2 CuZr phase can only be obtained from the glassy state by proper adjustment of the cooling rate. Annealing treatments of glassy alloys lead to the formation of metastable phases or the precipitation of the equilibrium phases, depending on the composition [20].

To the best of the authors' knowledge only the formation of martensite in Cu–Zr-based partially glassy alloys upon quenching has been reported so far [19,21–24]. This is due to the stresses in the material that occur during the solidification process that usually trigger martensite formation. In this paper we report for the first time the B2 to B19' transformation after plastically deforming partially crystalline Cu–Zr–Ti and Cu–Zr–Al bulk specimens. Furthermore, we investigate the effect of the addition of Ti and Al to the Cu–Zr system on the martensite transformation characteristics.

Cu₅₀Zr_{50–x}Ti_x (0 ≤ x ≤ 10) and (Cu_{0.5}Zr_{0.5})_{100–x}Al_x (x = 5, 6, 8) pre-alloys were synthesized by arc-melting the proper amounts of the constituting elements (purity of 99.99% or higher). Ti and Zr have similar electron configurations and hence similar chemical behavior;

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thus Zr was substituted by Ti. In case of the Cu–Zr–Al alloys the Cu:Zr ratio was kept constant and the chemically different Al was added. The pre-alloy ingots were remelted at least four times in order to ensure chemical homogeneity. Approximately 5–8 g pieces of the ingots were used to produce 40 mm long rods with a square cross section of $2.1 \times 2.1 \text{ mm}^2$ via injection casting into a copper mold. Furthermore, rods of $\text{Cu}_{50}\text{Zr}_{50}$, $\text{Cu}_{50}\text{Zr}_{42.5}\text{Ti}_{7.5}$ and $\text{Cu}_{47.5}\text{Zr}_{47.5}\text{Al}_5$, 2 mm in diameter and 50 mm long were obtained. Low-temperature differential scanning calorimetry (DSC) was carried out in a Diamond DSC (Pyris) with a heating rate of 20 K min^{-1} . High-temperature DSC measurements were performed in a Netzsch DSC 404 C at a heating rate of 40 K min^{-1} with ribbons that had been rapidly quenched by single-roller melt spinning in an argon atmosphere. The bulk samples were rolled in a hand-roller to a reduction in thickness of between 3% and 16%. The different undeformed and deformed partially crystalline bulk specimens were characterized with high-energy X-rays ($\lambda = 0.010436 \text{ nm}$) at the BW5 beam line of the HASYLAB in Hamburg. The diffraction angle in the present work is given as the scattering vector q : $q = \frac{4\pi \sin(\theta)}{\lambda}$. For the compression tests an Instron 5869 was used with a constant strain rate of $1 \times 10^{-4} \text{ s}^{-1}$. A laser-extensometer (Fiedler) monitored the strain directly at the sample.

First of all the impact of the addition of a third alloying element, i.e. Al or Ti, to the binary Cu–Zr system is considered. Our attention is focused here on the influence of Al and Ti on the stability of the high-temperature B2 CuZr phase. According to the binary Cu–Zr phase diagram [25] the B2 phase is only stable at temperatures above 988 K and should decompose eutectoidly into the equilibrium phases $\text{Cu}_{10}\text{Zr}_7$ and CuZr_2 at lower temperatures.

The inset in Figure 1 shows the DSC traces of (a) $\text{Cu}_{50}\text{Zr}_{50}$ and (b) $\text{Cu}_{50}\text{Zr}_{42.5}\text{Ti}_{7.5}$ ribbons. The shift of the B2 transformation towards higher temperatures in the ternary alloy (b) is clearly visible. The final endothermic events correspond to the melting of the alloys. It should be noted here that the $\text{Cu}_{50}\text{Zr}_{50}$ alloy melts congruently, whereas in the present case a splitting of the melting peak is observed. Most likely the composition is somewhat richer in Cu and, as a consequence, a minor volume fraction of $\text{Cu}_{10}\text{Zr}_7$ (which has a lower melting point than CuZr) is present besides B2 CuZr. The addi-

Figure 1. Variation of the B2 transformation temperature of the $\text{Cu}_{50}\text{Zr}_{50-x}\text{Ti}_x$ ($0 \leq x \leq 10$) and $(\text{Cu}_{0.5}\text{Zr}_{0.5})_{100-x}\text{Al}_x$ ($5 \leq x \leq 8$) alloys. The inset shows the high-temperature DSC traces of (a) $\text{Cu}_{50}\text{Zr}_{50}$ and (b) $\text{Cu}_{50}\text{Zr}_{42.5}\text{Ti}_{7.5}$ at a heating of 40 K min^{-1} . The B2 transformation temperature is indicated by the arrows.

tion of Ti apparently leads to the depression of the liquidus temperature. Figure 1 summarizes the variation of the B2 transformation temperature in the $(\text{Cu}_{0.5}\text{Zr}_{0.5})_{100-x}\text{Al}_x$ and $\text{Cu}_{50}\text{Zr}_{50-x}\text{Ti}_x$ alloys. The addition of Al does not alter the B2 transformation temperature significantly. It remains at a value of about 988 K, which coincides with the values stated for the $\text{Cu}_{50}\text{Zr}_{50}$ alloy [25]. For the $\text{Cu}_{50}\text{Zr}_{50}$ alloy in the present work the B2 transformation temperature was determined to be around 1000 K. The deviation with respect to the literature data is believed to result from the already mentioned minor volume fraction of $\text{Cu}_{10}\text{Zr}_7$ that is present in the investigated alloy. Obviously, the substitution of Zr by Ti increases the B2 transformation temperature and thus destabilizes the B2 CuZr phase. This in turn has a pronounced influence on the phase formation in the Cu–Zr–(Al,Ti) alloys upon cooling. In Figure 2 diffraction patterns of (a) $\text{Cu}_{50}\text{Zr}_{50}$ and (b) $\text{Cu}_{50}\text{Zr}_{42.5}\text{Ti}_{7.5}$ as-cast rods with a diameter of 2 mm are shown. The strongest reflections of the binary alloy can be indexed by martensitic CuZr (B19') and the high-temperature B2 phase. In fact, the B2 phase has a cubic primitive structure ($Pm\bar{3}m$) and not a body-centered cubic (bcc) phase as frequently stated [9,26,27], as all reflections including those with $h+k+l=2n+1$ are visible. Those should be extinct in the case of a bcc unit cell. Two kinds of martensites have been found for the CuZr intermetallic compound both corresponding to the B19' structure [10,26,28]: $P2_1/m$ is the “normal” martensite and Cm is a superstructure of the former [26]. Both are present in the investigated specimens. Obviously the cooling rate is not sufficient to prevent partial crystallization of the melt. At first the B2 CuZr phase precipitates and to some extent undergoes a martensitic transformation. The remaining melt solidifies into a glass.

In $\text{Cu}_{50}\text{Zr}_{42.5}\text{Ti}_{7.5}$ no reflections of the martensite can be detected. Instead the Bragg peaks come from B2 CuZr and $\text{Cu}_{10}\text{Zr}_7$. The disappearance of martensite is believed to result from the lowering of the martensite start temperature, M_s , as reported by Koval et al. [12]. The addition of 7.5 at.% Ti shifts M_s to a value of around 270 K [12]. As the B2 phase has a relatively narrow stability region due to the addition of Ti, the cooling rate is not sufficient to suppress a partial decomposition of the B2 phase that solidifies first upon cooling. It has been suggested [20] that the formation of

Figure 2. Diffraction patterns of (a) $\text{Cu}_{50}\text{Zr}_{50}$ and (b) $\text{Cu}_{50}\text{Zr}_{42.5}\text{Ti}_{7.5}$ as-cast rods with a diameter of 2 mm. Note that the strongest reflection of (b) at around $2\theta = 18^\circ$ has been cut off.

the CuZr_2 phase is more sluggish as Zr is diffusing more slowly than Cu [29] and, as a result, the Cu-rich equilibrium phase $\text{Cu}_{10}\text{Zr}_7$ precipitates first.

Next to the martensite formation upon quenching, which is well known and has been reported for many different systems [9,21,23,24,30–32], the martensitic transformation can be also induced by deformation. Figure 3 displays high-energy X-ray diffraction patterns of as-cast and deformed $\text{Cu}_{47.5}\text{Zr}_{47.5}\text{Al}_5$ and $\text{Cu}_{50}\text{Zr}_{45}\text{Ti}_5$ bulk specimens. In the as-cast state the reflections result solely from the B2 phase in both alloys (Fig. 3a and c). The $\text{Cu}_{47.5}\text{Zr}_{47.5}\text{Al}_5$ alloy was rolled to a thickness reduction of approximately 3%. The diffraction pattern of the rolled sample shows additional reflections besides the B2 Bragg peaks, which unambiguously belong to the B19' phase (Fig. 3b). The same observations hold true for $\text{Cu}_{50}\text{Zr}_{45}\text{Ti}_5$ (Fig. 3d). However, the B19' reflections are much stronger compared to those found for $\text{Cu}_{47.5}\text{Zr}_{47.5}\text{Al}_5$. This is believed to result from the more severe plastic deformation (approximately 16%). Obviously the volume fraction of transformed B2 phase strongly depends on the plastic strain imposed by rolling. In other words, the higher the degree of deformation, the more martensite forms. The scattering intensity of the crystalline phase is much stronger than that of the glassy phase and therefore the presence of the glassy phase is not clearly deducible from Figure 3. We therefore performed DSC measurements and calculated the crystalline volume fraction of the rods by comparing the enthalpy of crystallization $\Delta_{\text{cryst}}H$ of the bulk specimens with $\Delta_{\text{cryst}}H$ of a fully glassy ribbon. This yields a crystalline volume fraction of 70 ± 5 vol.% for $\text{Cu}_{47.5}\text{Zr}_{47.5}\text{Al}_5$ and of 90 ± 5 vol.% for $\text{Cu}_{50}\text{Zr}_{45}\text{Ti}_5$.

Louzguine-Luzgin et al. [33] reported a deformation-induced transformation in a $\text{Ti}_{60}\text{Fe}_{20}\text{Co}_{20}$ alloy. They claimed that the martensitic transformation contributes only to a minor part to the overall ductility of the alloy. In the present case the transformation from the B2 to the B19' phase is substantial, as can be judged from Figure 3. Thus, we believe that it plays a key role in the enhancement of the macroscopic deformability similar to that reported for $\text{Ti}_{50}\text{Cu}_{50}$ [30]. The second important factor is the structural heterogeneity resulting from the composite nature of the Cu–Zr–Ti and Cu–Zr–Al alloys [17,18].

Figure 3. $\text{Cu}_{47.5}\text{Zr}_{47.5}\text{Al}_5$ bar-shaped sample (a) before and (b) after deformation. The sample thickness has been reduced by approximately 3%. $\text{Cu}_{50}\text{Zr}_{45}\text{Ti}_5$ bar-shaped sample (c) before and (d) after deformation. The thickness reduction amounts to approximately 16%. The strongest reflections at $q = 26 \text{ nm}^{-1}$ have been cut off for (a) and (c).

In order to analyze the role of the martensitic transformation during deformation we cast a $\text{Cu}_{47.5}\text{Zr}_{47.5}\text{Al}_5$ 2 mm diameter rod with a composite microstructure and performed compression tests. As can be seen from the inset in Figure 4, B2 crystals are embedded in a featureless glassy matrix [32]. Prior to the test the top and bottom side of the compression test specimens with an aspect ratio of 2 had been ground and polished and the volume fraction of crystals had been determined optically. Figure 4 shows the results of the compression tests. Sample (a) was taken from the bottom of the rod and contained 30 vol.% of B2 crystals; sample (b) came from the top part of the rod and exhibited a crystalline volume fraction of 50 vol.%. The data are summarized in Table 1. A higher volume fraction of crystals leads to a lower yield strength (σ_{ys}). Yet the work hardening is significantly more pronounced. This is an effect also observed in other systems that can undergo a martensitic transformation [17,23,24,30,34]. It is strongly reminiscent of the transformation-induced plasticity (TRIP) effect and is not unusual in shape memory alloys (SMAs) [35]. Interestingly, the Young's modulus of the $\text{Cu}_{47.5}\text{Zr}_{47.5}\text{Al}_5$ BMG composites (Table 1) is lower than that of Cu–Zr-based glasses [17]. The Young's modulus of the B2 CuZr phase seems to be quite low and to result in this low value for the composites. As there are no reports about the mechanical properties of this phase, data from the NiTi system are used as estimates. For the B2 NiTi phase an elastic modulus of 57 GPa has been reported [36]. This value is in good accord with the Young's moduli determined in the present work for the composites (52–56 GPa). Most likely the B2 CuZr phase has an equally low Young's modulus as the B2 NiTi phase, indicating the low resistance of this intermetallic compound to elastic deformation. Apparently, the deformation behavior of the composites is governed by the crystalline phase.

One last aspect regarding the size of the B2 crystals that transform will be addressed here. From the diffraction pat-

Figure 4. True stress–strain curves of $\text{Cu}_{47.5}\text{Zr}_{47.5}\text{Al}_5$ under compressive loading from (a) the bottom part of a rod and (b) the top part of the rod. With increase in the crystalline volume fraction (b) the work-hardening effect is much more pronounced. The inset shows the cross-section of the specimen taken from the top of the 2 mm diameter rod. The darker phase consists of B2 crystals and the glassy phase appears brighter.

Table 1. Elastic/plastic properties of $\text{Cu}_{47.5}\text{Zr}_{47.5}\text{Al}_5$ BMG composites determined under uniaxial compressive loading.

Sample	Crystalline volume fraction (%)	E (GPa)	Yield strength $\sigma_{y0.2}$ (MPa)	Maximum stress σ_{max} (MPa)	Fracture strain ϵ_f (%)
(a)	30 ± 5	56 ± 3	1246 ± 20	1627 ± 10	14.9 ± 0.3
(b)	50 ± 5	52 ± 2	550 ± 20	1467 ± 10	11.3 ± 0.3

terns and the optical investigations it is clear that the crystalline precipitates have sizes of the order of several tens of micrometers. However, one can speculate that the martensitic transformation also occurs for much smaller grains and maybe even atomic clusters, i.e. on a nanometer scale. It has been shown [37,38] that the short-range order plays a crucial role in the local deformation of simulated Cu–Zr-based metallic glasses. In particular, the formation of shear transformation zones (STZs) is strongly influenced by the presence and packing of Cu-centered icosahedral clusters [37]. These icosahedral clusters furthermore control plastic flow [38]. To explain the large plasticity found in Cu–Zr-based glasses [8,17] one can additionally conjecture the presence of atomic clusters with an austenitic (B2) short-range order, which can transform to a martensitic (B19') short-range order or undergo a twinning-like mechanism upon deformation. In this way the imposed plastic strain could be accommodated without the generation of shear bands. Such “M-glasses” are expected to exhibit enhanced plastic deformability [19,39].

The results of this study can be summarized as follows. Addition of Ti destabilizes the B2 CuZr phase at lower temperatures. This promotes the precipitation of the equilibrium phases $\text{Cu}_{10}\text{Zr}_7$ and CuZr_2 in as-cast rods. In contrast, Al as a third element does not significantly influence the B2 transformation temperature. Partially crystalline $\text{Cu}_{50}\text{Zr}_{45}\text{Ti}_5$ and $(\text{Cu}_{0.5}\text{Zr}_{0.5})_{100-x}\text{Al}_x$ BMG alloys show a martensitic phase transformation from a cubic B2 structure to a monoclinic B19' structure induced by plastic deformation. The higher the degree of deformation, the higher the transformed volume fraction. This behavior leads to pronounced work hardening, which becomes obvious in compression tests. Such a strong work hardening is typical of SMAs. When the martensitic transformation is being exploited in the investigated alloy systems one should rather consider the binary Cu–Zr alloys close to the equiatomic composition and their ternary derivatives containing Al. Assuming the observed martensitic transformation also proceeds on the scale of the short-range order provides a possible explanation for the large macroscopic plasticity found in Cu–Zr-based glassy alloys in the vicinity of the equiatomic CuZr intermetallic compound. Thus, this group of martensitic BMGs and its composites may open a new field of applications, where so far only classical SMAs have been employed.

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- [1] K.H.J. Buschow, J. Appl. Phys. 52 (1981) 3319.
- [2] Z. Altounian, G.H. Tu, J.O. Strom-Olsen, J. Appl. Phys. 53 (1982) 4755.
- [3] A. Inoue, T. Masumoto, N. Yano, J. Mater. Sci. 19 (1984) 3786.
- [4] A. Inoue, W. Zhang, Mater. Trans. 45 (2004) 584.

- [5] D. Xu, B. Lohwongwatana, G. Duan, W.L. Johnson, C. Garland, Acta Mater. 52 (2004) 2621.
- [6] A. Inoue, W. Zhang, T. Tsurui, A.R. Yavari, A.L. Greer, Phil. Mag. Lett. 85 (2005) 221.
- [7] W.H. Wang, J.J. Lewandowski, A.L. Greer, J. Mater. Res. 20 (2005) 2307.
- [8] Z.W. Zhu, H.F. Zhang, W.S. Sun, B.Z. Ding, Z.Q. Hu, Scripta Mater. 54 (2006) 1145.
- [9] E.M. Carvalho, I.R. Harris, J. Mater. Sci. 15 (1980) 1224.
- [10] Y.N. Koval, G.S. Firstov, A.V. Kotko, Scripta Metall. Mater. 27 (1992) 1611.
- [11] G.S. Firstov, J. Van Humbeeck, Y.N. Koval, J. Phys. IV 11 (2001) 481.
- [12] Y.N. Koval, G.S. Firstov, L. Delaey, J. Van Humbeeck, Scripta Metall. Mater. 31 (1994) 799.
- [13] A. Inoue, W. Zhang, T. Zhang, K. Kurosaka, Acta Mater. 49 (2001) 2645.
- [14] A. Inoue, W. Zhang, T. Zhang, K. Kurosaka, Mater. Trans. 42 (2001) 1149.
- [15] L.Q. Xing, P. Ochin, J. Bigot, J. Non-Cryst. Solids 205 (1996) 637.
- [16] J. Eckert, J. Das, S. Pauly, C. Duhamel, J. Mater. Res. 22 (2007) 285.
- [17] J. Das, M.B. Tang, K.B. Kim, R. Theissmann, F. Baier, W.H. Wang, J. Eckert, Phys. Rev. Lett. 94 (2005) 205501.
- [18] K.B. Kim, J. Das, F. Baier, M.B. Tang, W.H. Wang, Appl. Phys. Lett. 88 (2006) 051911.
- [19] J. Das, K.B. Kim, W. Xu, B.C. Wei, Z.F. Zhang, W.H. Wang, S. Yi, J. Eckert, Mater. Trans. 47 (2006) 2606.
- [20] S. Pauly, J. Das, J. Eckert, Intermetallics (2008) submitted for publication.
- [21] J. Das, S. Pauly, C. Duhamel, B.C. Wei, J. Eckert, J. Mater. Res. 22 (2007) 326.
- [22] W.J. Ma, Y.R. Wang, B.C. Wei, Y.F. Sun, Trans. Nonferr. Metals Soc. China 17 (2007) 929.
- [23] S. Pauly, J. Das, C. Duhamel, J. Eckert, Metall. Mater. Trans. A 39A (2008) 1868.
- [24] F. Qiu, P. Shen, Z.G. Jiang, T. Liu, Q.C. Jiang, Appl. Phys. Lett. 92 (2008) 151912.
- [25] K.J. Zeng, M. Härmäläinen, H.L. Lukas, J. Phase Equilib. 15 (1994) 577.
- [26] D. Schryvers, G.S. Firstov, J.W. Seo, J. Van Humbeeck, Y.N. Koval, Scripta Mater. 36 (1997) 1119.
- [27] Z.Y. Liu, M. Aindow, J.A. Hriljac, I.P. Jones, I.R. Harris, Metastable, mechanically alloyed and nanocrystalline materials, ISMANAM 2000, 360-3, 2001, p. 223.
- [28] A.V. Zhalko-Titarenko, M.L. Yevlashina, V.N. Antonov, B.Y. Yavorskii, Y.N. Koval, G.S. Firstov, Phys. Status Solidi B 184 (1994) 121.
- [29] O. Taguchi, Y. Iijima, K. Hirano, J. Alloys Comp. 215 (1994) 329.
- [30] K.B. Kim, K.A. Song, X.F. Zhang, S. Yi, Appl. Phys. Lett. 92 (2008) 241915.
- [31] Y.F. Sun, C.H. Shek, F.S. Li, S.K. Guan, Mater. Sci. Eng. A 479 (2008) 31.
- [32] S. Pauly, J. Das, C. Duhamel, J. Eckert, Adv. Eng. Mater. 9 (2007) 487.
- [33] D.V. Louzguine-Luzgin, L.V. Louzguina-Luzgina, V.I. Polkin, A. Inoue, Scripta Mater. 57 (2007) 445.
- [34] Q.S. Zhang, W. Zhang, G.Q. Xie, K.S. Nakayama, H. Kimura, A. Inoue, J. Alloys Comp. 431 (2007) 236.
- [35] G.S. Firstov, J. Van Humbeeck, Y.N. Koval, Scripta Mater. 50 (2004) 243.
- [36] C.P. Frick, T.W. Lang, K. Spark, K. Gall, Acta Mater. 54 (2006) 2223.
- [37] C.E. Lekka, A. Ibenskas, A.R. Yavari, G.A. Evangelakis, Appl. Phys. Lett. 91 (2007) 214103.
- [38] Y.Q. Cheng, A.J. Cao, H.W. Sheng, E. Ma, Acta Mater. 56 (2008) 5263.
- [39] J. Das, S. Pauly, M. Boström, K. Durst, J. Eckert, ISMANAM 2007, accepted for publication.