Microstructure and Dynamic Compression Properties of PM Al6061/B₄C Composite

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Received: 2 April 2015 / Revised: 23 June 2015 / Published online: 25 September 2015
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Abstract Aluminum 6061 matrix composite reinforced by 35 wt% B₄C particle was fabricated by power metallurgy method. Then, the as-deformed composite was tested by quasi-static (0.001 s^{-1}) and dynamic ($760\text{--}1150\text{ s}^{-1}$) compression experiments. The Johnson–Cook plasticity model was employed to model the flow behavior. The damage mechanism of composite was analyzed through the microstructure observations. The results showed that the B₄C particles exhibited uniform distribution and no deleterious reaction product Al₄C₃ was found in the composite. Al6061/B₄C composite showed high yield strength, moderate strain rate sensitivity and strain hardening under the dynamic loading, and a constitutive model under dynamic compression was established based on Johnson–Cook model, and accorded well with experimental results. The microstructure damage was dominated by particle fracture and interface debonding, and the dislocation was observed in the composite at a higher strain rate.

KEY WORDS: Power metallurgy; Dynamic compression; Johnson–Cook model; Al6061/B₄C composite

1 Introduction

Metal matrix composites (MMCs) reinforced with ceramic particles, short fibers or whisker possess outstanding physical and mechanical properties [1–3], such as wear resistance, low density, low coefficient of thermal expansion, high strength and stiffness [4, 5]. These properties lead to numerous applications of these composites in

aerospace, electronic, military and automotive, etc. [6–8]. Generally, the MMCs are fabricated by spark plasma sintering, spray forming, powder metallurgy, stir casting, squeeze casting [9–12]. Among above manufacturing methods, the powder metallurgy is an attractive processing method for producing MMCs as it is relatively simple and offers a wide selection of matrix, species and content of reinforcement and processing conditions. In addition, the reinforcements are homogeneously distributed in the matrix even for the composites with high reinforcement content.

The most common reinforcements are Al₂O₃, SiC, C_f, B₄C, MgO, etc. [13]. Owing to the low density (2.53 g cm^{-3}), high melting point (2450 °C) and good hardness ($35\text{--}45\text{ GPa}$), B₄C has become one of the important reinforcements in the last few decades. Moreover, it exhibits excellent capacity for neutron absorption in isotope ¹⁰B (with an effective cross section of 3840 barns) and is one of the few elements that do not emit γ -rays on being radiated. B₄C/Al composites can be used in nuclear industries of protective screen, especially for spent

Available online at <http://link.springer.com/journal/40195>

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fuel storage as structural frame and functional pieces [14, 15]. However, the composite material may suffer from the violent impact such as earthquake, tsunami and terror attack in use. Previous research [16] has shown that the mechanical behavior of composite is very complex under the high strain rate as compared with the quasi-static. So, the characterization of dynamic constitutive behavior of Al6061/B₄C composite is very important for its engineering applications.

Therefore, the aim of the present work is to investigate the microstructure and dynamic properties of Al6061/B₄C composites with B₄C fabricated by PM. The dynamic properties were tested at room temperature using a split Hopkinson pressure bar (SHPB) apparatus with a momentum trap. The material parameters of existing Johnson–Cook model were determined, and the predicted results were compared with the experimental results. The relationship between the microstructures and the deformation was also discussed.

2 Materials and Methods

A commercial power of Al6061 with average diameter of 13 μm was chosen as matrix alloy, supplied from Beijing Xingrongyuan technology, Co., Ltd, China, and the chemical composition is shown in Table 1. B₄C particles with an average diameter of 18.5 μm were used as reinforcements, supplied from Dalian Boentan technology, Co., Ltd, China, and the chemical composition is shown in Table 2. Figure 1a, c shows that the powders are free from agglomerates, and Al6061 particle shape is spherical and B₄C particle is irregular with sharp edges and corners. Particle size distribution curves for Al6061 and B₄C particles are given in Fig. 1b, d, and bimodal distribution and normal distribution are found for the Al6061 particles and the B₄C particles, respectively.

The Al6061/B₄C composites with 35 wt% B₄C were produced by power metallurgy method. B₄C and Al6061 powder mixtures were subjected to mechanical milling, followed by vacuum hot pressing (VHP), hot extrusion at 500 °C and hot rolling, in sequence. During the mechanical milling process, the powders are subjected to ball–powder–ball collision causing severe plastic deformation, cold welding and fracturing of the powders [17]. Detailed parameters of the mechanical milling experiment are

Table 1 Chemical composition of Al6061 alloy powder (wt%)

Si	Fe	Cu	Mn	Mg	Cr	Zn	Ti	Al
0.6	0.7	0.25	0.15	0.8	0.1	0.25	0.15	Bal.

Table 2 Chemical composition of B₄C powder (wt%)

B	C	Ca	Fe	Si	F	Cl
80.0	18.1	0.3	1.0	0.5	0.025	0.075

presented in Table 3. During vacuum hot pressing, the mixture powders were pressed under a pressure of 100 MPa in vacuum (~ 0.1 Pa) at 600 °C for 40 min, graphite die and punches were used, and both the heating rate and cooling rate were kept at 15 °C min^{−1}. After hot pressing, the Al6061/B₄C composite ingots were extruded at 500 °C with the ratio of 10:1 at a constant speed of 10 mm s^{−1} to form 300 mm $\times$ 60 mm $\times$ 15 mm samples. Finally, hot rolling was carried out at 500 °C to obtain the plate with thickness of 7 mm.

The density of the Al6061/B₄C composites was measured by Archimedes method, and the samples were weighed by an analytical balance (TG328A), with an accuracy of ± 0.1 mg. Microhardness measurements were taken using a Vickers microhardness tester (HUAYIN HVS-1000A), and the reported data are the average of three samples. Scanning electron microscopy (SEM, JSM 6700F) was used to study the microstructure of composites. Phase evaluation of the composites was performed by a powder X-ray diffractometer (XRD, D/MAX2400, CuK α). DNS200 compression machine was used to perform quasi-static compression tests with a strain rate of 0.001 s^{−1} at room temperature. A conventional SHPB equipment with 37-mm-diameter bars was used for dynamic compression tests at strain rates ranging from 760 to 1150 s^{−1}. The specimens that were machined in cylindrical shape with 7 mm in length and 12 mm in diameter by electro-discharge machining were used for both dynamic and quasi-static compression tests. Both surfaces of the cylindrical specimens were finely ground and made as flat and as parallel as possible. The lateral surfaces of the samples after deformation were investigated by SEM and transmission electron microscopy (TEM, JEM 2100F).

3 Results and Discussion

The relative density and Vickers hardness of Al6061/B₄C composites are presented in Table 4. The results show that the relative density of VHP specimens is only $\sim 85.4\%$, and it reached to $\sim 96.7\%$ after hot extrusion. After subsequent hot rolling, the density approaches to $\sim 100\%$. The increase in relative density is due to the voids reduction and better distribution of B₄C during extrusion and rolling.

The Vickers hardness of Al6061/B₄C composites was continuously increased in the fabricate process and reached

Fig. 1 SEM images and particle size distribution curves of Al6061 **a, b** and B₄C **c, d** powders

Table 3 Experimental parameters of ball milling

Milling media	Ball size (mm)	Weight ratio of ball to powder	Milling speed (r min ⁻¹)	Milling time (min)
ZrO ₂	10 and 6	15:1	1500	30

Table 4 Relative density and average microhardness values of Al6061/B₄C at different period of production process

Process	Relative density (%)	Average microhardness (HV _{0.5})
Vacuum hot pressing	85.44 ± 1.2	121.8
Hot extrusion	98.67 ± 0.4	148.4
Hot rolling	99.43 ± 0.4	157.3

to the maximum value of 157.3 HV after hot rolling. The obtained superior hardness values could be explained as a compilation of contributions: The major factors include the B₄C reinforcement phase with high hardness value, homogeneous dispersion of the B₄C reinforcement in the Al6061 matrix and a good adhesion between the B₄C particle and Al6061 particle, in which the misfit strain results in geometrically necessary dislocation generation in the vicinity of B₄C particle.

Figure 2a shows the microstructure of Al6061/B₄C composites fabricated by powder metallurgy. B₄C particles distribute uniformly in the Al6061 matrix. The rearrangement of B₄C particles takes place during extrusion and rolling because the plasticized Al6061/B₄C composite is subjected to the severe deformation caused by high compressive stress. There are no voids or porosities at the interface, as shown in Fig. 2b. This indicates proper bonding of B₄C particles with the Al6061 matrix. These

excellent features are mainly attributed to the followings: During high-energy ball mill, the B₄C particles are embedded into the Al6061 powers, which is beneficial to improve the distribution uniformity of the mixed radius of B₄C particles. Besides, small B₄C particles usually distribute among the larger ones, which is also in favor of the increased density of the composite after vacuum hot pressing. After the application of extrusion and rolling, the porosity is decreased, the particles agglomeration is decomposed, and better particle distribution is obtained.

The XRD patterns of Al6061, B₄C and Al6061/B₄C MMCs are depicted in Fig. 3, which shows the presence of Al and B₄C. The B₄C particles are thermodynamically stable at the fabricated temperature. Usually, the reaction of B₄C and Al6061 in power metallurgy sometimes produces undesirable intermetallic phase (Al₄C₃) which will have a bad effect on the strength of Al6061/B₄C composites [18]. Arslan *et al.* [19] reported that the B₄C–Al

Fig. 2 SEM images of B₄C/Al6061 composite **a** and the interface between B₄C and 6061Al **b**

Fig. 3 XRD patterns of Al6061 **a**, B₄C **b** and B₄C/Al6061 **c**

composites are composed of various combination of Al₃BC, AlB₂, AlB₁₂C₂ and Al₄C₃ phases. However, Al₃BC, AlB₂, AlB₁₂C₂ and Al₄C₃ diffraction peaks were not detected by the XRD, which may be ascribed to the less quantity of them.

Figure 4 displays the engineering stress–strain curve of the Al6061/B₄C composites upon quasi-static compression under the strain rate of 0.001 s^{−1}. It can be found that the yielding strength is about 250 MPa, and work hardening occurs during the compression. Comparison of the responses of Al6061/B₄C composites under quasi-static loading with the dynamic loading (760–1150 s^{−1}) is presented in Fig. 5. It can be found that the composites upon varied dynamic loadings exhibit plastic flow and the strain rate is not constant. Here, the maximum plastic-flow stress is simply taken as the yield stress [20]. The yield stresses under strain rates 760, 857, 956 and 1150 s^{−1} are 387, 399, 403 and 427 MPa, respectively, which are all higher than

Fig. 4 Engineering stress–strain curve of the B₄C/6061Al composites upon quasi-static compression under strain rate of 0.001 s^{−1}

Fig. 5 Engineering stress–strain curves of the B₄C/6061Al composites upon dynamic loading compression under different strain rates in the range of 760–1150 s^{−1}

that of the yield stress (~ 250 MPa) upon quasi-static loading. Meanwhile, it can also be observed from Fig. 5 that the strain hardening and the flow stress increase with increasing strain rate. Hence, the material may be considered as moderate strain-rate-sensitive material [21].

Relationship between the maximum flow stress and strain rate of Al6061/B₄C composites is presented in Fig. 6. The values of maximum flow stress are within the range of 387–427 MPa, which increases with increasing strain rate. Based on the experimental data, the linear relationship between the maximum flow stress ($\sigma_{\max}$) and the strain rate ($\dot{\varepsilon}$) is approximately expressed as:

$$\sigma_{\max} = 0.0995\dot{\varepsilon} + 311.37 \quad (1)$$

Johnson–Cook model is a phenomenological model which is widely used to reproduce the response of the materials subjected to impact and penetration problems [22, 23] and connects the quasi-static and dynamic mechanical properties. The Johnson–Cook model can be described as

$$\sigma = (A + B\varepsilon^n) \left(1 + C \ln \dot{\varepsilon}^* \right) \left[1 - \left(\frac{T - T_{\text{room}}}{T_{\text{melt}} - T_{\text{room}}} \right)^m \right] \quad (2)$$

where A is the quasi-static yield stress, B is the modulus of strain hardening, n is the strain hardening exponent, C is the strain rate hardening constant, $\varepsilon^* = \varepsilon/\varepsilon_0$ is the normalized equivalent plastic strain rate ($\varepsilon_0 = 0.001 \text{ s}^{-1}$), m is the thermal softening constant, T is the material temperature, T_{room} is the room temperature, and T_{melt} is the melting temperature. B and n are determined by curve-fitting method from the stress–strain curve up to 10% of the maximum deformation of the specimens. C is the strain rate sensitivity parameter and is obtained by curve-fitting method from the stress–strain curve at different strain rates. In the present work, the dynamic compression was carried

out at room temperature. Thus, the temperature effect can be ignored [24]. As a consequence, the Johnson–Cook material model can be described as

$$\sigma = (A + B\varepsilon^n) \left(1 + C \ln \dot{\varepsilon}^* \right) \quad (3)$$

By fitting the curves in Fig. 4, the parameters A , B and n are obtained as 250, 518.46 and 0.682 MPa, respectively.

By fitting the curves, the experimental and predicted engineering stress–strain curves of Al6061/B₄C composites at different strain rates are shown in Fig. 7 based on the Johnson–Cook model, the constitutive relationship for Al6061/B₄C composites is obtained as $\sigma = (250 + 518.46\varepsilon^{0.682n}) \left(1 + C \ln \dot{\varepsilon}^* \right)$, and the C values at strain rates of 760, 857, 956 and 1150 s^{-1} are 0.0020, 0.0025, 0.0028 and 0.0034, respectively. The average value of the C is 0.0027. As a consequence, the constitutive relationship for the Al6061/B₄C composites with 35 wt% B₄C is: $\sigma = (250 + 518.46\varepsilon^{0.682n}) \left(1 + 0.0027 \ln \dot{\varepsilon}^* \right)$. The dynamic compression experiment curves presented are similar to sine wave distribution, which is because of the hardening–softening transition, since the higher the flow stress, the more the deformation energy transferred to heat, and the higher the temperature of the specimen.

After the compression test, all specimens were compressed into thin disks, and neither obvious broken nor the adiabatic shear band was observed. The SEM images in Fig. 8 indicate that the lateral surfaces of samples are deformed. Some voids, particles fracture and particles assembling are observed at strain rates of 1150 s^{-1} , as shown in Fig. 8c. The magnified SEM images of Al6061/B₄C composites after compression under the strain rate of 1150 s^{-1} are presented in Fig. 9, which indicate that the microstructural damage is dominated by those defects at large deformation. Big particles fracture is primarily parallel to the loading direction, as shown in Fig. 9b, due to the stress transferring to B₄C particles across particle/particle contacts. Some big B₄C particles are broken under high deformation. This has also been stated by Ahmed *et al.* [25] that the particle size $< 20 \mu\text{m}$ is not susceptible to fracture, but the rupture takes place in the matrix. There are two main factors responsible for the results: (1) More defects existing around the big B₄C particles can provoke the premature fracture under the high deformation; (2) the interfacial adhesion between the large B₄C particle and the 6061Al matrix is weak. The presence of the voids is caused by the debonding of particle/matrix and the particles assembling, as schematically illustrated in Fig. 9.

It is known that dislocation interaction in a polycrystalline metal is stronger at a higher strain rate. For example, metals and crystalline alloys with fcc, bcc and hcp structures normally exhibit a higher strength at dynamic strain

Fig. 6 Relationship between the maximum flow stress and strain rate of Al6061/B₄C composites

Fig. 7 Experimental and predicted engineering stress–strain curves of Al6061/B₄C composites at different strain rates

Fig. 8 SEM images of Al6061/B₄C composites after compression under different strain rates of 0.001, 760 and 1150 s^{−1}

rates than that at a quasi-static strain rate [26]. Typical TEM micrograph of dynamic compressed specimen is shown in Fig. 10. The dislocation density is high in the composites. It is generally accepted that the change of strain rate sensitivity of the Al6061/B₄C composites is related to the two followed aspects: (1) The change of the mobile dislocation density is due to the mismatch of plastic strain and the large differences of coefficient of thermal expansion between the B₄C ($4.4 \times 10^{-6} \text{ K}^{-1}$) and 6061Al ($23.6 \times 10^{-6} \text{ K}^{-1}$) matrix; (2) the resistance to the dislocation motion imposed by the B₄C particles at high strain rate deformation. As it is schematically illustrated in Fig. 11, representing the presence of B₄C particles within

the matrix can also take a role in hampering dislocation movement in the matrix and led to the dislocation multiplication. By the above analysis, it can be concluded that dislocation interaction has a significant effect on improving the yield strength of Al6061/B₄C composites under the dynamic loading.

4 Conclusions

1. B₄C particles are homogeneously distributed in the PM Al6061/B₄C composites. The specimens are free of voids and cracks, and the bonding interface between

Fig. 9 The magnified SEM images of Al6061/B₄C composites after compression under the strain rate of 1150 s⁻¹: **a** particle/matrix interface debonding; **b** particles fracture; **c** particles assembling

Fig. 10 Typical TEM micrograph of dynamic compressed specimen

B₄C particles and Al6061 matrix is good. The Vickers hardness and the relative density of the composites are 157.3 HV_{0.5} and 99.4%, respectively, after hot rolling.

2. The yield strength and strain hardening under the dynamic loading were found to be larger than that at the quasi-static condition. The maximum flow stress increases with the strain rate increasing. The constitutive relationship for Al6061/B₄C composites can be described as $\sigma = (250 + 518.46e^{0.682n})(1 + 0.0027 \ln \dot{\epsilon}^*)$.
3. Big particles deboned easier than smaller one under the dynamic loading, and the microstructural damage is dominated by big particles fracture and interface debonding. The yield strength and strain hardening are improved due to the dislocation interaction in the Al6061/B₄C composites.

Fig. 11 Schematic illustration of microstructure variation under high strain rate

Acknowledgments This work was financially supported by the Key Science and Technology Program of Shanxi Province, China (No. 20130321024) and the Graduate Innovation Project of Shanxi Province, China (No. B2014005). The authors are grateful to Prof. Z.H. Wang of Applied Mechanics and Biomedical Engineering, Taiyuan University of Technology, for his helpful discussion on the preparation of the experiment.

References

- [1] K.K. Deng, J.Y. Shi, C.J. Wang, X.J. Wang, Y.W. Wu, K.B. Nie, K. Wu, *Compos. A* **43**, 1280 (2012)
- [2] Y.Z. Li, Q.Z. Wang, W.G. Wang, B.L. Xiao, Z.Y. Ma, *Mater. Sci. Eng. A* **620**, 445 (2015)
- [3] Y.P. Huang, W.J. Zhang, L. Liang, J. Xu, Z. Chen, *Chem. Eng. J.* **220**, 143 (2013)
- [4] B. McWilliams, T. Sano, J. Yu, A. Gordon, C. Yen, *Mater. Sci. Eng. A* **577**, 54 (2013)
- [5] P. Barick, D.C. Jana, N. Thiagarajan, *Ceram. Int.* **39**, 763 (2013)
- [6] H. Joel, *Mater. Des.* **30**, 323 (2009)
- [7] Y.C. Feng, L. Geng, P.Q. Zheng, Z.Z. Zheng, G.S. Wang, *Mater. Des.* **28**, 2023 (2008)
- [8] C.S. Ramesh, R. Keshavamurthy, B.H. Channabasappa, A. Abrar, *Mater. Sci. Eng. A* **502**, 99 (2009)
- [9] B. Sivaiah, M. Saravanan, D. Ajay, J. Mater. Sci. Technol. **28**, 969 (2012)
- [10] K. Kalaiselvan, N. Murugan, S. Parameswaran, *Mater. Des.* **32**, 4004 (2011)
- [11] K.K. Deng, C.J. Wang, J.Y. Shi, Y.W. Wu, K. Wu, *Mater. Chem. Phys.* **134**, 581 (2012)
- [12] T. Varol, A. Canakci, S. Ozsahin, *Acta Metall. Sin. (Engl. Lett.)* **28**, 182 (2015)
- [13] A. Canakci, T. Varol, *Powder Technol.* **268**, 72 (2014)
- [14] J.J. Park, S.M. Hong, M.K. Lee, C.K. Rhee, W.H. Rhee, *Nucl. Eng. Des.* **282**, 1 (2015)
- [15] P. Zhang, Y.L. Li, W.X. Wang, Z.P. Gao, B.D. Wang, *J. Nucl. Mater.* **437**, 350 (2013)
- [16] W.H. Liu, Z.T. He, Y.Q. Chen, S.W. Tang, *Trans. Nonferrous Met. Soc. China* **24**, 2179 (2014)
- [17] T. Varol, A. Canakci, *Powder Technol.* **246**, 462 (2013)
- [18] K. Kalaiselvan, I. Dinaharan, N. Murugan, *Mater. Des.* **55**, 176 (2014)
- [19] G. Arslan, F. Kara, S. Turan, *J. Eur. Ceram. Soc.* **23**, 1243 (2003)
- [20] J.W. Qiao, M.Y. Chu, L. Cheng, H.Y. Ye, H.J. Yang, S.G. Ma, Z.H. Wang, *Mater. Lett.* **119**, 92 (2014)
- [21] S. Rawat, S. Chandra, V.M. Chavan, S. Sharma, M. Warriar, S. Chaturvedi, *J. Appl. Phys.* **116**, 213507 (2014)
- [22] D.N. Zhang, Q.Q. Shangguan, C.J. Xie, F. Liu, *J. Alloys Compd.* **619**, 186 (2015)
- [23] N.K. Singh, E. Cadoni, M.K. Singha, N.K. Gupta, *Mater. Des.* **32**, 5091 (2011)
- [24] J. Wang, H.W. Zhang, A.M. Wang, H. Li, *Acta Metall. Sin.* **48**, 636 (2012). (in Chinese)
- [25] M.E. Ahmed, S. Mohamed, A. Mohamed Taha, P. Heinz, J. Mater. Process. Technol. **213**, 1669 (2013)
- [26] T. Mukai, T.G. Nieh, Y. Kawamura, A. Inoue, K. Higashi, *Scr. Mater.* **46**, 43 (2002)