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Microstructure and mechanical properties of hot pressed WC matrix composites toughened by commercial Al₂O₃ and α -Al₂O₃ particulates

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Tungsten carbide (WC) matrix composites toughened by commercial alumina (Al₂O₃) composed of boehmite (AlOOH), χ -Al₂O₃ and amorphous Al₂O₃ (WA1) and α -Al₂O₃ (WA2), respectively, were prepared by high energy ball milling and the following hot pressing, respectively. The microstructure and mechanical properties of WA1 and WA2 were characterized and compared. The results showed that the content of ditungsten carbide (W₂C) in the sintered WA1 composites was much smaller than that of WA2 because the phase transformation of the commercial Al₂O₃ in WA1 suppressed the decarburization of WC; The Vickers hardness and fracture toughness of WA1 were higher than that of WA2 due to the smaller WC grain size and Al₂O₃ particulate size resulted from the phase transformation of the commercial Al₂O₃ in WA1. The lower density and worse particulate/matrix interfacial combining conditions of WA1 compared with that of WA2, resulting from the phase transformation of the transformation t

Key words: WC matrix composites, Hot pressing sintering, Mechanical properties, Microstructure-final.

Introduction

Tungsten carbide (WC) has achieved significant scientific and technological attention for use in cutting tools due to its excellent properties, such as high melting point, superior hardness, good electrical conductivity and intrinsic resistance to oxidation and corrosion at high temperature. However, the relatively low value of fracture toughness, poor ductility and shock resistance of monolithic WC limits its applicability under severe conditions such as for high speed cutting tools [1]. Metallic cobalt (Co) [2, 3], Nickel (Ni) [4] and ferrite (Fe) [5] are usually used as binder material to improve several mechanical properties of WC. Embedding the WC grains, the naturally ductile metals serve to offset the characteristic brittle behavior of WC ceramic, thus raise its toughness and durability. However, metallic binders result in reduced hardness and corrosion/ oxidation resistance, and enhance grain growth due to the rapid diffusion in the liquid phase [6, 7]. Especially, Co in the cutting tools made of WC-Co composites will be oxidized due to its low melting point and the high temperature induced by the high cutting speed. In addition, Co has other shortcomings, namely, poor corrosion resistance, high cost and environmental toxicity, which limit the use of WC-Co materials in industrial applications [8]. In the past few decades,

significant researches efforts have been focused on developing WC with none metallic binders [9-14].

 Al_2O_3 is another refractory oxide with low price, superior hardness, high resistance to oxidation and corrosion at high temperature. The melting point of Al_2O_3 is lower than that of Zirconium oxide (ZrO₂) and Magnesium oxide (MgO), according to the reported results, the sintering temperature of WC-Al₂O₃ is lower than that of WC-ZrO₂ and WC-MgO composites [10, 12, 15]. In addition, Al_2O_3 is inexpensive compared to Co, ZrO₂ and MgO, indicating that the WC-Al₂O₃ composites will be much more economical for industrial application.

It is well known that α -Al₂O₃ could be obtained by calcining the beohmite powders [16]. Alumina derived from the decomposition of beohmite powders undergoes a number of phase transformations before the stable α -Al₂O₃ is obtained [17]. As a result, the cost of α -Al₂O₃ powders was higher than that of beohmite powders. In general, WC-Al₂O₃ composites are produced by the compaction and sintering of WC and α -Al₂O₃ [18], however, there is interest and economical advantages in WC-Al₂O₃ composites produced directly by sintering of the WC and beohmite powders. Meanwhile the influence of phase transformation during hot pressing on the microstructure and mechanical properties of the WC-Al₂O₃ composites could be considered. Accordingly, commercial Al₂O₃ composed of AlOOH, χ -Al₂O₃ and amorphous Al₂O₃ were used as the starting materials to consolidate WC-Al₂O₃ composites (WA1). Comparing its mechanical properties with the WC-Al₂O₃ composites

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consolidated with WC and α -Al₂O₃ powders as the starting materials to investigate the effects of phase transformation on the mechanical properties of the WC-Al₂O₃ composites.

Some previous studies are carried out to investigate the synthesis of WC-Al₂O₃ powders [19-22] and the sintering of WC-Al₂O₃ composites with α -Al₂O₃ as the starting material [18, 23], the influence of sintering temperature and holding time on the densification, phase transformation, microstructure and properties of hot pressing WC-40 vol.%Al₂O₃ composites (WA1) is investigated by Qu and Zhu et al. [15], however, little information is available on the effects of phase trans-formation of Al₂O₃ starting materials on the micro-structure and mechanical properties of WC-Al₂O₃ composites.

Recently, we have attempted to consolidate the commercial Al₂O₃ composed of AlOOH, χ -Al₂O₃ and amorphous Al₂O₃ toughened WC-Al₂O₃ composites (these samples will be referred to as WA1) and α -Al₂O₃ toughened WC-Al₂O₃ composites (these samples will be referred to as WA2) using hot pressing method. The aims of present work are to investigate the microstructure and mechanical properties of WC matrix composites toughened by commercial Al₂O₃ and α -Al₂O₃ particulates and discuss the effects of phase transformation of commercial Al₂O₃ on the toughening of WC-Al₂O₃ composites.

Experimental Procedure

Commercial powders of WC (74 µm, 99% purity), Al₂O₃ (composed of AlOOH, χ -Al₂O₃ and amorphous Al₂O₃; 150 µm) and α -Al₂O₃ (74 µm) provided by Sinopharm Chemical Reagent Co. Ltd. were used in this study. The WC powders were mixed with 40 vol.% commercial Al₂O₃ powders composed of AlOOH, χ -Al₂O₃ and amorphous Al₂O₃ (WA1) and 40 vol.% α -Al₂O₃ (WA2), respectively. Then the mixed powders were ball milled using a QM-1SP4 planetary ball milling machine under argon gas atmosphere for 50h, respectively. The ball to powder weight radio was 10 : 1 and the rotation speed of the mill was 350 revolutions per minute (rpm). Both the vial and milling balls (10 mm in diameter) were made of cemented carbide materials.

According to the experimental results reported in [15], the optimum hot pressing process was 1540 °C for 90 min. The as milled powders were hot pressed using a vacuum hot pressing furnace (ZT-40-20YB, Shanghai Chen Hua Electric Furnace Co. Ltd., China) at 1540 °C under a pressure of 39.6 MPa (applied already from the start) in a vacuum (about 1.3×10^{-1} Pa) atmosphere for 90 min. The heating rate of the hot pressing was 10 °C/ min. Carbonic paper with a thickness of 0.2 mm was used to prevent adhesion between powders and the die. The thermoelectric couple and infrared thermometer were selected to measure the temperature, located in the

surface of the mold.

The hot pressed WA1 and WA2 samples were ground and polished by standard ceramographic methods. Phase identification of the sintered and polished WA1 and WA2 samples were investigated by X-ray diffraction (XRD) using a D/max-2550PC (Rigaku Co., Japan) Xray diffractometer with a Cu K_{α} radiation ($\lambda =$ 0.15418 nm) at 400 KV and 200 mA. The microstructure of the polished samples was characterized by S-4800 (Hitachi Co., Japan) field emission scanning electron microscope (FE-SEM) and Nanoscope (Veeco Co., USA) scanning probe microscope (SPM). Chemical composition was examined by energy dispersive spectroscope (EDS). The fracture surfaces of the sintered samples were observed by S-3400N (Hitachi Co., Japan) scanning electron microscope (SEM). Then samples were etched in a Murakami's reagent consisting of Fe₃[K(CN)₆] (10 g), KOH (10 g) and distilled water (100 ml) for 5 min to expose the grain boundary of WC. The polished and etched samples were characterized by JSM-5600LV (JEOL Co., Japan) scanning electron microscope (SEM). The WC grain size and Al₂O₃ particle size were estimated, respectively, by the linear intercept method. For each sample, at least three images were taken of the microstructure; in each image a minimum of five line segments were assessed. The density of the sintered samples was measured using water immersion method in accordance with Archimedes' principle. The hardness of sintered specimens (the average of 10 indentations) was determined using a HVS-50Z Vickers indenter with a load of 30 kg and a dwell time of 10s. The crack shapes of the specimen determined by repeated surface polishing were confirmed to be Palmqvist crack pattern and in accordance with the Shetty's model, so the fracture toughness (K_C) calculations were based on the crack length measure-ment of the radial crack pattern produced by Vickers indentations according to the formula estimated by Shetty et al [24]. The fracture toughness values were derived from the average of 10 measurements. The flexural strength of sintered samples was measured by the three-point flexural test method at room temperature according to the ASTM B312 standard. The samples were cut and ground into $3 \times 5 \times 15 \text{ mm}^3$ specimens. The three-point flexure tests were carried out on a WDW-100 (Changchun kexin experimental equipment Co. Ltd., China) test machine with a span of 10 mm and cross-head speed of 0.01 mm/min. The fracture strength of sintered samples (the average of three tests) was calculated using

$$\sigma = 3PL/2t^2w \tag{1}$$

where σ (MPa) is the flexural fracture strength, P (N) is the force required to rupture, L (L = 10 mm) is the length of the span of fixture, w (w = 5 mm) is the width of the specimen, and t (t = 3 mm) is the thickness of the specimen.

Results and discussion

Phase analysis of WA1 and WA2 composites

The XRD patterns of the starting and the as-milled powders (ball milled for 50 h) of WA2 composites were shown in Fig. 1. From Fig. 1a, it could be seen that the starting powders of WA2 was composed of WC and α -Al₂O₃. However, when the starting materials were ball milled for 50 h, only the peaks belonging to WC were detected in the XRD pattern (Fig. 1b), the α -Al₂O₃ peaks were not observed due to the trace amount of α -Al₂O₃ and the refinement of the ball milling process, the intensity of α -Al₂O₃ peaks were very low compared with that of WC. In addition, it could also be demonstrated that there were no composition changes took place and WC and α -Al₂O₃ did not react with each other during the ball milling process. Comparing of Fig. 1a and 1b, it indicated that the XRD peaks of WC broadened and the intensity decreased in response to the decreasing crystalline size, as well as increasing lattice strain.

Fig. 2a shows the XRD patterns of commercial Al_2O_3 powders used in this study. It indicated that the commercial Al_2O_3 powders were consisted of a great quantity of amorphous Al_2O_3 and trace amounts of



Fig. 1. XRD patterns of WC-40 vol.% α -Al₂O₃ powders ball milled for (a) 0 hour and (b) 50 hours.



Fig. 2. XRD patterns of (a) commercial Al_2O_3 powders used in this study, (b) WC powders premixed with 40 vol.% commercial Al_2O_3 powders ball milled for 0 h, (c) partially magnified XRD patterns of (b), and (d) WC powders premixed with 40 vol.% commercial Al_2O_3 powders ball milled for 50 h.

AlOOH and χ -Al₂O₃. XRD patterns of WC powders premixed with 40 vol.% commercial Al₂O₃ powders ball milled for 0h were shown in Fig. 2b. The peaks of AlOOH and χ -Al₂O₃ were not observed due to the small content and their low intensity compared with that of WC. However, the partially magnified XRD pattern of Fig. 2b, as shown in Fig. 2c, confirms the existence of AlOOH and χ -Al₂O₃. Fig. 2d shows the XRD patterns of WC-40 vol.%Al₂O₃ powders ballmilled for 50 hours. Only peaks belonging to WC were observed, indicating that no composition changes took place, and phase transformation of AlOOH, χ -Al₂O₃ and amorphous Al₂O₃ did not occur. During the ball milling process, the AlOOH and χ -Al₂O₃ were refined and the particles were too small to be detected. With the increase of milling time, the XRD peaks of WC broadened and the intensity decreased as observed in the milling process of WC- α -Al₂O₃ powders in response to decreasing crystalline size and increasing lattice strain.

The XRD patterns of the hot pressed WA1 and WA2 composites were shown in Fig. 3. The peaks of WC and α -Al₂O₃ could be observed in both WA1 and WA2 composites, indicating that both the sintered WA1 and WA2 composites were composed of WC and α -Al₂O₃, and the commercial Al_2O_3 composed of AlOOH, χ -Al₂O₃ and amorphous Al₂O₃ in WA1 starting powders transformed to α -Al₂O₃ during the hot pressing process due to the calcining effect. According to the results reported in [15], with the increasing hot pressing temperature, the starting powders of WA1 lost its physically bonded water at 769 °C; the amorphous Al₂O₃, AlOOH and χ -Al₂O₃ contained in WA1 starting powders were transformed to γ -Al₂O₃ when the sintering temperature was 961 °C; when the starting powders were hot pressed at 1100 °C for 90 min, all the transitional Al₂O₃ and amorphous Al₂O₃ were transformed to α -Al₂O₃ completely.

It could be seen from Fig. 3 that even when the starting powders were sintered at 1540 °C, no other new phases could be identified, indicating that there was no reaction and no interface phase between WC and Al_2O_3 and they had good adopting characteristics. W_2C peaks were also detected in Fig. 3a and 3b due to



Fig. 3. XRD patterns of (a) WA1 and (b) WA2 composites sintered at 1540 °C for 90 min.

the oxidation and decarburization of WC during the sintering process. The decarburization of WC to C, W_2C and W is thought to proceed in the following three stages [25]:

$$2WC \rightarrow W_2C + C \tag{2}$$

$$W_2C + 1/2O_2 \rightarrow W_2(C,O) \tag{3}$$

$$W_2(C,O) \to 2W + CO \tag{4}$$

The ball milled WC and Al_2O_3 powders have been expected to contain surface oxides. When the powders are hot pressed, the surface oxide is reduced. In this case, carbon is consumed in the reduction of surface oxide. Consequently, there is a lack of carbon to maintain the WC and WC is decarburized. W₂C phase forms due to the decarburization of WC during sintering [26]. Ma and Zhu et al [14] also observed the phenomenon of decarburization of WC phase and the formation of W₂C phase during the hot pressing sintering process.

In addition, it could be obviously observed that the intensity of the W_2C peaks of WA1 composites was much lower than that of WA2. The most important difference in the hot pressing processes of WA1 and WA2 was the phase transformation of commercial Al_2O_3 in WA1. So it can be reasonably deduced that the phase transformation of WA1 could suppress the decarburization of WC effectively. This may be related to the energy conditions in the hot pressing process.

The transformation of γ -Al₂O₃ to α -Al₂O₃ occurs by nucleation and growth process [27], its isothermal transformation kinetics could be analyzed by the Johnson-Mehl-Avrami theory [28-31], which is usually applied to a variety of interface-reaction-controlled or diffusion-controlled transformations. The theory develops an expression for the transformation rate of the form

$$\alpha = 1 - exp(-gIU^{m}t^{n}) \tag{5}$$

where α is the fraction transformed, g is a geometric shape factor, I is the nucleation rate, U is the growth rate, m is a constant dependent on the growth mechanism, t is the time, and n is a constant related to the physical characteristics of the system.

Plotting $\ln\ln(1-\alpha)^{-1}$ vs $\ln(t)$ allows the determination of n and the reaction rate constant k, because the slop and the intercept of the resulting straight line is n and $n\ln(k)$, respectively. The reaction rate constant k can be related to the activation energy (E_a) of the transformation through the expression

$$k = Aexp(-E_d/(RT)) \tag{6}$$

where A is a constant, E_a is the activation energy of the transformation, R is the gas constant and T is the absolute temperature. Bagwell and Messing have reported

that the activation energy (E_a) of the commercial boehmite and alkoxide boehmite is 567 KJ/mol and 547 KJ/mol, respectively [32].

So it is possible that the activation energy (E_a) absorbed before the phase transformation of commercial Al₂O₃ would consume part of the system energy, decreasing the remaining energy could be used for decarburization of WC, as a result, the decarburization of WC is suppressed and the content of W₂C correspondingly decreases. With the high hardness and low Young's modulus [33], W₂C is generally thought to be a brittle phase harmful to the properties of the sintered composites, the decreased content of W₂C is one of the reasons for the improved properties of the hot pressed WA1 composites.

Microstructure analysis of WA1 and WA2 composites

The FE-SEM images of the polished WA1 and WA2 samples showing the different dispersion states and morphologies of the toughening particulates were shown in Fig. 4. Two phases with gray and dark colors and the grey-dark concavities were observed in Fig. 4. The EDS data of polished WA1 samples were illustrated in Fig. 5. According to the EDS spot analysis (spectrum 2 and 1 in Fig. 5), it was indicated that the dispersed dark particulates were composed of Al and O elements, the surrounding grey matrix was rich in W and C elements. As a result, it can be confirmed that the dark particulates was Al₂O₃, the surrounding grey matrix was WC. While from the spectrum 3 of Fig. 5, it can be seen that the grey-dark concavity was rich in Al, O, W and C, indicating that the grey-dark concavity was composed of Al₂O₃ and WC. The grey-dark concavity may be



Fig. 4. FE-SEM images of the polished (a) WA1 and (b) WA2 composites sintered at 1540 °C for 90 min showing the different dispersion states and morphologies of the toughening particulates.



Fig. 5. FE-SEM images and EDS data of the polished WA1 sintered at 1540 $^{\rm o}{\rm C}$ for 90 min.

formed during the grinding and polishing process. Due to the low toughness of Al_2O_3 , the Al_2O_3 particulates were easily peeled off from the WC matrix, leaving a concavity on the polished surface.

From Figs. 4a and 4b, it could be observed that the Al_2O_3 particulates dispersed in the WC matrix homogeneously. However, the Al_2O_3 particulate size of WA1 was smaller than that of WA2, and the Al_2O_3 particulates in the WA1 composites existed with much more angularities (Fig. 4a), while the edges of the Al_2O_3 particulates in the WA2 composites were smooth



Fig. 6. SEM images of the polished and etched (a) WA1 and (b) WA2 composites sintered at 1540 °C for 90 min.

(Fig. 4b). The different sizes and morphologies of Al_2O_3 particulates would influence the mechanical properties of the hot pressed WC-Al_2O_3 composites.

The characteristic microstructures of the polished and etched samples are shown in Fig. 6. A number of SEM photos similar to that of Fig. 6 were used to evaluate the grain size of WC and a number of FE-SEM images similar to that of Fig. 4 were used to evaluate the particle size of Al₂O₃. The particle size of Al₂O₃ and the grain size of WC are summarized in Table 1. It can be seen from Table 1 that when the WA1and WA2 starting powders were sintered at 1540 °C for 90 min, the particle size of Al₂O₃ and the grain size of WC of WA1 composite was 1.80 µm and 2.79 µm, respectively. While the particle size of Al₂O₃ and the grain size of WC of WA2 composite which was larger than that of WA1composite was 1.94 µm and 3.92 µm, respectively.

It is reasonable that the phase transformation of the commercial Al_2O_3 during the hot pressing process of WA1 composites could decrease the particle size of Al_2O_3 . The transformation of γ to α -Al₂O₃ occurs by nucleation and growth process [27], and rearrangement of particles occurs during the transformation [17]. Because of the difference in specific volume of boehmite (0.332 cm³/g)

Table 1. The experimental density (ρ), relative density ($\rho_{\%}$), mechanical properties, WC grain size (d_w) and Al₂O₃ particle size (d_a) of WA1 and WA2 composites sintered at 1540 °C for 90 min.

Sample	Sintering Process	ρ	ρ‰	Vickers hardness (HV ₃₀)	Fracture toughness (K _C)	Flexural strength (σ)	Grain size of WC (d _w)	Particle size of Al ₂ O ₃ (d _a)
		g/cm ³	%TD	GPa	$MPa \cdot m^{1/2}$	MPa	μm	μm
WA1	1540 °C -90 min -39.6 MPa	10.74 ± 0.01	97.99	18.65 ± 0.57	10.43 ± 0.68	756.34 ± 32.53	2.79 ± 0.39	1.80 ± 0.33
WA2		10.78 ± 0.01	98.36	16.55 ± 0.4	8.52 ± 0.49	881.35 ± 46.06	3.92 ± 0.37	1.94 ± 0.28



Fig. 7. 3D SPM images showing the particulate/matrix interfacial microstructures of (a) WA1 and (b) WA2 composites sintered at 1540 °C for 90 min.

and α -Al₂O₃ (0.25 cm³/g), 24 vol.% porosity develops in the body after transformation [34]. A considerable amount of fine pores are redistributed throughout the compacts during sintering. Thanks to the press applied during the sintering process, the pores will be easily excluded.

According to the residual stress toughening mechanism, smaller size of secondary particles brings the improvement of the fracture toughness of WA1 composites [15]. So it can be deduced that there is a possibility that the phase transformation of the amorphous Al_2O_3 and transitional Al_2O_3 during sintering may be beneficial for the toughening of WA1 composites.

To investigate the effects of the morphology of Al_2O_3 particulates on the particulate/matrix interfacial combining conditions, the scanning probe microscope (SPM) analysis of WA1 and WA2 composites were carried out. Fig. 7 shows the particulate/matrix interfacial microstructure of the sintered WA1 and WA2 composites. It could be seen that the morphology of WA1 was rough and fluctuated, while the morphology of WA2 was smooth, indicating that the particulate/matrix interfacial combining conditions of WA2 composite was better than that of WA1 and the Al_2O_3 toughening particulates of WA2 had the continuous and compatible interfaces with the matrix.



Fig. 8. SEM images of the fracture surfaces of (a) WA1 and (b) WA2 composites sintered at 1540 $^{\circ}$ C for 90 min.

Due to the transformation of WA1 during the hot pressing process, the commercial Al₂O₃ particles should be rearranged to form α -Al₂O₃ and some vacancies and shrinkages can be generated leading to such morphology, however, the rearrangement of Al₂O₃ particles depend on the sintering parameters and the amount the α -Al₂O₃ nucleus [17]. Accordingly, the morphology would be different from that of WA2 composite without phase transformation during hot pressing process, resulting in the different particulate/matrix interfacial microstructure.

There is another possible reason for the unsatisfactory particulate/matrix interfacial combining conditions of WA1 composites. During the phase transformation, there was water vapor generated from WA1 composites due to the starting powders containing trace amount of boehmite (AlOOH). Most of the water vapor generated from the AlOOH conversion can escape from the composites, as there are continuous load and high temperature applied to the composites during hot pressing. Minor amount of the vapor become to be residual pores. As a result, the density of WA1 composite is lower than that of WA2 composite (as described in Table 1) and the pores deteriorate the particulate/matrix interfacial combining conditions of WA1 composite.

Since the particulate/matrix interfacial combining conditions and pores would affect the flexural strength obviously, this could be used to well explain why the flexural strength of WA1 composite was lower than that of WA2. Generally, the strength of a polycrystalline ceramic material is determined by many factors that include chemical composition, microstructure and surface conditions [35]. Beside the better combining conditions, the relative density of WA2 was higher than that of WA1, indicating the smaller amount of the pores in the WA2 sample, which would be beneficial for the high flexural strength. According to the toughening mechanics analyzed in [15], when particulate size was smaller than the critical one, the increase of it would increase the length of crack propagation path, increasing the flexural strength of the composites. That is, although the grain size of WA1 was smaller than that of WA2, both the two grain sizes did not exceed the critical one, as a result, the a little larger grain size of WA2 could be a favorable factor to the higher flexural strength of WA2.

SEM images of the fracture surfaces of WA1 and WA2 composites sintered at 1540 °C for 90 min were shown in Fig. 8. Comparing Fig. 8a with 8b, it could be illustrated that there were less pores in WA2, the grain shapes of WA2 were more regular than that of WA1, and the WC grain size of WA2 was larger than that of WA1, agreeing with the measured and analyzed results above. Meanwhile, it could be observed that the fracture surfaces of both WA1 and WA2 composites were characterized by a mixed mode of intergranular and transgranular fracture.

Mechanical properties of WA1 and WA2 composites

Table 1 summarized and compared the density and mechanical properties of the hot pressed WA1 and WA2 composites. As analyzed above, due to the residual pores generated during the hot pressing process, the density of WA1 composites was lower than that of WA2 composites; Both the WC grain size and Al₂O₃ particle size of WA1 composites were smaller than that of WA2 composites according to the calculated results of the SEM images of the polished and etched samples. The higher Vickers hardness and fracture toughness of WA1 composites were attributed to the smaller WC grain size and Al₂O₃ particle size deduced by the phase transformation of the commercial Al₂O₃. The lower flexural strength of WA1 composites than that of WA2 composites due to the worse particulate/matrix interfacial combining conditions and lower relative density of WA1 composites.

Conclusions

- (1) When the WA1 and WA2 starting powders were hot pressed at 1540 °C for 90 min, a relative density of 97.98% and an excellent Vickers hardness of 18.65 GPa, combining a fracture toughness of 10.43 MPa · m^{1/2} with an acceptable flexural strength of 756.34 MPa were achieved for the WA1 composites, while a relative density of 98.38% and a maximum hardness of 16.55 GPa, combining a fracture toughness of 8.52 MPa · m^{1/2} with an improved flexural strength of 881.35 MPa were obtained for the WA2 composites.
- (2) The phase transformation of the commercial Al₂O₃ in WA1 composites could suppress the decarburization of WC during the hot pressing process. The intensity

of W_2C in the final sintered WA1 composites was much smaller than that of the WA2 composites.

- (3) The Vickers hardness and the fracture toughness of WA1 composites were higher than that of WA2 due to the smaller WC grain size.
- (4) The relative density of WA1 composites was lower than that of WA2, in addition to the angular morphology of Al₂O₃ particulates, the particulate/matrix interfacial combining conditions of WA1 was worse compared with the continuous and compatible particulate/matrix interfacial combining conditions of WA2, inducing the lowered flexural strength of WA1 composites.

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