Room and Elevated Temperature Tensile Properties of SiC\textsubscript{f}/TC17 Composite

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SiC\textsubscript{f}/TC17 composite was prepared from precursor wires produced by magnetron sputtering. The results showed that full density of the composites was achieved and no porosity was detected. The fibers were arranged well forming a nearly hexagonal array. The longitudinal tensile properties of the SiC\textsubscript{f}/TC17 composite have been investigated at room and elevated temperature. The tensile strength of the composite was significantly increased, to 1717 MPa and 1341 MPa at room temperature and 500\textdegreeC respectively, due to the incorporation of SiC fiber. Scanning electron microscopy (SEM) was used to study the matrix deformation, interfacial debonding and fiber pull-out. Different damage models have been obtained by comparing the stress-strain curves and fractographs of tensile samples at room and elevated temperatures. The damage processes of SiC\textsubscript{f}/TC17 composite under tensile loading at room and elevated temperature are discussed.

Keywords: Titanium matrix composites, SiC fiber, magnetron sputtering, tensile performance

1. Introduction

It is believed that continuous SiC fiber reinforced titanium alloy matrix composites (TMCs) could be promising candidate materials for light weight components of aerospace structural applications because of their high specific strength, modulus and stiffness at room and elevated temperature\textsuperscript{2).}

The mechanical properties of TMCs, especially longitudinal tensile performances, are sensitive to the testing temperature because it affects the interface bonding strength and matrix yielding behavior. In the process of loading, the breakage of fiber initiated at the weakest fiber for intermediate and strong bonding, then the loading shed by the broken fiber was transferred to the neighbor fibers. However, for the weak bonding, the load shed by the fractured fibers was distributed equally among the intact fibers\textsuperscript{25).} Deformation of the matrix alloy leads to yielding of the TMCs. The yielding stress of TMCs was similar to the unreforced matrix alloy\textsuperscript{2).} Therefore, the mechanical properties of the TMCs also depends on the type of the matrix alloy.

Recently, a number of investigations have been conducted relating to the tensile properties, fracture mechanism and strength prediction of SiC\textsubscript{f}/Ti-6Al-4V composites\textsuperscript{19),} but the performance and damage processes of SiC\textsubscript{f}/TC17 composites were rarely reported. TC17 (Ti-5Al-2Sn-2Zr-4Cr-4Mo) alloy has a greater fracture strength and creep resistance and can be used at higher temperature than Ti-6Al-4V.

In this work, SiC\textsubscript{f}/TC17 composite was prepared by a method of precursor wire from magnetron sputtering, and the longitudinal tensile properties of the SiC\textsubscript{f}/TC17 composite were investigated at room and elevated temperature. Scanning electron microscopy (SEM) was used to observe the damage characteristics on fracture surface and longitudinal section. The damage processes of the SiC\textsubscript{f}/TC17 composite under tensile loading at room and elevated temperature are discussed.

2. Material Preparation and Tensile Tests

The material used in this study was SiC\textsubscript{f}/TC17 composite, SiC fibers coated with a C layer were produced by chemical vapor deposition of SiC on a tungsten wire at the Institute of Metal Research, Chinese Academy of Sciences. Fracture strength and diameter of the SiC fiber were 3500 MPa and 100 \textmu m, respectively. The composite material was processed by deposition of the TC17 alloy onto the SiC fibers using the magnetron sputtering process. The SiC\textsubscript{f}/TC17 composite precursor wires were cut to the required length and stacked into a tube of TC17; after vacuum pumping and sealing they were consolidated by hot isostatic pressing (HIP).

![Figure 1](image-url)
A representative cross-section of the composite rod was given in Figure 1(a), which showed the quality of the TMCs. The fibers were arranged well forming an imperfect hexagonal array. The micrograph also revealed that the matrix filled the space completely between the fibers and no porosity was detected which showed that the fabrication process was appropriate. There was a continuous interfacial reaction layer of about 1 μm thick around the fiber, Figure 1(b). Several researchers suggested that the main reaction product was brittle TiC which may cause the increase of interfacial hardness and brittleness.\textsuperscript{10-11}

Uniaxial tensile tests were performed on cylindrical specimens with a gauge diameter of 4.5 mm. The fiber volume fraction in the gauge length was about 30%. Uniaxial tensile tests were carried out for the unreinforced matrix and the TMCs at room temperature (RT) and 500°C in a servo-hydraulic testing machine to obtain the yield strength and the fracture strength.

3. Results

3.1 Tensile Tests

Stress-strain curves of TC17 and SiC/TC17 composite at room temperature and 500°C were plotted in Figure 2. Tensile test results were summarized in Table 1. It could be seen that TC17 has a yield strength of 959 MPa at room temperature which reduces to 613 MPa at 500°C; fracture strength of the TMC is 1717 MPa and 1341 MPa at room temperature and 500°C, respectively. Therefore, the mechanical behavior was improved significantly at room temperature and 500°C due to incorporation of the fibers.

![Stress-strain diagram for the SiC/TC17 composite and the TC17 alloy. (TMCs; curves were plotted to fracture. TC17; curves were interrupted at 2% strain)](image)

<table>
<thead>
<tr>
<th>Material</th>
<th>Testing Temperature (°C)</th>
<th>Yield strength (MPa)</th>
<th>Fracture Strength (MPa)</th>
<th>Failure strain (%)</th>
<th>Elastic modulus (GPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>SiC/TC17</td>
<td>25</td>
<td>1150</td>
<td>1717</td>
<td>0.91</td>
<td>195</td>
</tr>
<tr>
<td>SiC/TC17</td>
<td>500</td>
<td>—</td>
<td>1341</td>
<td>0.79</td>
<td>184</td>
</tr>
<tr>
<td>TC17</td>
<td>25</td>
<td>959</td>
<td>1059</td>
<td>&gt;8.00</td>
<td>113</td>
</tr>
<tr>
<td>TC17</td>
<td>500</td>
<td>613</td>
<td>726</td>
<td>&gt;8.00</td>
<td>77</td>
</tr>
</tbody>
</table>

The stress-strain curve of SiC/TC17 composite at room temperature has a bilinear shape, typical for composites when loaded along the fiber direction\textsuperscript{10-12}. The elastic region I (characterized by the elastic modulus) ended when the matrix began to deform plastically. The knee point was taken as the composite yield strength, and it was followed by another linear region II with a smaller slope. At the start and the end of region II there were some slight fluctuations of the straight line, which indicated that fracture may occur in the matrix, fiber or interface. Thomas and Winstone\textsuperscript{59} considered that a large number of cracks were generated in the matrix at the start of region II and the amount of fiber breakage rapidly increased at the end.

The stress-strain curve of SiC/TC17 composite at 500°C was a nearly straight line with slight curvature which was mainly borne by the elastic deformation of the fibers and the plastic deformation of the matrix. With the increase of the load, deformation of the matrix was no longer synchronized with the fiber because of interfacial debonding, then plastic deformation of the matrix occurred. So the tensile process of composite materials is not of purely elastic deformation.

3.2 Fractography

Figure 3 presents typical tensile fracture surfaces of the SiC/TC17 composite at room temperature. From Figure 3(a), two different regions were observed on the fracture surface, corresponding to a flat region (region I) and a ductile rough region (region II). On the entire fracture surface only a small number of fibers were pulled-out and the average pull-out length was about 0.2 mm. The pull-out length of the fiber depends on the fiber strength and the interfacial shear strength. In the flat region I, matrix surface was flat and the fracture pattern is divergent along the growth direction of columnar crystals on the surface of the precursor wires. Degree of interface debonding was very small, and a few residual C-coating layers were observed to attach to the edge of the groove caused by some fiber pull-out, as shown in Figure 3(b), which indicated that interface bonding was strong in this region.
In order to reveal the fracture mechanism of the fiber, the polished longitudinal cross-section of fracture samples at room temperature was observed. The longitudinal cross-section micrographs were shown in Figure 4(a). Multiple fracture of each fiber was observed and the fracture plane was almost perpendicular to the loading direction. Fiber fracture occurred just under the fracture surface at a distance of approximately 0.1 ~ 0.4 mm. Fiber fracture was not observed far away from the fracture surfaces. Matrix morphology close to the fracture surface was shown in Figure 4(b). The fracture surface of the matrix was flat with little plastic deformation, and nearly no internal micro-cracks could be detected.

Figure 5(a) shows a longitudinal cross-section of a fractured sample. More multiple fractures occurred on each fiber; the fracture plane of fiber was almost perpendicular to the loading direction and some fracture surfaces were aligned at a certain angle to the loading direction. In Figure 5(b), an example of the matrix morphology near the fracture surface with extensive plastic deformation, dominated by lots of micro-cavities was shown.

4. Discussion

When composites were loaded along the fiber direction, the damage mechanisms include matrix deformation and failure, fiber fracture, frictional sliding and pull-out, interface debonding and crack deflection, etc. The presence of strong or weak bonding influenced the damage mechanisms of TMCs, thereby affecting the fracture processes of the composites properties. So the fiber/matrix interface strength was critical and
played the most important role in failure.

At room temperature, SiC/TC17 composite samples showed the characteristics of strong bonding; the composite tensile behavior was controlled by the strain of the fiber, and composite tensile fracture determined by fiber fracture. Both the fiber and matrix deformed elastically in region I of the stress-strain curve (up to approximately 0.55% strain). The fiber continued to deform elastically, at the same time the matrix deformed plastically at the beginning of region II of the stress-strain curve. Because of the residual tensile stress, the real yield strength of the matrix was lower than that of the unreinforced alloys. Then the fibers near the yielding matrix carried more loads, and with the load increasing, the fibers started breaking due to the presence of randomly distributed flaws in the fibers. The cracks occurred in the fibers either split the C-coating layer extending along longitudinal direction, or are blunted leading to an increase in stress right at the crack tip around the interface reaction layer (Figure 7(a)). The loading shed by the broken fiber was transferred to the neighbor fibers. With the load increased, the distributed fiber fractures continued extending. When a cluster of adjacent broken fibers reached some critical size, the catastrophic failure of the composite was triggered. The flat region (region I on the fracture surface at RT) could be formed when the cracks initiated and extended. However, the rough region (region II on the fracture surface at RT) could be formed when the catastrophic failure of the composite was triggered.

![Figure 7. Damage mechanisms of the SiC/TC17 composite at room temperature (a) and 500°C (b)](image)

At elevated temperature (500°C), interface bonding of SiC/TC17 composite samples was weak, and the composite fracture is dominated by fiber fracture and matrix yielding. In the process of tensile loading, the weakest fiber failed firstly, and interface debonding and crack deflection took place immediately, see Figure 7 (b). At this time, for the matrix, load transfer from the matrix to the fibers was interrupted. In addition, the residual tensile stress did not completely release and the yield strength of the matrix at 500°C was lower than that at room temperature (613 MPa and 959 MPa respectively). So matrix yielding could occur at a smaller stress, For the fiber, the load shed by the fractured fibers was transferred homogeneously to the other intact fibers at the same section, and the other fibers fractured randomly. With the load increasing, the fractured fibers were pulled out with the frictional sliding at C-coating layer/reacton layer interface and lots of micro-cavities were formed inside of the matrix. When the randomly fractured fibers reached some critical volume, the cracks formed by broken fibers and micro-cavities in the matrix propagated and connected, and catastrophic failure of the composite was triggered. Therefore, the irregular fracture surface of SiC/TC17 composite was obtained.

5. Conclusions

(1) The yield strength of TC17 alloy was 959 MPa at room temperature which reduced to 613 MPa at 500°C. The fracture strength of SiC/TC17 TMC was 1717 MPa at room temperature, and 1341 MPa at 500°C.

(2) Matrix yielding contributed to the shape of the stress-strain curves of SiC/TC17 composite, and the curve was bilinear at room temperature, and has a slight curvature at 500°C.

(3) Interface strength played the most important role in determining the failure morphology and damage mechanisms. At room temperature, fracture surface was mainly a flat region, and almost brittle fracture occurred in the matrix. At elevated temperature, the fracture morphology was dominated by fiber pull-out, giving an extremely irregular surface, and the matrix broke in a ductile manner with dimples.

(4) At room temperature, the tensile behavior of SiC/TC17 composite was controlled by fiber fracture; at 500°C, the tensile behavior of SiC/TC17 composites was dominated by fiber fracture and matrix yielding simultaneously.

REFERENCES


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