Tensile Behavior of SiCf/TC17 Composites (Postprint)

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Abstract

This study investigates the tensile properties at room temperature and elevated temperature (773 K), along with the fracture mechanisms of SiCf/TC17 composites. Results demonstrate that the stress-strain curves of SiCf/TC17 composites at room temperature and elevated temperature exhibit distinct characteristics, influenced by fiber linear elastic deformation and the extent of matrix yielding. The room-temperature fracture mechanisms primarily involve multiple fractures of the reaction layer, single fracture of fibers, and brittle fracture of the matrix, whereas the elevated-temperature fracture mechanisms mainly consist of multiple fiber fractures, ductile fracture of the matrix, and extensive interface debonding. The fiber cumulative damage theory proves suitable for estimating the fracture strength of SiCf/TC17 composites, wherein the room-temperature fracture strength conforms to the local load-sharing model when the critical number of fractured fibers is greater than or equal to 3, while the elevated-temperature fracture strength follows the global load-sharing model. By integrating the fracture mechanisms and strength estimation results, the tensile fracture process of SiCf/TC17 composites at room temperature and elevated temperature is elaborated in detail.

Full Text

Preamble

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Study on Tensile Behavior of SiCf/TC17 Composites

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Abstract
The tensile properties and fracture mechanisms of SiCf/TC17 composites at room temperature and 773 K were investigated. The results show that the shapes of the stress-strain curves are influenced by fiber elastic deformation and matrix yielding, exhibiting a bilinear appearance at 298 K and slight curvature at 773 K. The primary room-temperature fracture mechanisms include multiple fractures of the interfacial reaction layer, single fiber fracture, and matrix brittle fracture, while elevated-temperature fracture mechanisms consist mainly of multiple fiber fractures, matrix ductile fracture, and extensive interface debonding. Fiber cumulative damage theory is suitable for estimating the fracture strength of SiCf/TC17 composites, where the room-temperature fracture strength follows the local loading sharing model when three or more fibers fail, and the elevated-temperature fracture strength follows the global loading sharing model. Based on fracture mechanisms and strength estimation results, the tensile fracture processes at room and elevated temperatures are discussed in detail.

Keywords
Titanium matrix composite, SiC fiber, tensile property, fracture mechanism, fracture process

Introduction

Continuous SiC fiber-reinforced titanium matrix composites offer significantly higher specific strength, specific modulus, and service temperature than conventional titanium alloys, making them highly promising for applications such as rotors and blades in aero-gas turbine engines, where they can substantially reduce engine weight and improve efficiency [1,2]. The tensile properties along the fiber axis represent one of the most important mechanical characteristics of these composites and have received extensive attention and in-depth research. Influencing factors include matrix and fiber properties [3,4], fiber volume fraction [5], composite processing [6], gauge length [7], deformation rate [8], and service environment [9,10], rendering analysis of tensile properties and failure mechanisms particularly complex and subject to varying interpretations. The widely accepted failure mechanism is the fiber cumulative damage model, wherein weak fibers undergo random fracture during loading due to strength dispersion, transferring their load to intact fibers. Depending on whether stress concentration from broken fibers alters the randomness of cumulative damage, the model is categorized as either global loading sharing (GLS) or local loading sharing (LLS) [5,11]. The GLS model assumes that stress concentration is insufficient to change the random nature of fiber damage, with load shared by all unbroken fibers across the section, whereas the LLS model considers local stress concentration non-negligible, with load from broken fibers transferred to neighboring fibers, increasing their fracture probability.

Despite existing failure models and fracture mechanisms for SiC fiber-reinforced titanium matrix composites, a complete description of the fracture sequence of fibers, reaction layers, and matrix, as well as the overall fracture process, remains lacking, particularly for SiCf/TC17 composites, whose mechanical properties are rarely reported. TC17 alloy, a key material for manufacturing turbine engine disc forgings, has become a preferred matrix for titanium matrix composites. Therefore, investigating the room- and elevated-temperature tensile properties and fracture mechanisms of SiCf/TC17 composites holds significant practical importance for their application. In this work, SiCf/TC17 composites were fabricated using magnetron sputtering precursor wire method combined with hot isostatic pressing (HIP), and tensile tests were conducted at 298 K and 773 K. By analyzing fracture surface morphology and internal microstructure, combined with calculations from fiber cumulative damage models, the room- and elevated-temperature tensile fracture mechanisms were proposed, and the complete fracture processes were summarized.

Experimental

The SiC fibers used in this study were produced by chemical vapor deposition (CVD) with a diameter d_f of 103 μm and a carbon coating thickness of 1.6 μm. The characteristic strength σ_0 calculated using Weibull distribution was 3611.4 MPa, with a Weibull modulus m of 8.0. The nominal composition of TC17 alloy is Ti-5Al-2Sn-2Zr-4Mo-4Cr (mass fraction, %). TC17 alloy was deposited onto SiC fiber surfaces using a dual-target magnetron sputtering system to prepare SiCf/TC17 composite precursor wires. The precursor wires were then placed in TC17 alloy cans, vacuum-sealed, and HIP-processed to obtain SiCf/TC17 composite rod samples.

Both SiCf/TC17 composites and TC17 alloy rods subjected to the same HIP process were machined into tensile specimens with the profile dimensions shown in [Figure 1: see original paper]. For SiCf/TC17 composite specimens, the core composite diameter in the gauge section was 3.6 mm, with a composite can wall thickness of 0.45 mm, and the fiber volume fraction V_f throughout the specimen was 33%. Specimen surfaces were mechanically polished to eliminate machining marks. Tensile testing was performed on an Instron 5582 universal testing machine at 298 K and 773 K, with the fiber axis parallel to the loading direction and an average strain rate of 1 mm/min.

Fracture surfaces and longitudinal sections near the fracture were examined using an S-3400N scanning electron microscope (SEM). Longitudinal section samples were prepared by first cutting a section 5 mm below the fracture surface perpendicular to the fiber direction using wire electrical discharge machining. The samples were then hot-mounted, mechanically ground, and polished. To avoid fiber damage during grinding, sandpapers with sequentially finer grits of 600, 1000, and 2000 were used. Finally, the sample surfaces were etched using a titanium alloy etchant (3% HF + 10% HNO_3 + 87% H_2O by volume).

Results

2.1 Microstructure of SiCf/TC17 Composites

[Figure 2: see original paper] shows the microstructure of the HIP-processed SiCf/TC17 composite rod. The fibers are uniformly distributed in a near-hexagonal arrangement, with no evidence of pores, gaps, or incomplete consolidation across the entire cross-section. As shown in [Figure 2b: see original paper], the SiC fibers are well-bonded to the matrix, forming an interfacial reaction layer approximately 1 μm thick, composed primarily of TiC [12,13], which increases hardness and brittleness in the interface region. After HIP, the residual unconsumed carbon coating thickness is about 1.6 μm, with no cracking observed at the three interfaces: C coating/fiber, C coating/reaction layer, and reaction layer/matrix. Additionally, the matrix surrounding the fibers exhibits fine equiaxed $\alpha+\beta$ microstructure, where the dark gray regions in [Figure 2b: see original paper] correspond to the α phase and the light gray regions to the β phase.

2.2 Tensile Properties at Room and Elevated Temperatures

The room- and elevated-temperature tensile properties of SiCf/TC17 composites and TC17 alloy are listed in . The composite fracture strengths at 298 K and 773 K are 1717 MPa and 1341 MPa, respectively, representing improvements of 62.1% and 84.7% over the TC17 matrix. The fiber reinforcement effect is more pronounced at elevated temperature because matrix properties degrade significantly while fiber properties remain largely unchanged [10]. Additionally, compared with room temperature, the matrix elastic modulus decreases by 36 GPa at 773 K, whereas the composite modulus decreases by only 12 GPa, indicating that the composite modulus is primarily controlled by the fiber modulus since fiber modulus shows no obvious decrease at elevated temperature.

[Figure 3: see original paper] presents the tensile stress-strain curves for SiCf/TC17 composites and TC17 alloy at room and elevated temperatures. The composite exhibits a typical bilinear stress-strain curve at room temperature, with the inflection point at 1150 MPa considered the yield strength [7,14]. Before the inflection point, the composite undergoes elastic deformation with linear stress-strain behavior, where the slope represents the composite elastic modulus. Beyond the inflection point, slight matrix yielding reduces the slope. Small fluctuations appear at the beginning and end of the second linear segment, indicating possible local fracture of matrix, fibers, or interfaces. At 773 K, the stress-strain curve shows near-curvilinear characteristics due to reduced matrix yield strength and interface debonding during loading, which decreases load transfer capability and causes extensive matrix plastic deformation, altering the elastic nature of the stress-strain curve. The yield strength cannot be directly obtained from the 773 K curve but can be calculated indirectly using the matrix stress-strain curve at 773 K and considering residual stress effects. The TC17 alloy yield strength at 773 K is 613 MPa ([Figure 3: see original paper]), and finite element calculations [15] show an axial residual tensile stress of 166 MPa in the matrix at 773 K, meaning the matrix yields prematurely at 447 MPa (TC17 yield strength minus axial residual stress). This stress corresponds to a strain of 0.57% on the TC17 alloy stress-strain curve at 773 K, which represents the composite yield strain. Using the same method, the composite yield strength is calculated as 1006 MPa, showing only -2.1% deviation from the value obtained directly from the stress-strain curve (1150 MPa), confirming the accuracy of this method. However, this approach is unsuitable for estimating fracture strength because composite fracture is governed by fiber strength rather than matrix strength, and after matrix yielding, fiber fracture and interface debonding cause fiber strain, matrix strain, and composite strain to diverge.

2.3 Fracture Morphology

2.3.1 Room-Temperature Fracture Morphology [Figure 4a: see original paper] shows the overall room-temperature tensile fracture surface of SiCf/TC17 composite, clearly revealing the external can and core composite. The typical can fracture morphology ([Figure 4b: see original paper]) exhibits numerous dimples, indicating substantial plastic deformation before fracture. No brittle fracture features are observed on the can fracture surface, confirming ductile fracture of the can occurring at the final stage of tensile loading. The core composite fracture can be divided into five distinct regions based on surface height. Regions 1, 2, 3, and 5 show relatively flat matrix fracture surfaces, while region 4 shows significant surface undulation.

Magnified observation of the flat region 3 ([Figure 4c: see original paper] and [Figure 4d: see original paper]) reveals a flat matrix fracture surface with patterns radiating outward along the precursor wire columnar grain growth direction, characteristic of brittle fracture. Fiber fracture surfaces also show clear brittle fracture features with diverse textures. Most fiber fracture planes are nearly coplanar with the surrounding matrix, with no obvious interface cracking. Only a few fibers show pull-out, with limited pull-out length. Using SEM focal depth measurements, the average fiber pull-out height is approximately 0.2 mm. Additionally, small amounts of residual carbon coating adhere to the inner walls of holes left by pulled-out fibers, indicating that fiber fracture initiates at the fiber core and propagates outward. When cracks reach the carbon coating, the graphite-like layered structure parallel to the fiber axis readily undergoes coating splitting, forming crack blunting or deflection, though this phenomenon is uncommon in the flat regions of the room-temperature fracture surface. Regions 1, 2, and 5 show fracture characteristics similar to region 3.

The undulating region 4 ([Figure 4e: see original paper] and [Figure 4f: see original paper]) shows significantly greater matrix surface roughness than region 3, with numerous dimples in the matrix, confirming ductile fracture after plastic deformation. This suggests region 4 fractured after regions 1, 2, 3, and 5. In region 4, fiber fracture planes are not coplanar with the surrounding matrix but vary in height following matrix surface undulations. Interface cracking is more pronounced than in region 3, though the number and length of pulled-out fibers do not increase significantly, indicating strong fiber/matrix interfacial bonding at room temperature.

2.3.2 Elevated-Temperature Fracture Morphology [Figure 5a: see original paper] shows the macroscopic fracture morphology of SiCf/TC17 composite at 773 K. The can fracture surface ([Figure 5b: see original paper]) remains unchanged from room temperature, showing dimples and ductile fracture. However, the core composite fracture morphology changes dramatically, with extensive matrix surface undulation, extreme irregularity, and no relatively flat regions or distinct fracture planes as observed at room temperature.

[Figure 5c: see original paper] presents a typical local view of the 773 K fracture surface. Fiber fracture planes are not coplanar with the surrounding matrix, and extensive fiber pull-out occurs, with an average pull-out length of approximately 0.5 mm. Both the number and length of pulled-out fibers exceed those at room temperature, indicating weaker interfacial bonding at 773 K. In addition to fiber pull-out, some tungsten cores also exhibit "pull-out" ([Figure 5d: see original paper]), suggesting decreased W/SiC interface bonding strength at elevated temperature. Furthermore, longitudinal cracks parallel to the fiber axis appear in the matrix between closely spaced fibers ([Figure 5e: see original paper]), promoting interface debonding that bypasses fibers and connects with other longitudinal cracks, forming an extended longitudinal crack plane. This results from excessive circumferential residual tensile stress in the thin matrix between fibers. Longitudinal cracks also appear in room-temperature fractures ([Figure 4d: see original paper]) but are less observable due to limited matrix plastic deformation and minimal interface debonding. According to literature [15], circumferential residual tensile stress is higher at room temperature than at elevated temperature, making longitudinal crack planes easier to form at room temperature. The multiple flat regions observed in room-temperature fractures result from crack planes being arrested by longitudinal crack planes during propagation. [Figure 5f: see original paper] confirms extensive matrix plastic deformation during elevated-temperature tensile testing, characteristic of the entire fracture surface, establishing ductile fracture at elevated temperature. Additionally, besides C coating/reaction layer interface debonding, W/SiC interface cracking occurs at elevated temperature, with the reaction layer adhering to the SiC side, confirming decreased W/SiC interface bonding strength at high temperature.

2.4 Fracture Mechanisms

2.4.1 Room-Temperature Fracture Mechanisms The longitudinal section morphology of flat regions 1, 2, 3, and 5 in the room-temperature fracture surface ([Figure 4a: see original paper]) is shown in [Figure 6: see original paper]. The composite fracture surface appears as a regular zigzag line ([Figure

6a: see original paper]), indicating crack deflection or convergence of two crack planes during propagation. The fiber damage zone extends within 300 μm of the fracture surface, with most fibers fracturing once and a few fracturing multiple times; no fiber fractures are observed beyond this region. Most fiber fracture planes are flat and nearly parallel to the fracture plane (perpendicular to the fiber axis), indicating axial tensile loading dominance. [Figure 6b: see original paper] shows a broken fiber near the fracture surface with reaction layer firmly attached to the matrix containing transverse cracks. Most reaction layer cracks do not cause matrix fracture, though a few near fiber fracture planes extend into the matrix, becoming crack sources for subsequent matrix fracture ([Figure 6c: see original paper]). [Figure 6d: see original paper] shows typical matrix morphology near the fracture surface, with a flat fracture surface showing no obvious deformation, indicating transgranular fracture. This demonstrates that matrix fracture occurs after surrounding fiber fracture, caused by reaction layer crack propagation into the matrix.

The longitudinal section morphology of the undulating region 4 ([Figure 4a: see original paper]) is shown in [Figure 7: see original paper]. The fracture surface is not on the same plane. The fiber damage zone length is significantly greater than in flat regions, with fiber fracture observable beyond 1 mm from the fracture surface. Most fibers exhibit multiple fractures; no fiber fractures are found in regions farther from the fracture surface. Most fiber fracture planes are flat and perpendicular to the fiber axis, though some are angled less than 90° to the axis with severe damage, numerous fiber fragments near the fracture, and multiple fractures of the tungsten core. Additionally, plastic deformation is evident in the matrix near fractured fibers ([Figure 7b: see original paper]), indicating that fracture of severely damaged fibers results not from axial overload causing rapid crack propagation from surface or internal defects, but from local compression by the matrix creating multiple interacting cracks. After removing SiC and W core fragments, numerous transverse cracks are visible in the reaction layer ([Figure 7c: see original paper]), which did not extend into the matrix. [Figure 7d: see original paper] shows typical matrix morphology near the fracture surface, with substantial plastic deformation and internal micro-cavities. This indicates that matrix fracture in this region occurs after surrounding fiber fracture, resulting from ductile fracture due to excessive axial load, with cracks initiating from internal micro-cavities rather than from reaction layer cracks near broken fibers.

Based on the degree of matrix deformation, the undulating region fractured after the flat regions, corresponding to the end of the second linear segment in the room-temperature stress-strain curve, while fracture of multiple flat regions corresponds to the beginning of this segment. The small fluctuations in the second linear segment likely result from specimen vibration when transverse crack planes are arrested by longitudinal crack planes.

2.4.2 Elevated-Temperature Fracture Mechanisms The longitudinal section morphology near the 773 K fracture surface is shown in [Figure 8:

see original paper]. The composite fracture surface is highly irregular, with obvious matrix plastic deformation and fiber pull-out ([Figure 8a: see original paper]). The fiber damage zone extends up to approximately 2 mm from the fracture surface, with fibers undergoing multiple fractures and more severe damage than at room temperature. Most fiber fracture planes are angled to the fiber cross-section, indicating shear stress during fracture. Most fibers exhibit pulverized fracture without causing multiple tungsten core fractures. Some tungsten cores also show pull-out ([Figure 8b: see original paper]), which comprehensively demonstrates fiber pulverized fracture, interface debonding, crack deflection, fiber sliding, matrix plastic deformation near interfaces, and "tungsten core pull-out" —all mechanisms that absorb crack tip energy and hinder crack propagation. These result from reduced radial residual compressive stress at the interface at elevated temperature, decreasing interfacial bonding strength. Reaction layer transverse cracks still appear near fiber fracture planes at elevated temperature ([Figure 8c: see original paper]), but with reduced crack density and increased spacing compared with room temperature, due to decreased interface shear strength at high temperature. Additionally, small amounts of residual carbon coating adhere to reaction layers near many fiber fracture planes, with these coatings fracturing multiple times together with the reaction layer, indicating that fiber fracture originated from internal crack initiation. When cracks reach the carbon coating, they blunt or deflect without extending into the matrix. [Figure 8d: see original paper] shows typical matrix morphology near the fracture surface, with numerous cavities indicating substantial plastic deformation. No brittle matrix fracture features are observed throughout the longitudinal section.

At elevated temperature, matrix yielding occurs prematurely due to: (1) reduced matrix yield strength, (2) residual axial tensile stress not fully released at the test temperature, and (3) decreased load transfer capability due to weakened interface bonding, increasing the load borne by the matrix. Consequently, the matrix yields before fibers fracture under axial load. Matrix deformation creates compressive loads on fiber surfaces that increase with deformation, causing fibers to fracture along the matrix necking direction (55° to the loading axis), forming multi-lamellar "pancake structures" ([Figure 9a: see original paper]) [16]. These fiber fractures typically initiate at the W/SiC interface, where reaction products such as W_2C and WSi_2 form brittle phases with a thin reaction layer (~120 nm) and low tensile strength [17]. After W/SiC interface reaction layer fracture, SiC fracture and W/reaction layer interface debonding occur. When SiC fragments in the "pancake structure" are removed, intact tungsten cores remain ([Figure 9b: see original paper]), indicating that "tungsten core pull-out" differs fundamentally from fiber pull-out—the latter involves entire fibers pulling out from the interface after fracture. Multiple tungsten core fractures without pull-out occur at room temperature due to strong interface bonding. Comparison shows that multiple fracture mechanisms at elevated temperature are similar to those in the undulating region at room temperature (matrix plastic deformation, interface cracking, multiple fiber fractures), but are more prevalent, pronounced, and decisive for the fracture damage process at elevated temperature. Although similar mechanisms appear in the room-temperature undulating region, they occur less frequently than in flat regions, making the flat region mechanisms dominant at room temperature.

Discussion

3.1 Room- and Elevated-Temperature Tensile Fracture Strength

The fracture strength of SiC fiber-reinforced titanium matrix composites is typically estimated using the rule of mixtures (ROM) [18], which assumes uniform fiber tensile strength and identical stress distribution on all fibers—clearly unrealistic and leading to large errors. In reality, fiber tensile strength exhibits dispersion, and stress distribution varies due to material structure and processing. The GLS and LLS models account for fiber strength dispersion and load redistribution after weak fiber fracture, providing more accurate fracture strength estimates for analyzing composite fracture mechanisms and processes.

3.1.1 GLS Model Curtin [19,20] developed the GLS model, which treats the composite as a series of continuous, relatively independent links, each with length L equal to twice the fiber slip length in the matrix. When fibers fracture in a weak link, load is shared by all unbroken fibers across the entire section. Random cumulative fiber damage triggers final fracture, while stress concentration from fiber fracture is insufficient to alter the randomness of cumulative damage and is therefore neglected. The composite fracture strength using the GLS model is calculated as [4,11,18,21]:

$$\sigma_{\text{comp}} = V_f \sigma_c \left(\frac{1}{m+1}\right)^{1/m} + (1 - V_f)\sigma_{my}$$

where σ_{comp} is composite fracture strength, V_f is the fiber volume fraction in the gauge section (33%), m is the Weibull modulus (8.0), σ_{my} is the matrix tensile stress at composite fracture, and σ_c is the critical fiber slip strength, expressed as [5,8,14]:

$$\sigma_c = \sigma_0 \left(\frac{L_0 \tau}{d_f \sigma_0}\right)^{1/(m+1)}$$

where σ_0 is the fiber Weibull characteristic strength, τ is the interface slip stress, L_0 is the composite tensile specimen gauge length (22.5 mm), and d_f is the fiber diameter (103 μm).

Based on the material processing and fiber-matrix interface structure, and referencing literature [22-24], τ values of 138 MPa and 78 MPa were used for room temperature and 773 K, respectively. The σ_{my} value must account for axial residual tensile stress in the matrix. At room temperature, the SiCf/TC17 composite fracture strain is 0.91%, meaning the matrix experiences 0.91% strain from external loading at composite fracture. From the matrix stress-strain curve at room temperature ([Figure 3: see original paper]), this strain corresponds to 951.3 MPa. Adding the inherent axial residual tensile stress of 275.6 MPa [15] gives a matrix tensile stress of 1226.9 MPa at composite fracture. Similarly, σ_{my} at 773 K is 735.6 MPa. These calculated matrix stresses significantly exceed the tensile fracture strengths of monolithic TC17 alloy, likely due to fine grain strengthening from the matrix grain size (<10 μm). At room temperature, fibers fracture after reaching their tensile strength, so σ_0 is 3611.4 MPa. However, at 773 K, fiber fracture under compressive matrix loading does not reach the fiber tensile strength, so σ_0 must account for matrix yielding. At 773 K, composite yield and fracture strains are 0.57% and 0.79%, respectively, with fiber fracture strain between these values; an average of 0.68% was selected. Since room-temperature fiber fracture strain is 0.91%, the σ_0 value at 773 K is 3611.4 MPa$\times$0.68%/0.91% = 2698.6 MPa. The GLS model estimates for room- and elevated-temperature fracture strengths are listed in .

3.1.2 LLS Model The LLS model assumes that fracture of a weak fiber creates stress concentration on neighboring intact fibers, increasing their fracture probability. When a critical number of neighboring broken fibers form a cluster, composite fracture occurs. The LLS model expression for composite fracture strength when fracture is caused by i neighboring broken fibers forming a critical cluster is [14,18]:

$$\sigma_{\text{comp}} = V_f \sigma_c \left(\frac{1}{i+1}\right)^{1/m} \left(\frac{C}{L_0}\right)^{1/m} + (1-V_f)\sigma_{my}$$

where i is the number of neighboring broken fibers required to form a critical cluster, N is the number of fibers in the gauge section (575), and C is expressed as [14]:

$$C = \frac{C_j}{(i+1)!} \left(\frac{\sigma_c}{\tau}\right)^i$$

where C_j is a length constant, with C_1 taken as the gauge length (22.5 mm). Expressions for C_2, C_3, and C_4 are [14]:

$$C_2 = \frac{1}{2} \int_0^{d_c} \left(\frac{x}{d_c}\right)^{m+1} dx$$

$$C_3 = \frac{1}{3} \int_0^{d_c} \left(\frac{x}{d_c}\right)^{2(m+1)} dx$$

$$C_4 = \frac{1}{4}\int_0^{d_c}\left[\left(\frac{x}{d_c}\right)^{3(m+1)} + 3k_1\left(\frac{x}{d_c}\right)^{2(m+1)}\right]dx$$

where k_1 is the stress concentration factor (4% [14,21]) and d_c is the critical fiber slip length, expressed as [5,8,14]:

$$d_c = \frac{d_f\sigma_c}{2\tau}$$

In LLS model calculations, V_f, σ_0, L_0, m, σ_{my}, and τ values are identical to those used in the GLS model. Using equations (3)-(8), composite fracture strengths were calculated for critical clusters formed by 1-4 neighboring broken fibers, with results shown in . At room temperature, the GLS model overestimates composite fracture strength, with results clearly inconsistent with experimental values. In contrast, LLS model calculations for $i = 3$ and $i = 4$ agree well with experimental values, with deviations of only -1% and 2%, respectively, while $i = 1$ and $i = 2$ show larger deviations. This indicates that room-temperature damage in SiCf/TC17 composites follows LLS fracture mode, where fracture of a few weak fibers creates significant stress concentration on surrounding matrix and fibers, leading to neighboring fiber fracture. When three or more neighboring broken fibers form a critical cluster, rapid crack propagation occurs. González et al. [14] and Li et al. [21] also found good agreement between LLS model predictions and experimental tensile strengths for SiCf/Ti-6Al-4V composites at room temperature. At 773 K, GLS model calculations are significantly lower than experimental values, while all four LLS calculations are substantially below experimental values. This demonstrates that elevated-temperature fracture follows GLS mode, where interface debonding and crack deflection after weak fiber fracture prevent stress concentration on neighboring fibers, resulting in random fiber fracture. The fracture surface and longitudinal section morphologies at room and elevated temperatures correspond well to LLS and GLS mechanisms, respectively, validating the accuracy of these models for predicting fracture strength and analyzing fracture modes in SiCf/TC17 composites.

3.2 Fracture Processes

3.2.1 Room-Temperature Fracture Process In SiC fiber-reinforced titanium matrix composites with carbon coatings, the interfacial reaction layer consists primarily of brittle TiC phase with numerous defects and tensile strength far lower than that of fibers and matrix. Under tensile loading, the reaction layer is most prone to crack initiation, creating stress concentration at crack tips [25]. However, the high-strength SiC fiber inside prevents crack propagation through the fiber, and the ductile TC17 matrix outside can accommodate reaction layer cracks through slight yielding. Therefore, during initial loading, reaction layer cracks cannot penetrate SiC fibers or extend into the matrix, only causing limited C coating/reaction layer interface debonding or multiple reaction layer fractures. As loading increases and the stress-strain curve enters the second linear segment, random fracture of SiC fibers with surface or internal defects occurs due to fiber strength dispersion and reaction layer microcracks. SiC fiber cracks blunt or deflect limited distances upon reaching the interface, creating stress concentration in surrounding matrix and neighboring fibers that causes fracture of adjacent SiC fibers in the same cross-section. When the number of neighboring broken fibers reaches three, these fiber cracks extend into the matrix via reaction layer crack planes, coalescing into a larger crack plane that propagates rapidly. If this crack encounters longitudinal crack planes or reaches the can/composite interface, it deflects or arrests until other critical fiber cluster crack planes extend and intersect, forming a larger flat crack plane or multiple flat crack planes at different sections. The specimen then enters the rapid fracture stage (late second linear segment of the stress-strain curve), where the remaining unbroken sections cannot sustain the released load, causing matrix yielding and fracture of remaining fibers under combined axial tension and radial compression, followed by instantaneous fracture of remaining matrix and can, completing specimen failure. The main room-temperature tensile fracture process is illustrated in [Figure 10: see original paper].

3.2.2 Elevated-Temperature Fracture Process [Figure 11: see original paper] schematically shows the elevated-temperature tensile fracture process. Compared with room temperature, the matrix yields prematurely after loading begins. The reaction layer fractures randomly first due to matrix deformation, causing surrounding C coating/reaction layer interface debonding. Reaction layer cracks deflect along the interface for some distance before load transfers back to the reaction layer, causing secondary or multiple fractures. This process is similar to that at room temperature, but the critical length for load transfer back to the reaction layer increases at elevated temperature due to easier interface debonding, resulting in larger spacing between multiple reaction layer cracks. As loading increases and matrix deformation grows, compressive loads develop on SiC fiber surfaces, causing crack initiation at the internal W/SiC interface that propagates rapidly, leading to random pulverized fiber fracture and "pancake structure" formation. Cracks generated within SiC fibers can cause carbon coating splitting if they enter the coating, or interface debonding if they reach the C coating/reaction layer interface; both scenarios deflect cracks and prevent extension into the matrix. However, interface debonding eliminates load transfer, increasing tensile load on matrix near fiber fracture planes. Particularly when a cluster of neighboring fibers fractures near the same cross-section, the dispersed matrix between broken fibers bears the load entirely, forming micro-cavities in overloaded regions. When the number of broken fibers reaches a critical level, massive plastic deformation and microvoid coalescence fracture occur in matrix and can near the cross-section where fiber fracture is relatively concentrated. The final fracture surface connects with various fiber fracture planes, completing specimen failure. In fact, the final stage of elevated-temperature fracture closely resembles that at room temperature, differing only in that the same mechanisms are more intense and pronounced at elevated temperature.

Conclusions

1. The room-temperature tensile stress-strain curve of SiCf/TC17 composites shows typical bilinear shape, while the elevated-temperature curve exhibits near-linear shape with some curvature. The primary factors determining these characteristics are fiber elastic deformation and matrix yielding.
2. The main room-temperature fracture mechanisms include: multiple reaction layer fractures, single fiber fracture, fiber crack extension into matrix, and matrix transgranular fracture. The main elevated-temperature fracture mechanisms include: interface debonding, crack deflection, multiple reaction layer cracks, carbon coating splitting, multiple fiber fractures, fiber pull-out, tungsten core "pull-out," matrix plastic deformation, and matrix micro-cavities.
3. The room-temperature tensile damage mode in SiCf/TC17 composites follows the LLS model, where fracture occurs after three or more neighboring broken fibers form a critical cluster. The elevated-temperature tensile damage mode follows the GLS model, where random and multiple fiber fractures lead to material failure.
4. The room-temperature fracture process proceeds as: multiple reaction layer fractures → random fracture of a few weak fibers → critical fiber cluster formation → crack plane development in broken fiber clusters and nearby matrix → crack plane extension and connection with other crack planes → formation of larger crack plane → rapid fracture of remaining fibers, matrix, and can. The elevated-temperature fracture process proceeds as: matrix yielding → random multiple fiber fractures → multiple reaction layer fractures → fiber-matrix interface debonding → matrix micro-cavity formation → rapid fracture of remaining fibers, matrix, and can.

References

[1] Christoph L, Frank K, Joachim H, Wolfgang A K. Aerosp Sci Technol, 2003; 7: 201
[2] Nicolas C, Frédéric F, Serge K. Aerosp Sci Technol, 2003; 7: 307
[3] Naseem K, Yang Y Q, Luo X, Huang B, Feng G H. Mater Sci Eng, 2011; A528: 4507
[4] Thomas M P, Winstone M R. Compos Sci Technol, 1999; 59: 297
[5] Gundel D B, Wawner F E. Compos Sci Technol, 1997; 57: 471
[6] Fukushima A, Fujiwara C, Kagawa Y, Masuda C. Mater Sci Eng, 2000; A276: 243

[7] Weber C H, Chen X, Connell S J, Zok F W. Acta Metall Mater, 1994; 42: 3443
[8] Gálvez F, González C, Poza P, Llorca J. Scr Mater, 2001; 44: 2667
[9] Kagawa Y, Fujita T, Okura A. Acta Metall Mater, 1994; 42: 3019
[10] Peters P W M, Hemptenmacher J. Composites, 2002; 33A: 1373
[11] Truon A, Ccota J, Mainí P, Trias D, Mayugo J A. Compos Sci Technol, 2005; 65: 2039
[12] Baik K H, Grant P S. Scr Mater, 2001; 4: 607
[13] Yang Y Q, Zhu Y, Ma Z J, Chen Y. Scr Mater, 2004; 5: 385
[14] González C, Llorca J. Acta Mater, 2001; 49: 3505
[15] Zhang X. PhD Dissertation, Institute of Metal Research, Chinese Academy of Sciences, Shenyang, 2012
[16] Garcíia-Leiva M C, Ocaña I, Martíin-Meizoso A, Martíinez-Esnaola J M, Marqués V, Heredero F. Eng Fract Mech, 2003; 70: 2137
[17] Cheng T T, Jones I P, Shatwell R A, Doorbar P. Mater Sci Eng, 1999; A260: 139
[18] Hill R A. J Mech Phys Solids, 1965; 13: 213
[19] Curtin W A. J Am Ceram Soc, 1991; 74: 283
[20] Curtin W A. Composites, 1993; 24: 98
[21] Li J K, Yang Y Q, Yuan M N, Luo X, Li L L. Trans Nonferrous Met Soc China, 2008; 18: 523
[22] Chandra N, Ghonem H. Composites, 2001; 32A: 575
[23] Chandra N, Ananth C R. Compos Sci Technol, 1995; 54: 87
[24] Kalton A F, Howard S J, Janczak-rusch J, Clyne T W. Acta Mater, 1998; 46: 3175
[25] Jeng S M, Yang J M, Yang C J. Mater Sci Eng, 1991; A138: 169

Note: Figure translations are in progress. See original paper for figures.

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