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Improvement of strength of B/Al composite by thermal-mechanical cycling①

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Abstract: The mechanical properties of B/Al composite were measured at room temperature in the as-fabricated condition and after thermal-mechanical cycling(TMC). The effects of TMC on microstructure and tensile fracture behavior of B/Al composite were studied using transmission electron microscope(TEM) and scanning electron microscope(SEM). The fibers/matrix interfaces are degraded during TMC, the extent of which is enhanced with increasing the cycles, causing a measurable decrease of stage Ⅰ modulus of the B/Al composite. The TMC induces the dislocation generation in the aluminum matrix and the dislocation density increases with the cycles. The synergistic effect of the matrix strengthening and the interfacial degradation during TMC is found to play an important role in controlling the changes of tensile strengths and fracture behavior of the composite. The ultimate tensile strength of the composite increases with the cycles increasing. The interfaces in the B/Al composite change from the strongly-bonded states toward the appropriately-bonded ones with increasing the cycles. TMC will provide an approach of improving the strength of B/Al composites.

Key words: B/Al composite; thermal-mechanical cycling; microstructure; mechanical properties

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1 INTRODUCTION

B/Al composite exhibits high uniaxial specific strength, modulus and stiffness, attracting great attention in the aerospace applications[1-4]. Thermal-mechanical cycling (TMC) is one of the most severe environmental conditions for the composites used on spacecraft. Substantial degradation of the interfacial bonding such as sliding or debonding may result from the continuous variation of near-interface stresses due to thermal-mechanical cycling. It is necessary to study the mechanical behavior of B/Al composites subjected to thermal mechanical cycling. Wright[5] reported the effect of cyclic temperature variation on the B/Al composites. It was shown that the strength degradation of the composites depended on the matrix alloys and the upper cyclic temperature. This effect was attributed to the decreased fiber strength due to ratcheting of the matrix surface and a decrease in the effective interfacial strength. Grimes et al[6] found that the thermal degradation in tensile strength of unidirectional B/Al composites was due to a decrease in fiber strength resulted from the interface reaction. Marfia et al[7] presented that the thermal cycling degraded the strengths of the composite. The fracture mode of the thermally cycled specimens showed a transition toward the cumulative one, the characteristic of weakening interfaces caused by a reversible stress field was induced by differential thermal expansion.

Skinner et al[8] examined the effect of thermal cycling on toughness of the B/Al composites. Changes in toughness of the B/Al composites were correlated with the matrix softening, interface reaction products and fiber notch sensitivity. Izdinsky et al[9] investigated the effect of thermal cycling in the temperature range from room temperature up to 500 ℃ on mechanical properties of the B/Al composites. The experimental results showed that the thermal cycling resulted in significant decrease in ultimate tensile strength of the composites and the boron fibers, as well as the ultimate fracture strain. The transition from the cumulative to the noncumulative failure mode for the composites was revealed. The interfacial reaction was found as the reason for the change. Most of the work described above studied the changes in mechanical properties of the B/Al composites after thermal cycling between room temperature and elevated temperatures in the absence of external stresses. The mechanism for the above experimental results was mainly focused on the interfacial reaction and the oxidation of fibers at elevated temperatures. Little work has been done on the effect of thermal cycling with a given applied load, namely the thermal-me-

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chanical cycling. The present paper aimed to study the mechanical properties and microstructure of B/Al composites under TMC.

2 EXPERIMENTAL

The composite used in this study was an aluminum alloy 5A06 (AA5A06) matrix unidirectionally reinforced by boron fibers with a diameter of 140 μm, which was fabricated by the vacuum hot-isostatical press/diffusion bonding technique. A stack of 9 foil-fiber plies with an average of 56 fibers/cm for each ply was pressed to 1.4 mm thick B/Al sheet. The nominal fiber volume fraction was 50%. Fig. 1 shows a transverse cross section of the composite in the as-fabricated condition.

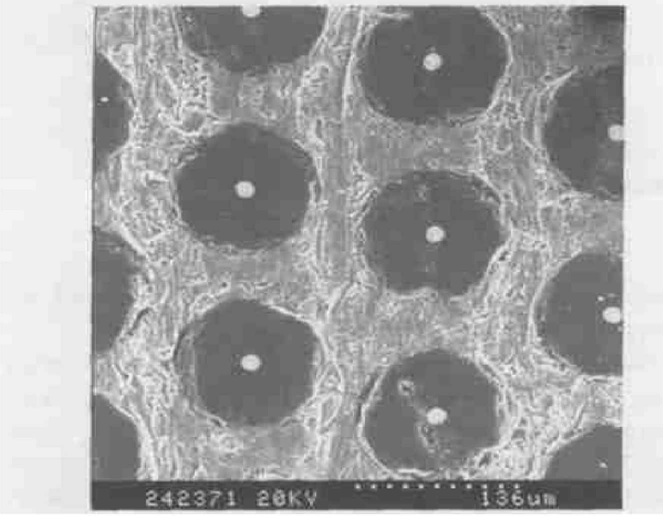

Fig. 1 Cross-sectional morphology of as-fabricated B/Al composite

Specimens for TMC test were cut from the B/Al composite sheet by electrical discharge machining. The size of the specimens was 125 mm × 10 mm × 1.4 mm. The longitudinal direction of the specimens was parallel to the fiber direction. After polishing and cleaning, the specimens were mounted in a pure copper clamp plate, with one end fixed to rigid frame and the other exerted with external tensile stress of 10 MPa. The applied stress was parallel to the longitudinal direction of the specimens. The TMC for 500, 1 000, 1 500 and 2 000 cycles was carried out in the temperature interval from − 125 ℃ to 125 ℃ with the heating rate of 1.25 ℃/s and the cooling rate of 2.5 ℃/s. The hold-time at both the lower and upper temperatures was 60 s.

A CM-12 transmission electron microscope (TEM), operated at 120 kV, was used to observe the microstructure of the specimens subjected to 1 000 or 2 000 cycles. The thin foils for TEM observation were prepared by an ion milling. After cycling to a predetermined number of cycles, the specimens were removed and aluminum alloy gripping tabs (25 mm × 10 mm × 2 mm) were attached to each end of the specimens with a contact adhesive. The room-temperature tensile tests were conducted on an Instron machine with sufficiently precise alignment. The crosshead speed was 0.5 mm/min. The tensile fracture surfaces were coated with Au thin layers and examined using a S-570 scanning electron microscope (SEM), operated at 20 kV.

3 RESULTS AND DISCUSSION

3.1 Microstructure

Fig. 2 shows the change in dislocation configur-

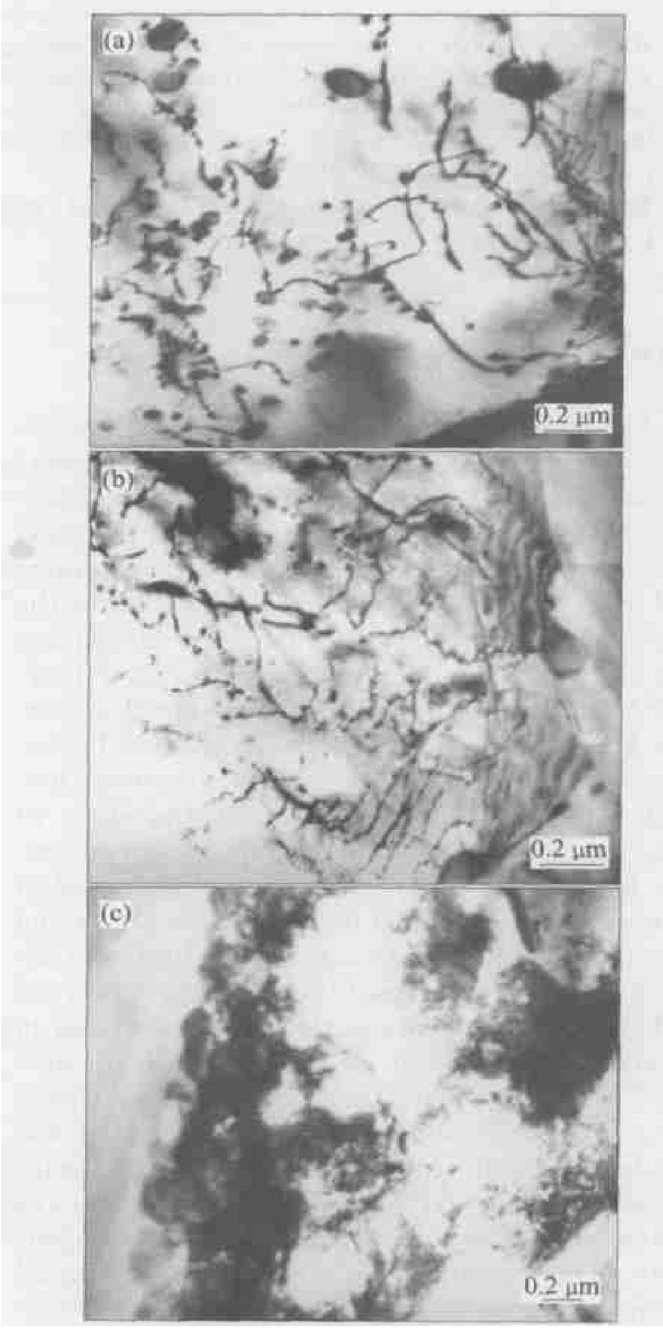

Fig. 2 TEM micrographs showing change in dislocation configuration in B/Al composites before and after TMC

(a) —As-fabricated; (b) —1 000 cycles; (c) —2 000 cycles ration in the matrix of B/Al composites after TMC. The dislocations are relatively straight in the as-fabricated B/Al composite, while are gradually tangled with increasing TMC cycles. The dislocation density increases with increasing TMC cycles.

3.2 Mechanical properties

It was found that the TMC exerted fewer effect on the stage II modulus of the B/Al composites, while the stage I modulus decreased significantly with increasing the cycles, as shown in Fig. 3.

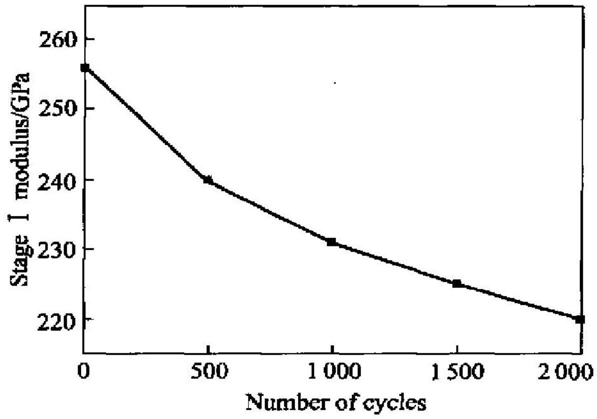

Fig. 3 Stage I modulus vs TMC cycles for B/Al composite

The ultimate tensile strength of the B/Al composite as a function of TMC cycles is shown in Fig. 4. It is indicated that the tensile strength increases with the number of cycles increasing.

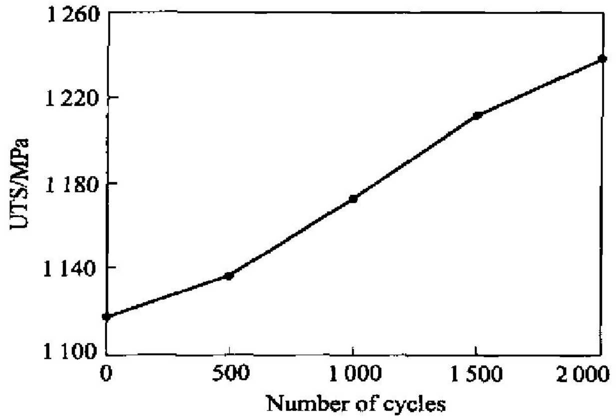

Fig. 4 Ultimate tensile strength vs TMC cycles for B/Al composites

3.3 Fractographs

Fig. 5(a) shows the tensile fracture surfaces for as-fabricated B/Al composite. The fracture path in the specimens is nearly planar, leading to a relatively flat fracture surface without evident pull-out of boron fibers. Each broken fiber remains round and is surrounded with the aluminum matrix with elongated dimples, indicating that the matrix fracture mode is ductile. Most of the B/Al interfaces still remain well bonded after the fracture. Cracks are initiated from the tungsten cores in the fibers and propagated quickly, and then cause fibers to be broken in advance of matrix fracture. The fracture surface shows that the B/Al composite in the as-fabricated condition has strongly-bonded interfaces.

The fractographs of B/Al composites subjected to TMC are shown in Fig. 5(b) and (c), suggesting a typical mixed failure mode consisting of the ductile matrix fracture, the flat fiber fracture and the fiber pull-out. When the number of TMC cycles is lower, most of broken fibers remain round and some fibers become elliptical on the fracture surfaces. The surfaces of the pull-out fibers can be clearly observed with aluminum matrix material adhering to it. The maximum pull-out height of boron fibers reaches approximately 300 μm, as shown in Fig. 5(b).

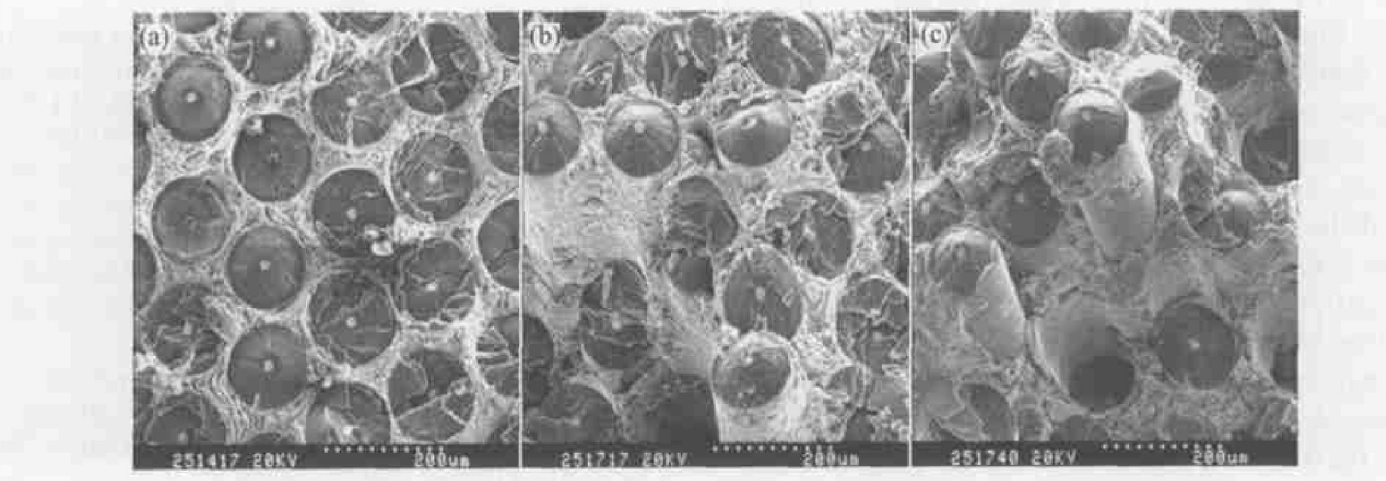

Fig. 5 Fractographs of B/Al composite before and after TMC

(a) —As-fabricated; (b) —1 000 cycles; (c) —2 000 cycles

When further increasing the number of TMC cycles, the notable feature of the fracture surfaces is that many fibers are pulled out smoothly, as shown in Fig. 5(c). The maximum pull-out height of boron fibers is about 400 μm. The surfaces of the pull-out fibers are not covered with a residual layer of aluminum and the groove is accordingly left cleanly at the interface, indicating that the bonding of fiber/matrix interfaces becomes poorer than that in the early stage of TMC.

3.4 Discussion

The mixture rule used to predict the elastic modulus of FMMCs is based on the perfect bonding interfaces[10, 11]. The elastic modulus of the composites can be mainly determined by the load-transfer from the matrix to fibers and only the degradation of the interface will decrease the elastic modulus. Therefore, the decrease of elastic modulus of the B/Al composites suggests that the interfaces are degraded during TMC. For the B/Al composite, the coefficient of thermal expansion (CTE) of boron fiber is 5×10^{-6} ℃$^{-1}$, while that of the aluminum matrix is 24×10^{-6} ℃$^{-1}$[12]. Therefore, during thermal cycling, the thermal mismatch strains occur due to the great difference of the coefficients of thermal expansion between the matrix and fibers. The strain induces thermal residual stresses, which will be superposed upon externally applied stress, and may be sufficient to cause "fatigue" at the fiber/matrix interfaces and then lead to the degradation of interfacial bonding, the extent of which increases with increasing the cycles. The interfacial degradation results in a measurable decrease of elastic modulus of the B/Al composite during TMC.

The change in ultimate tensile strength of the B/Al composites under TMC can be analyzed with respect to the matrix strengthening and the interfacial degradation. During thermal cycling, the thermal mismatch stresses can be induced due to the great CTE difference between the matrix and fibers. For example, during cooling of the TMC, the matrix is subjected to a very large tensile stress, while a compressive stress during heating. The thermal residual stresses superposed on external tensile stresses may be sufficient to result in the dislocations generation in the aluminum matrix (see Fig. 2). Consequently, the matrix can be strengthened, improving the strength of the composites during TMC. In a word, the increase in the strength of B/Al composite is related to the increase of dislocation density in the matrix.

On the other hands, the interfacial strength can also strongly influence the ultimate tensile strength of the B/Al composite. Neither the strongly-bonded nor the weakly-bonded interfaces are favorable to the strength of the composite[13-15]. Only the interfaces with an appropriate interfacial strength may be beneficial to the strength of the B/Al composite. The as-fabricated B/Al composite exhibits a strongly-bonded interface feature (see Fig. 5(a)). After subjected to the TMC with the applied stress of 10MPa in the temperature interval of −125 ℃−125 ℃, the interfaces are degraded and an appropriate interfacial strength gradually develops (see Fig. 5(b) and (c)). Therefore, the tensile strength of the B/Al composite will increase with the TMC cycles on the base of the matrix strengthening and the slight interfacial degradation.

4 CONCLUSIONS

1) The interfaces will be degraded during TMC, the extent of which is enhanced with increasing the cycles, causing a measurable decrease of the stage Ⅰ modulus of the B/Al composite.

2) The thermal residual stresses superposed upon externally applied tensile stresses can be sufficient to induce the dislocation generation in the aluminum matrix. With increasing the TMC cycles, the dislocation density increases.

3) With increasing the TMC cycles, the interfaces in the B/Al composite change from the strongly-bonded to the appropriately-bonded and the ultimate tensile strength of the composite increases. This effect can be ascribed to the matrix strengthening and the interfacial degradation during TMC.

4) TMC will provide an approach of improving the strength of B/Al composite.

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