In this work, the authors report on the formation of toughened TiB₂-based ceramic composites using Ti₆₀.₈Al₂₅Nb₁₀ⅉ₃Mo₁B₀.₂ (at%) metallic glass powder as sintering aid fabricated by spark plasma sintering. The TiB₂ ceramic powder blended with 5 wt% metallic glass sintering aid exhibits an interesting sigmoidal shrinkage behavior which is argued to be attributable to the viscous flow of metallic glass powder. The fabricated TiB₂-based ceramic composites have high relative density of 99.1% and high fracture toughness of 9.9 Mpa·m⁰.⁵/². The fracture toughness values obtained in this study are related to the mechanical response of the TiAl phase, which transformed from the metallic glass sintering aid. The results obtained herein provide a novel pathway for the fabrication of high-performance ceramic composites.

1. Introduction

The titanium diboride (TiB₂) ceramic possesses a series of notable physical, mechanical, and chemical properties, such as high melting point, high hardness, high strength-to-weight ratio, and good corrosion and wear resistance. These excellent properties render it an outstanding engineering material for various applications, including high-temperature structural components, wear-resistant parts, cutting tools, and lightweight ballistic protection armor materials.[1–3] In such applications, both relative density and fracture toughness represent two parameters that are frequently used to evaluate the effectiveness and service life of TiB₂ components.

It is well established that the densification of monolithic TiB₂ requires high sintering temperatures, sometimes as high as 2200 °C, and long sintering times due to its strong ionic and covalent bonding, low self-diffusion coefficient and the presence of an oxygen rich surface layer (primarily TiO₂ and B₂O₃).[4,5] These characteristics of TiB₂ make it challenging to achieve near full density, even when using pressure-assisted sintering methods such as hot pressing and hot isostatic pressing.[6] This has limited the use of monolithic TiB₂ in engineering applications.[7,8] Compared with conventional sintering methods, spark plasma sintering (SPS) is a comparatively novel sintering technique allows the compaction of ceramics to be sintered with higher density at lower temperature with decreased sintering times.[9] During the SPS sintering process, spark discharge, joule heating, and plastic deformation effects all contribute to densification behavior.[10] In particular, the presence of a spark discharge has been argued to effectively disrupt any surface oxide films that may be present on the ceramic particles.[6] As a result, grain growth can be prevented due to the accelerated densification kinetics. Hence, the SPS provides notable benefits as an effective method to fabricate TiB₂ ceramics as well as its composites.
To fabricate bulk TiB₂-based ceramic composites with high relative density and fracture toughness, a sintering aid is generally required. Inspection of the published literature shows that Cr₂B₃ ceramics, MoSi₂ compounds, metallic Ti, Fe–Ni–Ti–Al and Fe and Ni have all been utilized as sintering aids to improve relative density and mechanical response because these sintering aids have lower melting point, better wettability, or better plastic deformation ability. From the results provided in these and other studies, an important question emerges. That is, are there other sintering aids, not previously considered, that can possibly lead to enhanced density and fracture toughness in TiB₂ ceramics? To that effect, we propose that in fact a metallic glass is an ideal sintering aid for TiB₂ and the basis for this suggestion can be rationalized by the following points:

1) Metallic glass powder exhibits viscous flow behavior prior to crystallization, as compared with its crystalline counterparts. This will greatly accelerate the densification process during sintering, thereby resulting in a crystallized bulk alloy with a high relative density.

2) Sintered and crystallized bulk alloys obtained from a metallic glass sintering aid present high room-temperature strength and ductility. This is typically attributed to grain refinement (nanocrystalline or ultrafine grains) induced by controlling nucleation and growth, an equiaxed microstructure, controlling crystallization kinetics and accelerating densification relative to bulk alloys that contain nanocrystalline or prealloyed powders as sintering aid;

3) Sintered and crystallized bulk alloys obtained from a metallic glass sintering aid reportedly exhibit excellent high-temperature mechanical properties. Moreover, controlled additions of Ti and Al into the sintering aid can prevent formation of undesirable brittle secondary borides such as MB, M₂B, and M₂₃B₆.

In view of the above argument, in this work, Ti₆₀.₈Al₂₅Nb₁₀V₃Mo₁B₀.₂ (at%) metallic glass powder was first introduced as a sintering aid to toughen TiB₂-based ceramic composites fabricated by spark plasma sintering. Ti₆₀.₈Al₂₅Nb₁₀V₃Mo₁B₀.₂ high-temperature alloy composition was selected due to its excellent high-temperature mechanical properties. The fabricated TiB₂-based ceramic composites have high relative density and fracture toughness, superior to those of TiB₂-based ceramic composites containing metallic compound and ceramic sintering aids.

2. Experimental Section
2.1. Sample Preparation

The as-received titanium (∼300 mesh, 99.7%), aluminum (∼300 mesh, >99.7%), niobium (∼325 mesh, 99.95%), vanadium (∼300 mesh, 99.9%), molybdenum (∼300 mesh, 99.95%), boron (∼325 mesh, 99.9%), and titanium diboride (3–5 μm, 99.9%) powders were obtained from Zhongnuo New Materials Co., Ltd., Beijing, China. These elemental powders were initially mixed for 24 h with a stoichiometry of Ti₆₀.₈Al₂₅Nb₁₀V₃Mo₁B₀.₂ (at%) in a blending machine. Mechanical alloying of the homogeneously mixed elemental powders was performed at a rotation rate of 248 rpm under a purified argon gas atmosphere (99.999%, 0.5 MPa) in a high-energy planetary ball mill (QM-2SP20-CL, apparatus factory of Nanjing University). Stainless steel vials and balls were utilized as the milling media with a ball-to-powder mass ratio of 7:1. The milling process was interrupted regularly every 10 h for cooling the milled powders to room temperature. Meanwhile, approximately 3 g of the as-milled powders were taken out from the mill containers every 10 h for various characterizations until achieving maximum content of glassy phase in the as-milled powder after 70 h milling. Subsequently, composite powders were prepared by mixing of 95 wt% as-received TiB₂ ceramic powder and 5 wt% 70 h-milled Ti₆₀.₈Al₂₅Nb₁₀V₃Mo₁B₀.₂ alloy powder, and were blended homogeneously for 12 h at 100 rpm in the ball-mill. To avoid introducing oxygen impurity, all manipulation was performed under an argon atmosphere via vacuum glove-box.

Bulk Ti-based alloy and TiB₂-based ceramic composites were fabricated, respectively, from the 70 h-milled Ti₆₀.₈Al₂₅Nb₁₀V₃Mo₁B₀.₂ alloy powder and the aforementioned mixed composite powders by SPS (Dr. Sinter825, Sumitomo CoalMining Co. Ltd., Japan) in a cylindrical graphite die with an inside diameter of 20 mm as well as a protective layer of tantalum foil with a thickness of 30 μm. The sintering temperatures were selected as 1200, 1300, 1400, and 1500 °C with a heating rate of 100 °C/min. The rationale used for the selection of these sintering temperatures is to obtain TiAl phase transformed from the as-milled powder after 70 h milling, which was not released until completed soaking for 5 min at the target sintering temperature under an argon atmosphere, and the specimens were cooled down to room temperature. In addition, instantaneous values of sintering parameters, such as temperature, punch displacement, and time, were recorded every 2 s by the attached software. The densification rate of the mixed composite powders was determined by calculating the derivative of punch displacement versus time.

2.2. Phase Composition and Microstructure

The evolution of structure as a function of milling time of the as-milled Ti₆₀.₈Al₂₅Nb₁₀V₃Mo₁B₀.₂ powders was confirmed by X-ray diffraction (XRD, D8 Advance, Bruker Co., Saarbrücken, Germany) with Cu Kα radiation. The thermal stability was measured by differential scanning calorimetry.
Fig. 1. XRD patterns (a) and DSC curves (b) of the milled Ti60.8Al25Nb10V3Mo1B0.2 alloy powders after different milling times.

(DSC, Netzsch STA 409 C) under a high-purity argon atmosphere at a heating rate of 20°C min⁻¹. Transmission electron microscope (TEM; FEI, Eindhoven, The Netherlands) was used for microstructural observation of the 70 h-milled Ti60.8Al25Nb10V3Mo1B0.2 alloy powder. The phase composition and microstructure of the fabricated bulk Ti-based alloy and TiB₂-based ceramic composites were evaluated by XRD and high-resolution scanning electron microscopy (SEM, Nova Nano 430, FEI, Hillsboro, OR, USA) couple with energy dispersive spectroscopy (EDS).

2.3. Mechanical Properties

In order to study the mechanical behavior of the fabricated bulk Ti-based alloy under compression, cylindrical specimens of 3 mm in diameter and 6 mm in length were tested in a universal testing machine (MTS testing system) under quasi static loading at a strain rate of 5 x 10⁻⁴ s⁻¹, and a small strain gauge was used to calibrate and measure the strain during loading. Based on Archimedes principle, the density of the TiB₂-based ceramic composite was measured using water. Its theoretical density was calculated to be 4.526 g cm⁻³ according to the rule of mixtures. The hardness (HV10) was evaluated using a Vickers hardness tester (4305VA, Wilson Wolpert Co. Ltd., Shanghai, China) with a load of 10 kgf for 20 s. The fracture toughness (KIC) was calculated based on the radial crack length produced by the Vickers (HV10) indentation according to Anstis’ formula [28]

\[ K_{IC} = \frac{16}{3} \left( \frac{E}{H} \right)^{1/2} \frac{P}{C^{3/2}} \]  

where E is the elastic modulus (GPa); H is the Vickers hardness (GPa); P is the indentation load (kgf); and C is the total crack length (μm) from each corner of the indent to the tip of the corresponding crack.

3. Results and Discussion

Figure 1 shows XRD patterns and DSC curves of the milled Ti60.8Al25Nb10V3Mo1B0.2 powders after different milling times. The starting mixture of elemental powders reveals diffraction patterns of Ti, Al, Nb, V, and Mo. After 10 h milling, the major peaks associated with the starting elemental powders decrease significantly. With increased milling time, both peak broadening and a gradual decrease in intensity of peaks can be observed due to reduction in grain size, increase in lattice distortion and decrease in crystallinity. The obvious appearance of diffuse diffraction implies formation of a glassy phase. After 70 h milling, the Ti60.8Al25Nb10V3Mo1B0.2 alloy powder has completely transformed into glassy phase. It has an onset crystallization temperature (Tc) of 560°C; crystallization event ends at 685°C, called Tc. Its heat release of crystallization, or activation energy for the viscous flow obtained by integrating the area under the peak in DSC trace (Figure 1b), is 25.2 J g⁻¹ at a heating rates of 20°C min⁻¹. The obtained heat release of crystallization here is comparable to other mechanically alloyed Ti-based metallic glass powders [29,30].

Further TEM analysis was employed to confirm the exactly existence of glassy phase in the 70-milled Ti60.8Al25Nb10V3Mo1B0.2 alloy powder (Figure 2). The corresponding selected area diffraction pattern in Figure 2a clearly shows a diffuse halo and bright diffraction rings. High-resolution TEM images confirm that this is indeed a glassy phase (Figure 2b). Meanwhile, there still exist some remained Ti₃Al nanocrystalline phases (Figure 2b) surrounded by the glassy matrix, also as seen by the dark regions in Figure 2a. Considering that viscous flow behavior prior to crystallization...
for metallic glass powder can greatly accelerate the densification process during sintering, thereby resulting in a high relative density for a crystallized bulk alloy.[19,20] These results suggest that the Ti_{60.8}Al_{25}Nb_{10}V_{3}Mo_{1}B_{0.2} metallic glass powder would enhance the sintering mechanism by facilitating viscous flow and thereby leading to a high sintering density sintering aid.

Figure 3 shows XRD pattern and SEM image of the sintered and crystallized Ti_{60.8}Al_{25}Nb_{10}V_{3}Mo_{1}B_{0.2} bulk alloy by heating to 1200 °C at 100 °C min^{-1} and holding for 5 min.

Figure 4 shows the densification curve corresponding to the TiB_{2}-based ceramic composite powder containing 5 wt% sintering aid of the 70 h-milled Ti_{60.8}Al_{25}Nb_{10}V_{3}Mo_{1}B_{0.2} metallic glass powder using heating rate of 100 °C min^{-1}. The results in this figure suggest that there are three shrinkage stages, or three instantaneous maximum values of shrinkage rate corresponding to the temperature ranges: 585-695, 695-730, and >730 °C, respectively. It is interesting to note that compared with a single instantaneous maximum value of shrinkage rate below 730 °C in the case of sintering TiB_{2}-based ceramic composite powder containing 5 wt% (Fe-Ni-Ti-Al) crystalline sintering aid,[15] the material containing 5 wt% metallic glass sintering aid exhibits two instantaneous maximum values of shrinkage rate. The first shrinkage stage occurs within the temperature range of 585-695 °C, approximately corresponding to the onset (560 °C) and completion (685 °C) crystallization temperatures of the 70 h-milled Ti_{60.8}Al_{25}Nb_{10}V_{3}Mo_{1}B_{0.2} metallic glass powder in Figure 1b, and has a sigmoidal curve and a significant shrinkage displacement above 1 mm. This particular sigmoidal shrinkage behavior is ascribed to the viscous flow of the metallic glass sintering aid, consistent with results reported for the sintering of monolithic Ti_{40.6}Zr_{9.4}Cu_{37.5}Ni_{9.4}Al_{3.1} metallic glass powder during SPS.[20] Also, the fact that the first instantaneous maximum value of shrinkage rate is similar to that of monolithic Ti_{40.6}Zr_{9.4}Cu_{37.5}Ni_{9.4}Al_{3.1} metallic glass powder,[20] provides support to the suggestion that this behavior is related to the viscous flow of pure metallic glass powder. Moreover, this shrinkage behavior will accelerate densification relative to the crystalline counterpart powder,[20] thereby leading to a higher density. The second
shrinkage stage, or instantaneous maximum value of shrinkage rate, is between 695 and 730 °C, occurs at a temperature that is slightly higher than that corresponding to the completion crystallization temperature (685 °C) of the metallic glass powder (Figure 1b), and is attributed to presence of a substantial amount of quasi-amorphous layers at the grain boundaries[19] in crystallized nanocrystalline powder transformed from the metallic glass sintering aid. When the sintering temperature exceeds 730 °C, the third shrinkage stage is attributed to the crystallized Ti_{60.8}Al_{25}Nb_{10}V_{3}Mo_{1.2} sintering aid with grown ultrafine grains. The different peaks, or instantaneous maximum value of shrinkage rates, may be related to formation of Ti(O, C, N) solid solution and Al_{2}O_{3} phases and the decomposition of the TiAl phase at 1 200 and 1 300 °C in the crystallized Ti_{60.8}Al_{25}Nb_{10}V_{3}Mo_{1.2} sintering aid, as discussed about phase identification in Figure 6 and 7.

The changes in relative density of the TiB_{2}-based ceramic composites depend strongly on sintering temperature, as shown in Figure 5. A rapid increase in the relative density is observed in the range of 1 200–1 300 °C. When the sintering temperature reaches 1 300 °C, the relative density remains relatively unchanged and approaches 99.0% of theoretical; further increases in the sintering temperature do not appear to improve this value. It is apparent from the results summarized in Table 1 that this value of relative density is significantly higher than that of TiB_{2}-based ceramic composites with the same amount of sintering aid of CrB_{2} ceramic,[11] MoSi_{2} compound,[12,13] and Fe–Ni–Ti–Al.[15]

Figure 6 shows XRD patterns for the TiB_{2}-based ceramic composites containing 5 wt% metallic glass sintering aid fabricated using different sintering temperatures. At a sintering temperature of 1 200 °C, the fabricated TiB_{2}-based ceramic composite consists of two primary constituent phases: TiB_{2} and M_{2}B (M = Nb, V, and Mo). Because of the small volume fraction of the Ti_{3}Al it fails to appear in the XRD spectra. When the sintering temperature increases to 1 300 °C, the main constituent phases include: TiB_{2}, TiAl, and Ti(O, C, N) phases, accompanied with the disappearance of formed M_{2}B phase. When the temperature is increased to 1 400 °C, the content of TiAl decreases slightly and the content of TiO(C, N) increases largely, accompanied with the appearance of small amount of the Al_{2}O_{3} phase. With further increases in temperature to 1 500 °C, the TiAl phase cannot be detected. According to the binary aluminum–titanium phase diagram,[27] the Ti_{3}Al phase decomposes into the TiAl and α-Ti phases, and the TiAl and β-Ti phases when the temperature exceeds 1 200 and 1 300 °C, respectively. Thus, the formation of the Ti(O, C, N) solid solution phase with interstitial atoms (O, C, N) at 1 300 °C is attributed to the reactions between the impurities (O, C, N) on the starting TiB_{2} powders, surface and the decomposed α-Ti.[3] The increased TiO(C, N) content

Table 1. Relative density, microhardness, elastic modulus, and fracture toughness of the TiB_{2}-based ceramic composites containing 5 wt% MG sintering aid.

| Material compositions | Processing details | Relative density [%] | Crack lengths [μm] | Elastic modulus [GPa] | Vickers hardness [GsPa] | Fracture toughness [MPa m^{1/2}] |
|-----------------------|--------------------|----------------------|-------------------|----------------------|------------------------|-----------------------------|
| Monolithic TiB{\textsuperscript{2}} & SPS, 1 500, 10, 40 & 96.1 & – & 397.9 & 17.4 & 5.4 |
| TiB_{2}–5 wt% CrB{\textsubscript{2}} & HP, 1 750, 60, 35 & 94.8 & – & – & 23.7 & 4.3 |
| TiB_{2}–5 wt% MoSi{\textsubscript{2}} & HP, 1 700, 60, 30 & 98.8 & – & – & 26.2 & 5.0 |
| TiB_{2}–2.5 wt% Ti & SPS, 1 650, 5, 50 & 99.1 & – & – & 26.8 & 5.9 |
| TiB_{2}–5 wt% Fe–Ni–Ti–Al & SPS, 1 500, 5, 30 & 97.6 & 115.8 & 471.2 & 19.7 & 6.2 |
| TiB_{2}–5 wt% TiAlNbVMoB & SPS, 1 300, 5, 30 & 99.8 ± 0.51 & 86 ± 1.63 & 431 ± 8.82 & 21 ± 1.43 & 9.0 |
| MG, present work & SPS, 1 400, 5, 30 & 99.1 ± 0.51 & 83 ± 1.08 & 454 ± 7.93 & 20 ± 0.95 & 9.9 |
| TiB_{2}–5 wt% TiAlNbVMoB & SPS, 1 400, 5, 30 & 99.8 ± 0.71 & 116 ± 0.96 & 478 ± 8.37 & 23 ± 0.91 & 6.0 |

Crack lengths from each corner of the indent to the tip of the corresponding crack in the TiB_{2}-based ceramic composites is measured by SEM and used to calculate fracture toughness. For comparison, related parameters of monolithic TiB_{2} ceramics as well as TiB_{2}-based ceramic composites with various sintering aids are also presented.

HP; Hot pressing; SPS, Spark plasma sintering; MG, Metallic glass.

Fig. 6. XRD patterns for the TiB_{2}-based ceramic composites containing 5 wt% metallic glass sintering aid fabricated at different sintering temperatures.
with the increased sintering temperature may result from more O impurity uptake from surrounding environment. Moreover, the disappearance of formed M₂B (M = Nb, V, and Mo) phase at 1300°C may be due to the decomposed α-Ti. The constituted elements M ( = Nb, V, and Mo) may react with the decomposed α-Ti and thus stabilize and form β-Ti phase, having an positive influence on the mechanical properties of the TiB₂-based ceramic composites as presented in Figure 8 and Table 1. Likewise, the formation of a small quantity Al₂O₃ at 1400°C can be ascribed to the reaction between the decomposed Al with introduced minor oxygen existing in the powders during the sintering process. This suggestion is consistent with the fact that the Ti₃Al aid used in sintering the TiB₂ ceramic powder decomposed at 1350°C into Ti and Al and decomposed Al reacted with O to form Al₂O₃ particles. In addition, the disappearance of the TiAl phase at 1500°C may simply be attributable to its small volume fraction and partial decomposition.

Figure 7 shows back-scattered SEM images of the TiB₂-based ceramic composites containing 5 wt% metallic glass sintering aid fabricated at different sintering temperatures. The gray black region corresponds to the TiB₂ ceramic phase; a small part of gray white phase is confirmed to be Ti(O, C, N) solid solution, and the content of Ti(O, C, N) increased significantly with increasing sintering temperature. Moreover, the bright white phase corresponds to the TiAl intermetallic compounds, and with increasing sintering temperature, its proportion decreases gradually. In addition, the Al₂O₃ phase is associated with dark spots (see arrows), and it is evenly distributed between the ceramic particles. The average chemical compositions of the Ti(O, C, N) solid solution, TiAl intermetallic compound, and Al₂O₃ phases determined by EDS in the TiB₂-based ceramic composites fabricated at 1500°C are Ti₃9.8O₃6.9C₃.3, Ti₅₀.₅Al₃₉.₀Fe₂.₉O₄.₂C₃.₇, and Al₂₃₆O₆₉.₇Ti₆.₇, respectively. This proves that the correctness of phase identification in Figure 6 and 7.

Figure 8 presents the influence of the sintering temperature on grain size, microhardness, elastic modulus, and fracture toughness of the TiB₂-based ceramic composites containing 5 wt% metallic glass sintering aid. Generally, the hardness of
bulk materials reveals a strong dependence on grain size and relative density. The grain size of the TiB₂ phase in the fabricated TiB₂-based ceramic composites revealed in Figure 8a is below 4.9 μm, suggesting an absence of grain growth as compared with the as-received TiB₂ powder. The hardness trend is similar to that of relative density and it remains relatively constant when sintering temperature exceeds 1300 °C. Therefore, it can be deduced that the hardness is primarily affected by the relative density. As seen in Figure 8b, the elastic modulus increases with the increased sintering temperature. This may be principally attributed to the elimination of pores. However, the fracture toughness increases first with the increased sintering temperature, has maximums at the sintering temperatures of 1300 and 1400 °C, and decreases with the further increased sintering temperature to 1500 °C. Its change is consistent with the content change in TiAl phase (Figure 6) in the fabricated TiB₂-based ceramic composites. This indicates that the increased fracture toughness may be due to the enhancement in bond strength between the TiB₂ phase and the crystallized TiAl phase from sintering aid. As expected, the TiB₂-based ceramic composites fabricated between 1400 and 1400 °C exhibit an excellent combination of microhardness and fracture toughness (Table 1). The measured fracture toughness is superior to those for Ti₃Al-based ceramic composites with the same content sintering aid of CaB₂ ceramic,[11] 4.3 MPa m¹/², MoSi₂ compound[12] 5.0 MPa m¹/², metallic Ti,[13] 6.2 MPa m¹/², and Fe-Ni-Ti-Al[14] 6.2 MPa m¹/². The fracture toughness can be attributed to the outstanding mechanical response of the constituent phase TiAl transformed from the metallic glass sintering aid.[26]

Through observing the indentation topography after the Vickers hardness measurement in Figure 9a, cracks propagate from the four corners of the crease, respectively. It can be observed in Figure 9b that fracture occurs primarily via a transgranular mechanism and that some crack deflection is also evident. The transgranular fracture mode is a result of the high density as well as stronger bonding between the TiB₂ and sintering aid.
[14] Z. H. Zhang, X. B. Shen, F. C. Wang, S. K. Lee, Q. B. Fan, M. S. Cao, Scr. Mater. 2012, 66, 167.
[15] C. Yang, H. Guo, D. G. Mo, S. G. Qu, Y. Q. Li, W. W. Zhang, L. C. Zhang, Materials 2014, 7, 7105.
[16] D. A. Hoke, M. A. Meyers, J. Am. Ceram. Soc. 1995, 78, 275.
[17] G. Wang, J. Shen, J. F. Sun, Y. J. Huang, J. Zou, Z. P. Lu, Z. H. Stachurski, B. D. Zhou, J. NonCryst. Solids 2005, 351, 209.
[18] S. Scudino, B. Bartusch, J. Eckert, J. Phys. Conf. Ser. 2009, 144, 012097.
[19] V. V. Dabhade, T. R. R. Mohan, P. Ramakrishnan, Mater. Res. Bull. 2007, 42, 1262.
[20] L. H. Liu, C. Yang, Y. G. Yao, F. Wang, W. W. Zhang, Y. Long, Y. Y. Li, Intermetallics 2015, 66, 1.
[21] L. H. Liu, C. Yang, L. M. Kang, Y. Long, Z. Y. Xiao, P. J. Li, L. C. Zhang, Mater. Sci. Eng. A 2016, 650, 171.
[22] L. H. Liu, C. Yang, F. Wang, S. G. Qu, X. Q. Li, W. W. Zhang, Y. Y. Li, L. C. Zhang, Mater. Des. 2015, 79, 1.
[23] C. Yang, L. H. Liu, Q. R. Cheng, D. D. You, Y. Y. Li, Mater. Sci. Eng. A 2013, 580, 397.
[24] C. Yang, L. H. Liu, Y. G. Yao, Y. H. Li, Y. Y. Li, J. Alloys Compd. 2014, 586, 542.
[25] M. G. Barandika, J. M. Sánchez, T. Rojo, R. Cortés, F. Castro, Scr. Mater. 1998, 39, 1395.
[26] R. U. Vaidya, Z. Jin, C. Cady, G. T. Gray III, D. P. Butt, Scr. Mater. 1999, 41, 569.
[27] J. C. Schuster, M. Palm, J. Phase Equilib. Diff. 2006, 27, 255.
[28] G. R. Anstis, P. Chantikul, B. R. Lawn, J. Am. Ceram. Soc. 1981, 64, 533.
[29] Y. Y. Li, C. Yang, W. P. Chen, X. Q. Li, J. Mater. Res. 2008, 23, 745.
[30] L. C. Zhang, Z. Q. Shen, J. Xu, Mater. Sci. Eng. A 2005, 394, 204.
[31] J. M. Sánchez, I. Azcona, F. Castro, J. Mater. Sci. 2000, 35, 9.
[32] A. Mukhopadhyay, T. Venkateswaran, B. Basu, Scr. Mater. 2015, 69, 159.