TENSILE PROPERTIES OF COATED CARBON FIBER REINFORCED MAGNESIUM COMPOSITES①

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ABSTRACT The effects of fiber coating (C/SiC/SiO_2) on interfacial characteristic and fracture behavior of carbon fiber reinforced magnesium composites have been investigated in an effort to better understand the factors contributing to the tensile strength. Varying the coating thickness was found to strongly affect axial strength of the composites, basically consistent with the coating model. Due to the narrow range of the appropriate coating thickness, its optimum value was given.

Key words C/Mg composites coating thickness tensile properties

1 INTRODUCTION

In continuous fiber reinforced metal matrix composite materials, the fiber/matrix interfacial zone affects many of the mechanical properties of interest. In order to control the structure and the properties in this zone, many ways including fiber coating have been adopted. In our previous work[1], a C/SiC/SiO_2 coating on carbon fiber suitable for C/Mg composites was designed according to following principles: the inside pyrocarbon layer near carbon fiber was used to control interfacial bonding strength, the outside silica layer acted as wetting agent, the intermediate silicon carbide layer between pyrocarbon and silica was used to protect the fibers against attack by magnesium. This coating demonstrated a good efficiency and the corresponding composite exhibited high ultimate tensile strength (UTS) up to 1 050 MPa. In order to better understand the factors contributing to the tensile strength, the present study deals with the effects of the coating thickness on interfacial characteristic and fracture behavior of composite.

2 EXPERIMENTAL

The composites used in this investigation were prepared by a vacuum infiltration technique. The matrix was a Mg-4Al alloy, its ultimate tensile strength (UTS) and yielding strength in cast were 170 MPa and 90 MPa respectively. Continuous PAN-based fibers with a tensile strength of 3 300 MPa were used as the unidirectional reinforcement and its nominal volume fraction was 33%. A chemical vapor deposition (CVD) process was used to prepare C/SiC/SiO_2 coating by coating these three films successively on fiber surface. The thicknesses of each layer and finally formed coating was measured by a scanning electron microscopy (SEM). Since silica layer will react with molten magnesium during infiltration, its thickness was predetermined to be 0.05 μm so as not to interfere with metallurgy of the matrix.

Cylindrical composite tensile samples with a diameter of 6 mm and a gauge length of 30 mm were tested on an Instron 1196 testing machine at a cross-head speed of 0.5 mm/min. At least five samples were tested in each condition. Finally, the broken samples were subjected to fractographic examination using SEM.

3 RESULTS AND DISCUSSION

3.1 Effect of pyrocarbon layer thickness

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Fig. 1 shows the variation of the UTS of composites as a function of the thickness of pyrocarbon layer, h_C, while the SiC layer thickness remains a theoretically permissible value of 0.2 μm. As can be seen from Fig. 1, the UTS increases, reaches maximum, and then decreases with increasing h_C. Thus, it is found that there is an optimum value of h_C to obtain high-strength composite. This variation is consistent with that of UTS as a function of interfacial bonding strength and indicates reasonableness of following implications. For one thing, the interfacial bonding strength is correlated to h_C and can be controlled by varying h_C. Secondly, if the interfacial bonding strength is measured experimentally, its variation as a function of h_C will be obtained. This is of great value and is not yet available. Thirdly, pyrocarbon with the optimum value of h_C = 0.05 μm, at which the UTS of composites investigated in this work is the highest, makes fiber capability fully exploit during loading, whereas above or below this value reduces reinforcing efficiency.

There are obvious changes in the fractograph of the tensile specimens with different h_C. On a fiber scale, the fracture surface of composite without pyrocarbon layer is flat with little pull-out of fibers (Fig. 2a), while that with pyrocarbon layer thickness of 0.05μm shows fibrous with a modest amount of pull-out of the fibers (Fig. 2b), and that with pyrocarbon layer thickness of 0.10 μm shows severe longitudinal shear and extensive fiber pull-out (Fig. 2c). All of these fracture surfaces are typical of strong, weak and too weak interfacial bonding[2], which further confirms that the changes in h_C are directly related to changes in the interfacial bonding strength, and consequently to changes in the UTS.

In general, the fracture of the composite is believed to be initiated by random fracture of

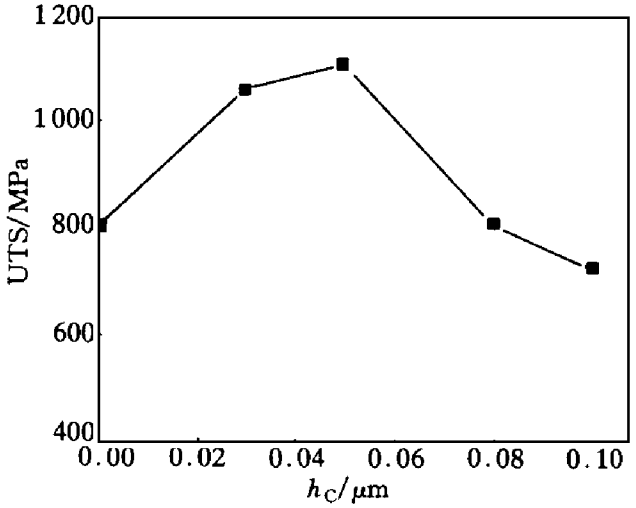

Fig. 1 The effects of h_C on the UTS for C/Mg composite

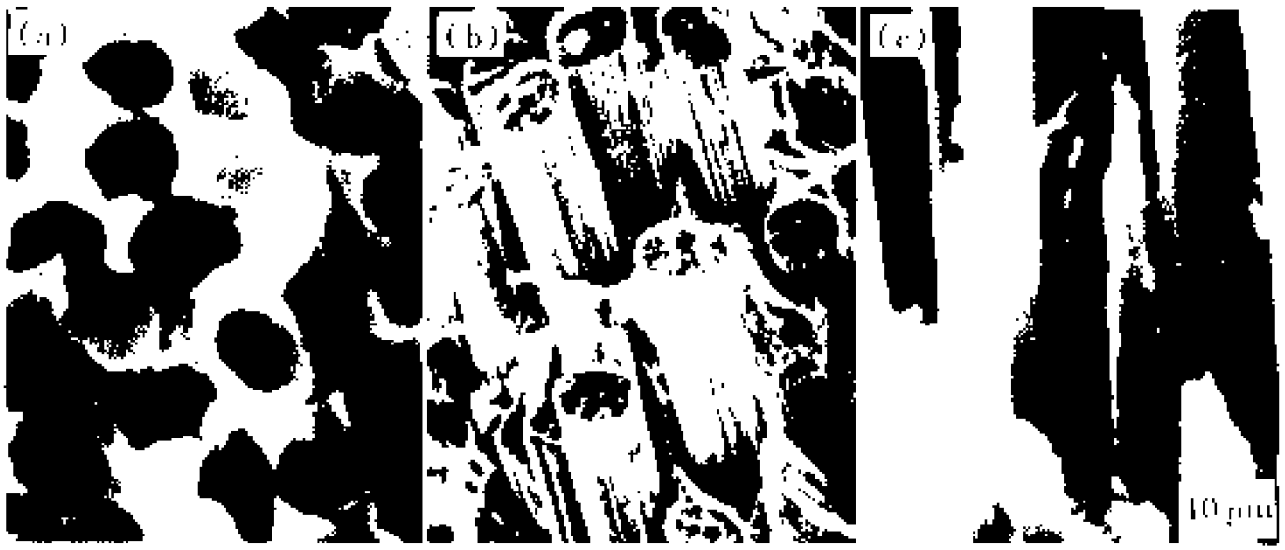

Fig. 2 SEM micrographs of C/Mg composites

(a) —without pyrocarbon layer; (b) —h_C = 0.05 μm; (c) —h_C = 0.15 μm

the weakest fibers because of the statistic scatter of the fiber strength, cracks formed by the fiber breakage will induce a stress concentration on adjacent fibers. Without pyrocarbon layer, the composite features a strong interface bonding due to the chemical interaction between the silica layer and magnesium[3]. Therefore, interfacial debonding cannot occur and further fiber fracture occurs due to the stress concentration. The critical defect sizes required to initiate brittle failure of the composite are on the order of a few fiber diameters, the failure of several adjacent fibers in the same plane may be sufficient to initiate failure of the composite. The crack can easily propagate across the fibers with little deflection along the interface. As a result, the fracture surface is relatively flat and fracture stress is low. It should be noted, however, in the same state of strong interface and resultant brittle failure, composite with a single silica coating has a much lower UTS[4] of 640 MPa than the composite investigated here. This should be due to the former composite lacking of the SiC layer, and it is known that, without the protection of the SiC layer, the penetration of magnesium into the carbon fibers as well as the reaction between aluminum (alloying element) and the fibers further prevent the realization of the strengthening potential[4].

In the composite with pyrocarbon thickness of 0.05 μm, there exists a weak interface due to a small fracture energy ($< 1\ J/m^2$) of pyrocarbon[5]. The stress concentration induced by fiber breakage may be eliminated or reduced to a rather low degree through interfacial debonding. In such a case, small amounts of fiber breakage do not lead to brittle fracture of the composite. Conversely, with the increment of applied stress, the others will break separately from their own flaws. When only a few amounts of fibers break, microcrack due to fiber breakage exhibits no interaction because of random flaw distribution. Upon loading to a certain stress level, the amount of microcrack becomes larger and some of microcrack may be connected to microscopic shearing crack through interfacial debonding and matrix shearing. Subsequent damage accumulation leads to composite failure by rapid propagation of the macroscopic shearing crack. Since the initiation and propagation of these different cracks occur at different stress levels, they can only be joined by matrix shearing and interfacial splitting. In other words, the fracture occurs with a cumulative mode where fibers break at a maximum stress as given by the Weibull distribution function and this is reflected in the high UTS. It is worth noting that the idea of weakening interface to gain high UTS can also be applied to other composite systems, such as C/Al and SCS/Ti composites[6, 7].

If the stress concentration induced by fiber breakage may be eliminated through interfacial debonding, load transfer capability from the matrix to the fibers has a dominant effect on the UTS of the composite. The pyrocarbon layer with a thickness of 0.10 μm seems to be associated with too weak interfacial bonding. During loading, it is very easy for the microcrack originated by fiber breakage to propagate and the microscopic shearing crack is formed at a lower stress. This reduces dramatically the load transfer efficiency due to a large length of the fibers which can not support the full load. Upon further loading and progressive occurring of shearing crack, the composite has to rely a large scale upon interfacial splitting and matrix shearing. Finally the composite fractures at a lower stress and its fracture surface is very fibrous in appearance.

3.2 The effect of SiC layer thickness

Fig. 3 shows the variation of the UTS of the composites as a function of SiC layer thickness, h_{SiC}, while h_C remains the optimum value of 0.05 μm. As can be seen, the UTS decreases with increasing h_{SiC} when h_{SiC} is above 0.20 μm. This is consistent with the coating model[1] and is considered to be the degradation of coated fiber strength due to h_{SiC} beyond the permissible value.

However, the UTS of the composite increases with increasing h_{SiC} when h_{SiC} is below 0.20 μm. This variation is in disagreement with the model in which the UTS should remain constant, and may be interpreted from the view-

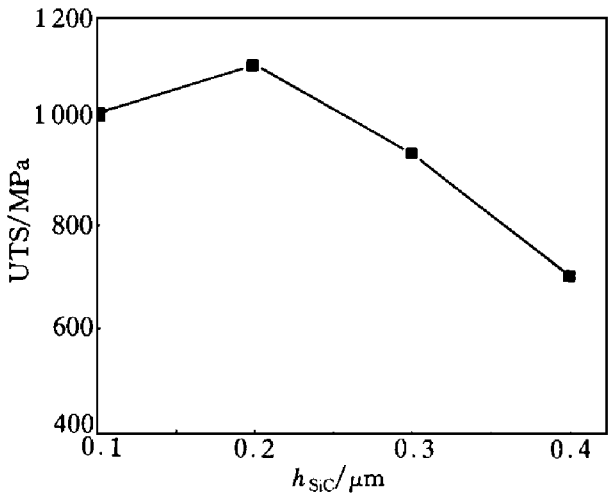

Fig. 3 Effect of h_{SiC} on the UTS for C/ Mg composites

point of internal stress. Typically metal matrix composites are fabricated at high temperatures, as the composite cools, the matrix begins to solidify and stresses are generated because of the differences among the thermal expansions of the fibers and the matrix. In the case of C/ Mg composite, the matrix contracts and the fibers expand as the material cools, thereby generating undesirable residual stress in all phases of the composite. For instance, if a model based on a two-dimensional composite plate which consists of a thin top plate (interphase) of thickness h and a thick substrate of thickness t, the stress in the interphase then can be obtained by[2]

$$\sigma = \frac{E_s t^2}{6(1-\nu_s)hR} \quad (1)$$

where E_s and ν_s are the Young's modulus and the Poisson's ratio of the substrate respectively, and R is the curvature of the composite plate.

It can easily be seen that the thicker the interphase layer, the lower the stress. In other words, the use of a thick coating on the fibers is more beneficial to mitigate potentially harmful residual stresses and raises the composite strength.

4 CONCLUSION

The interfacial characteristic and fracture behavior of carbon fiber reinforced magnesium composites are strongly influenced by the thickness of the C/ SiC/ SiO_2 coating on carbon fibers. The optimum thicknesses of pyrocarbon layer, the silicon carbide layer and the silica layer are 0.05, 0.20 and 0.05 μm respectively, at which the composite exhibits ideal interfacial bonding and the highest ultimate tensile strength.

REFERENCES

1 Zhang Kun *et al*. Acta Mater Compos Sinica, (in Chinese), 1996, 13(3): 44.
2 Peters P W M. In: Ochiai S (eds), Mechanical Properties of Metallic Composites. Marcel Dekker, Inc., 1991.
3 Katzman H A. J Mat Sci, 1987, 22(1): 144.
4 Zhang Kun. PhD thesis, (in Chinese). Northwestern Polytechnical University, 1995.
5 Evans A G, Marshall D B. Acta Metall, 1989, 37 (10): 2567.
6 Yu J K, Li H L, Shang B L. J Mat Sci, 1994, 29 (1): 2641.
7 Yang J M, Jeng S M, Yang C J. Mat Sci Eng, 1991, A138(2): 155.

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