Compressive behavior of C/SiC composites over a wide range of strain rates and temperatures

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ABSTRACT

The mechanical behavior of two-dimensional (2D) carbon fiber reinforced silicon carbide (C/SiC) composites is investigated at both quasi-static and dynamic uniaxial compression under temperatures ranging from 293 to 1273 K. Experimental results show that temperature and strain rate dramatically affect the compressive behavior of 2D C/SiC composites. If the temperature is below 873 K, the compressive strength increases with rising temperature. The reason is that the release of thermal residual stress enhances the compressive strength and this enhancement is more significant than the strength degradation due to the high temperature induced oxidation. In contrast, when the temperature rises above 873 K, the compressive strength decreases as temperature rises due to the stronger effect of oxidation induced strength degradation. Moreover, the degradation of compressive strength at strain rate of 10^-4/s and temperatures above 873 K is much more obvious than those at higher strain rates, and the strain rate sensitivity factor of compressive strength increases remarkably at temperature above 873 K. Post-deformation observation shows that failure angles and fracture surfaces are also strongly dependent on testing temperature and strain rate. The change of interfacial strength at high strain rate or high temperature is responsible for the variations.

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1. Introduction

Due to their high strength and stiffness, low density and excellent high temperature sustainability, continuous carbon fiber reinforced silicon carbide (C/SiC) composites have been widely used in high temperature structures. When C/SiC composites are used as heat shields for space shuttles or hypersonic aircrafts, a major problem is that they are frequently subjected to severe circumstances such as the impact of space debris, in some cases even accompanied by high temperature [1,2]. For example, the loss of the space shuttle Columbia in 2003 was caused by the impact of insulation foam on the composite leading edge [2]. Therefore, a full understanding of their mechanical properties under coupling conditions of elevated temperatures and high strain rates is important to the safety and reliability of composite structures.

In recent years, investigations have been conducted on high temperature properties of C/SiC composites. Cheng [3-5] studied the oxidative behavior of C/SiC composites at high temperatures and found the mass variation is strongly dependent on temperature. Zhang [6] compared the tensile behavior of two-dimensional (2D) composites in simulated re-entry environment as well as in vacuum. The results showed a significant reduction of strength when the composites were tested in re-entry environment. The strength reduction was attributed to the combined effects of the applied load and high temperature oxidation. The out-of-plane
compressive strength of 2D C/SiC composite in air was studied by Niu et al. [7] at room temperature, 973, 1373 and 1573 K, respectively under quasi-static compressive loading. The experimental results showed the 2D C/SiC composites displayed a gradual strength increase at temperatures below 973 K. At temperatures above 973 K, stress-oxidation played a more important role in the response of compressive strength in the air and led to the decreased compressive strength.

More recently, studies have been focused on the dynamic properties of the C/SiC composites. Li [8,9] studied the compressive behavior of 2D and three-dimensional (3D) C/SiC composites at strain rates ranging from $10^{-4}$ to $10^{3}$/s. The results indicated that at room temperature the strain rate had an insignificant influence on the mechanical properties of the material. Tensile experiments at high strain rates on 2D C/SiC composites showed that the strain rate sensitivity of in-bundle interface is mainly responsible for the strain rate sensitivity of the tensile strength [10].

Although high temperature properties of C/SiC composites have been widely investigated, limited literature was focused on the stress/strain responses under the coupling conditions of strain rate and temperature, which are essential for the impact resistance design of C/SiC thermal structures. In this paper, mechanical behavior of 2D C/SiC composites is investigated under both quasi-static and dynamic loading conditions along with elevated temperatures. The temperature and strain rate dependencies of compressive strength are investigated and the post-deformation failure is studied using optical microscope and scanning electron microscope (SEM).

2. Experimental procedure

2.1. Preparation of composite

The 2D C/SiC composite specimens were provided by the State Key Laboratory of Solidification Processing at Northwestern Polytechnical University, China. T300-1K fiber was the reinforcing phase and the volume fraction was about 40%. First, the fiber preform was prepared by 2D braided method into carbon cloth; then plain carbon clothes were stacked in parallel and SiC matrix was deposited by chemical vapor infiltration method at 1273 K. To improve the anti-oxidation capability at high temperatures, after the specimens were machined to final geometry, SiC coating with a thickness of 30–50 µm was deposited on the surface of specimens at 1273 K using chemical vapor deposition [6]. The specimen was cylindrical with 9 mm in diameter and 5 mm in thickness. The compressive loading direction was along with the thickness direction of the C/SiC plate.

2.2. Experimental procedure

2.2.1. Quasi-static experiments

The quasi-static experiments were conducted using a universal testing machine with the maximum load capacity of 25 KN. The specimens were heated to the testing temperatures using an electric furnace with a temperature control system at a heating rate of 30 K/min. Due to the low heat transfer capability of C/SiC composites, after the specimens were heated to experimental temperature, they were held for additional 20 min to achieve homogenous temperature distribution. During quasi-static experiments, the strain rates were controlled at $10^{-4}$ and $10^{3}$/s, and the temperatures were controlled at 293, 573, 873, 1073 and 1273 K, respectively.

2.2.2. Dynamic experiments

The high strain rate experiments were performed using the split Hopkinson pressure bar (SHPB), as shown in Fig. 1. The primary components of the classic SHPB system are two long elastic bars: the incident and transmission bars. A short specimen with a length to diameter ratio of below 1 is sandwiched between the two bars. When a projectile launched from a gas gun impacts the free end of the incident bar, a compressive pulse is generated and travels along the incident bar towards the specimen. Due to the mismatch of wave impedance between the specimen and loading bars, while reaching the bar/specimen interface, part of the stress pulse is reflected, and part of it is transmitted into the transmission bar. Assuming homogeneous deformation and uniform stress state in the sample, the strain, stress and strain rate can be calculated as [12]:

$$
\sigma_S = E \left( \frac{A_S}{A_C} \right) e_S = \frac{2C_0}{t_S} \int_0^t \ddot{e}_S dt \quad \dot{e}_S = \frac{2C_0}{t_S} \ddot{e}_S
$$

where $e_S$ and $c_S$ are the transmitted and reflected strain waves which can be measured by the strain gages on the loading bars; $C_0$, $E$ and $A_c$ denote the longitudinal elastic wave velocity, Young’s modulus and cross-sectional area of the loading bars, respectively; $l_S$ and $A_S$ are the length and cross-sectional area of the specimen. The classic SHPB technique provides a way to measure the stress-strain responses of materials at room temperature in the strain rate range of $10^2$–$10^3$/s.

Fig. 1 – High-temperature SHPB set-up with a synchronically assembled heating system [11] 1 inlet valve; 2 magnetic outlet valve; 3 main air chamber; 4 air chamber; 5 strike; 6 incident bar; 7 transmission bar; 8 plunger; 9 heating furnace; 10 air pipe; 11 electrical wire; 12 time relay; 13 switch; 14 sample; 15 sleeve; 16 thermocouple.
In classic SHPB experiments, the incident and transmission bars must be in close touch with the specimen. In case of high temperature experiments, if the loading bars are heated with the specimen, a strong temperature gradient within the bars will be generated, which will cause the gradient of both the elastic modulus and longitudinal elastic wave velocity within the loading bars; therefore, adjustment is necessary during the post-processing of the experimental data. However, such technique has limitations since the temperature gradient is hard to be directly measured or estimated. At the same time, thermal cycling associated with repeated testing at high temperature may cause permanent altering of the mechanical properties of Hopkinson bars. To overcome this limitation and improve the high temperature experimental capability of SHPB, a synchronically assembled heating system (see Fig. 1) was designed to be incorporated into the conventional SHPB apparatus. This technique enables us to preheat the specimen with the incident and transmission bars kept away from the heating zone of the furnace. After the desired specimen temperature has been achieved, the bars are brought into close contact with the specimen automatically (for more details, see Ref. [11]).

One thing worth being noticed is that during the high temperature and high strain rate coupling experiment using the synchronically assembled heating system mentioned above, the incident stress wave arrives at the incident bar-specimen interface shortly after the incident and transmitted bars are brought into close contact with the specimen. This means a time delay, named cold contact time, exists before the specimen is loaded but after the "hot" specimen contacts with the "cold" loading bars. Since during the cold contact time heat can be conducted away from the "hot" specimen into the "cold" loading bars, it should therefore be minimized. For our experimental setup, the cold contact time was directly measured using a sample electrical circuit (see more detail in Ref. [11]). The results show the maximum cold contact time is within 5 ms. Even in such short cold contact procedure, heat lose may cause temperature decrease in the specimen. To evaluate the influence of cold contact time on specimen temperature, numerical simulation was employed with the ABAQUS/Explicit code. An axisymmetric finite element model is employed since both the C/SiC specimen and loading bars are cylindrical. Fig. 2 shows the FEM model of heat transfer simulation. Here, the dimensions of the model are the same as the experiment. The SiC anti-oxidation coating was also included in the simulation. The available 4-node linear axisymmetric heat transfer quadrilateral elements are used.

The variation of thermal conductivity coefficients of SiC and C/SiC with temperature used in FEM simulation is also shown in Fig. 3. One thing worth mentioning is that in the FEM simulation of heat transfer, to simplify the problem, the heat conductivity property of C/SiC composite is assumed to be homogenous, and thermal conductivity parameters in the out-plane direction of the C/SiC composite specimen (same as axial direction) were employed. In FEM simulation, the initial temperature of loading bars was set as 293 K. For the specimen, the initial temperature was set as 573 and 1273 K corresponding to the minimum and maximum experimental temperatures when the synchronically assembled heating system was employed. One thing should be noticed that due to limited data, it was assumed that the thermal contact conductance between the specimen/bar interfaces was constant in simulation.

The temperature distributions along the axis of the specimen at different cold contact time and initial temperatures are shown in Fig. 4, in which the horizontal axis represents the position away from the center of specimen. The positions 2.0 and −2.0 correspond to the specimen/bar interfaces.

Clearly, the temperature distribution within the specimen is not uniform. The temperature for the position closer to the center of the specimen remains to be the initial temperature or non-significant drop. The farther away to the center of the specimen, the lower local temperature is. At the same time, there is a significant drop in temperature within the specimen as cold contact time increases. However, a slight temperature gradient forms within the local region of the loading bars closer to the specimen/bar interface. The longer the cold contact time is, the higher local temperature achieves within the loading bars. Since the cold contact time for our experiments does not exceed 10 ms, the temperature drop within more than 90% of the specimen does not exceed 20 and 50 K at the initial temperature of 573 and 1273 K, respectively. Thus the temperature drop due to the "hot" specimen contact with the "cold" loading bars before the specimen is loaded is negligible since it is less than 5% of initial temperature almost within the whole specimen. Meanwhile, the highest temperature within the loading bars does not exceed 335 K at the initial temperature of 1273 K within a 5 ms cold contact time. The influence of such a non-significant temperature gradient for the elastic modulus of martensitic steel bars is also limited. Thus, it is reasonable to
ignore the thermal diffusion effects between the “hot” specimen and “cold” loading bars. By using this technique, the dynamic mechanical responses of the 2D C/SiC composites at 293, 573, 873, 1073 and 1273 K can be determined.

For dynamic experiments at room temperature (293 K), a steel jacket was used to limit the compressive strain in order to prevent specimens from being loaded for the second time (see Fig. 1 in Ref. [13]). However, due to the difficulty in incorporating the steel jacket into the synchronically assembled system, the strains of the specimens tested at high temperatures were not controlled in this work. As a result, for high strain rate experiments, only fracture surfaces at room temperature are presented.

In SHPB experiments, to ensure the validity of experimental data, the stress-wave loading should be designed such that the specimen deforms uniformly under a dynamically equilibrated stress state as well as at a constant strain rate. However, in most cases, for brittle materials, an equilibrated stress state or constant strain rate is difficult to be achieved by using the trapezoidal incident pulse. The brittle specimen may start to fail from impact end at a very early stage of loading before dynamic equilibrium is achieved. Moreover, at the plateau of the trapezoidal loading pulse, the specimen may deform at drastically decreased strain rates, which fails to meet the assumption of constant strain rate deformation [14]. To satisfy these experimental conditions, a thin copper film was used as the pulse shaper to achieve a ramp incident pulse.

### 2.3. Microscope characterization

The failure patterns were observed using optical microscope and SEM (JSM6460). Since experiments at high temperatures and strain rates were carried out using SHPB with synchronically assembled system without steel jacket and the specimens were over-loaded and broken into small pieces, which were spread in the electronic furnace, it is hard to recover the tested specimens. This makes it very hard to compare the failure patterns and corresponding fracture surfaces with those deformed under single load. Thus, in case of dynamic
loading condition, both optical and SEM observations were carried out only on specimens tested at 293 K. Meanwhile, since the specimens subjected to quasi-static compression at high temperatures underwent severe oxidization while being naturally cooled down to room temperature together with the furnace, for specimens tested quasi-statically at higher temperatures the specimens were only subjected to optical observation and the fracture surfaces were not examined.

3. Results and discussions

3.1. Stress equilibrium and constant strain rate

To validate the equilibrated stress state, a stress equilibrium coefficient is defined as:

$$\alpha(t) = \frac{[\sigma_1(t) - \sigma_2(t)]}{[\sigma_1(t) + \sigma_2(t)]/2}$$

where $\sigma_1(t)$ and $\sigma_2(t)$ are stresses on the incident bar/specimen and transmitted bar/specimen interfaces respectively. The stress equilibrium coefficient is shown in Fig. 5(a). It shows that a dynamic equilibrium deformation state, corresponding to the stress of 60 MPa, is achieved after the first 7 $\mu$s initial loading stage, and the stress equilibrium coefficient was less than 3%. A nearly constant strain rate is also achieved with the strain ranging from 0.7% to 3.5%, as shown in Fig. 5(b). Within this deformation stage, the stress increases from 140 MPa to its peak, 353 MPa, and then drops off rapidly. Considering both dynamically equilibrated stress state and constant strain rate conditions, the experimental data are valid within the strain ranging from 0.7% to 3.5% which covers the major deformation stages of the composites.

3.2. Stress vs. strain curves

Fig. 6(a) shows two representative quasi-static stress–strain curves at a strain rate of 0.0001/s at 293 and 1273 K respectively. Here, for better observation, the curve at 1273 K was plotted with a 2% strain offset. Both curves can be roughly divided into four deformation phases according to the variation of modulus. In the initial phase (phase I), the modulus increases gradually with strain, mainly due to the closing of such defects as micro-vacancies or micro-cracks formed during the processing of composites. In phase II, the modulus keeps a constant value and the stress increases nearly linearly with strain. Then the modulus decreases gradually until the stress reaches a peak value. This deformation stage can be classified as phase III, which is characterized by an obvious nonlinear stress–strain relationship. This feature can be attributed to energy-absorbing processes such as the evolution of micro-cracks, the fiber/matrix interfacial failure and fracture/pull-out of carbon fibers [10]. With the accumulation and coalescence of micro-cracks, macro-cracks are formed at the end of this phase. Then it is phase IV, with the propagation of macro-cracks resulting in dramatic reduction in the load-bearing capacity of the material system. Different from traditional brittle materials, such as ceramic and rocks, catastrophic brittle failure is not observed. It is reasonable that the fracture of the SiC matrix may cause a drop in the loading capacity of the material after the peak stress [13]. Meanwhile, additional energy is needed for the pullout or failure of the carbon fibers, which naturally leads to a higher failure strain of the composites. One thing worth mentioning is that the quasi-static stress–strain curve at 293 K was cutoff artificially to avoid confusion with that at 1273 K when Fig. 6(a) was plotting. While in dynamic experiments, as a steel jacket was used to limit the maximum strain to a certain level when the C/SiC composites was tested at 293 K, the final strain seems to be smaller than that at 1273 K.

Similar features can also be observed at higher temperatures. Fig. 6(b) shows the stress–strain curves at two different temperatures under dynamic loading condition. Here strain offsets were also made and each curve was divided into four phases, similar to the above. It should be pointed out that at high strain rates those data in phase I are less reliable due to the non-equilibrium stress state between two sides of specimen at the initial stage of loading.
3.3. Strain rate effect

Fig. 7(a) plots the compressive stress–strain curves at room temperature under both quasi-static and dynamic loading conditions. Unlike traditional metals and ceramics that usually show linear elastic behavior before yield or failure, the C/SiC composites displays a nonlinear stress strain relation. Such behavior is an indication of subcritical crack growth. Similar behavior was also observed by other researchers in fiber reinforced composite materials such as C/SiC, C/C and natural fiber reinforced composites [8,10,14–21]. The evolution of pre-existing defects, such as micro-voids, micro-cracks, was believed to result in the reduction in stiffness as strain increases. Meanwhile, it is clear that the compressive strength of C/SiC composites increases with the applied strain rate. For example, the compressive strength of the material increases from about 250 MPa at the strain rate of $10^{-3}$/s to 330 MPa at 1500/s with a relative increment of about 32%.

The strain rate sensitivity of the material may be caused by two factors. On the one hand, SiC matrix is strain rate dependent [22]; on the other hand, the pre-existing micro-cracks caused by the thermal expansion coefficients mismatch between fiber and matrix also contributes to the rate sensitivity based upon the crack growth dynamics [8,23]. Under compressive loading, the nucleation and propagation of micro-cracks cause the failure of the composites. It should be pointed out that both the nucleation and propagation of micro-cracks are loading rate dependent [24]. Under quasi-static loading, the micro-cracks have enough time to nucleate and extend; while in case of dynamic loading, the applied load pulse is usually tens of microseconds, which tends to increase the rate of the micro-crack nucleation and extension, and thus higher compressive strength is achieved [25]. Moreover, higher strain rate also leads to more severe stress concentration near the tips of micro-cracks, which may facilitate the damage evolution. Thus, strain corresponding to compressive strength of the composites decreases with increasing strain rate.

Fig. 7(b) shows the corresponding compressive strength vs. logarithmic strain rate curve. Although there is scatter associated with the experimental data (nature of the brittle material systems due to the inherent distribution of defects in the material system), close inspection of the experimental results
reveals that the compressive strength of the material varies almost linearly with the logarithm of strain rate. A similar phenomenon is also observed in 3D needle-punched C/SiC composites [9].

To quantify the strain rate effect on the compressive strength of C/SiC composites, the sensitivity factor of compressive strength to strain rate is defined as follows:

$$m = \frac{\partial \log \sigma_b}{\partial \log \dot{\varepsilon}}$$

(3)

where $\sigma_b$ and $\dot{\varepsilon}$ are the compressive strength and strain rate respectively. Fig. 8 gives the strain rate sensitivity factor of compressive strength at different temperatures. It is evident that the variation of $m$ is insignificant at temperatures below 873 K. However, at temperatures above 873 K, $m$ increases dramatically with rising temperature.

### 3.4 Temperature effect

Fig. 9 presents the compressive strength vs. temperature curves at different strain rates. Unlike the traditional materials whose strengths decrease monotonously with increasing temperature, the C/SiC composites exhibit a critical compressive strength at temperature of about 873 K and then the compressive strength decreases at higher temperatures. It should be noted that due to the interval of experimental temperature, the C/SiC composites exhibit a critical compressive strength at temperature of about 873 K and then the compressive strength decreases at higher temperatures. It should be pointed out that the reason that the compressive strength is achieved might lie between 873 and 1073 K. Since the intrinsic strengths of both silicon carbide ceramic matrix and carbon fiber can be maintained without any degradation or even slightly increase at very high temperature up to more than 1273 K [26], the variation of compressive strength at different temperatures shown in Fig. 9 is mainly caused by the high temperature oxidation as well as the thermal residual stress (TRS), which induced the alteration of interfacial bonding strength between fiber and matrix [4,6,7,26–28]. It should be pointed out that the reason that the compressive strength at 1273 K in case of strain rate 0.0001 seems is not reduced as remarkably as those at higher strain rates might be due to the large scattering of the C/SiC composites.

Since the composites are fabricated at high temperature, i.e. typically 1273 K of the chemical vapor infiltration (CVI) processing of C/SiC composites, TRS are often generated in the composites upon cooling from processing to room temperatures due to extensive mismatch of the coefficients of thermal expansion between C fiber and SiC matrix. Such TRS is known to play an important role in the mechanical behavior of C/SiC composites as well as in damage (matrix cracking, fiber–matrix failure). Consider a composite with ideal interface bond and without any matrix cracks associated with relief of the TRS from processing to room temperature, the axial residual stress in a non-cracked matrix then can be classically expressed by [29]:

$$\sigma_{\text{TRS}}(T) = E_f E_m V_f \left( \frac{x_f - x_m}{E_f + E_m} \right)$$

(4)

where $\sigma_{\text{TRS}}$ refers to axial thermal residual stress in matrix, $E$ the elastic modulus, $V_f$ the volume fraction of fiber, $x$ the thermal expansion coefficient, $T_e$ the experimental temperature and $T_p$ is the preparation temperature. The subscripts ‘f’ and ‘m’ present the fiber and matrix respectively. For the C fiber, the thermal expansion coefficient in axial direction is $-0.1–1.1 \times 10^{-6}/K$ while for the SiC matrix, the thermal expansion coefficient is $4.6 \times 10^{-6}/K$. That means in the present C/SiC, the SiC matrix is in residual tensile stress which leads to opening micro-cracks in the materials once cooled down from the processing temperature to room temperature [29]. However, it should be pointed out that although the TRS induced cracks can lead to release of TRS, the formation of local crack can only reduce the local TRS to a value below the critical stress for crack propagation. On the other hand, in the region where TRS is below the critical crack initiation stress, there will be no TRS induced cracks. That is, the TRS cannot be fully released through cracking. Therefore, it is believed that a certain amount of TRS, which is lower than the critical levels to initiate the local cracking, is retained in the material system even though some TRS induced cracks exists.

Except for the TRS in axial direction of carbon fiber, however, due to higher thermal expansion coefficient in radial direction of carbon fiber $1.0 \times 10^{-6}/K$ than SiC matrix, tensile TRS is generated within both the matrix and the fiber in the
radial direction of fiber when the composites are cooled down from its preparation temperature (1273 K) to room temperature [6]. Since the existence of radial TRS increase or decrease the fiber–matrix interfacial strength depending on whether they are compressive or tensile stresses [6,30,31], it is reasonable to believe that the generation of thermal residual tensile stress in radial direction tends to decrease the fiber–matrix interfacial strength of the composites, thus weaken the compressive strength of the composites. If the testing temperature is between room and preparation temperature, the amount of TRS generation decreases with increasing testing temperature, thus the material tends to have higher compressive strength as the testing temperature getting close to the preparation temperature (1273 K).

Besides the TRS, oxidation embrittlement is another continuous problem that plagues the C/SiC composites, which involves the ingress of oxygen through matrix cracks and the subsequent reactions of oxygen with matrix and fibers [32]. When the composite specimens are tested at high temperatures in the air, the oxidative reactions of carbon fiber and silicon carbide matrix can be written as:

\[
2C(S) + O_2(g) \rightarrow 2CO(g)
\]  
\[
2SiC(S) + 3O_2(g) \rightarrow 2SiO_2(g) + 2CO(g)
\]

The occurrence of such reactions inside the composites requires micro-cracks acting as channels for oxygen transfer. Fig. 10 shows thermal gravimetric analysis (TGA) curves of carbon fiber and SiC matrix respectively. It can be seen that for carbon fiber slight weight change occurs at ~723 K. At temperature above ~873 K TGA curve drops sharply which indicates remarkable oxidation happens. Similarly, serve oxidation of SiC only starts at temperature above ~920 K. Meanwhile, it should also be pointed out that although a layer of anti-oxidation SiC coating was deposited on the surface of the composite specimens, the surface morphology shows pre-existing micro-cracks due to the TRS (see Fig. 11). At the early stage of deformation, the micro-cracks in surface coating as well as in matrix act as channels for oxygen transferring. With the increase of applied load, pre-existing cracks open and new cracks nucleate and propagate, which enhance the oxidative damage and thus accelerate material failure.

Since TRS and oxidation play competitive roles in the mechanical response of C/SiC composites, the competition between them determines the trend of compressive strength at elevated temperatures. On the one hand, the temperature increase releases the TRS and thus strengthens the composites. On the other hand, in case of temperatures above 723 K, oxidation damage tends to weaken the material. However, although the oxidative reaction of Eq. (5) starts from 723 K, it only becomes pronounced at temperatures above 873 K [33,35]. Therefore, it is reasonable to believe that the oxidative damage induced degradation of compressive strength is limited at temperatures below 873 K and thus can be overwhelmed by the strengthening effect due to the release of TRS. The compressive strength increases with temperature until a certain maximum value is reached at temperature \( T_p \) (873 K < \( T_p < 1073 \) K) where the oxidation induced degradation is comparable with the reinforcement by increased interfacial strength. When specimens are tested above 1073 K, the oxidation of SiC enhances the oxidation damage of the composites and thus leads to the reduction in compressive strength.

It is also interesting to note that at temperatures above 873 K the higher the applied strain rate is, the less degradation of compressive strength is. At the strain rate of \( 10^{-4}/s \), a 24% reduction in compressive strength is observed at 1273 K, compared with that at 873 K. However, as the strain rate increases to \( 10^{-2}/s \) and above, the reduction in compressive strength is no more than 10%. Since the oxidative damage is responsible for the reduction in compressive strength, the observed phenomena as well as higher strain rate sensitivity at temperatures above 873 K (Fig. 8) suggests the strain rate (time) dependence of oxidative damage. Considering that same oxidation time is experienced when the C/SiC composite specimens are heated to the expected experimental temperature, the same amount of oxidative damage, namely pre-oxidation, will be induced before the specimens are subjected to different strain rates. Since lower strain rate means longer loading time, more severe stress-oxidation, which refers to the oxidative damage during the deformation process, is expected [6,7]. This may explain why the degradation of compressive strength is more severe as temperature increases from 873 to 1273 K at the strain rate of \( 10^{-4}/s \). In con-
Contrast, at high strain rates the stress-oxidation damage is negligible since the loading time is in the regime of tens of microseconds. Thereby, only pre-oxidation contributes to the oxidative damage and the strength loss is not so remarkable at high strain rates as temperature increases from 873 to 1273 K. Despite its time dependence, the stress-oxidation induced damage also relies on experimental temperature. At temperatures above 1073 K, oxidation damage of carbon fiber as well as silicon carbide matrix is intensified. Thus the strain rate sensitivity of compressive strength increases remarkably above 873 K (as seen in Fig. 8).

3.5. Macro- and micro- structural observations

Fig. 12 shows the optical images of ruptured specimens recovered from quasi-static experiments (0.0001/s) at three selected temperatures (293, 873 and 1273 K) as well as dynamic strain rate (1500/s) at room temperature (293 K). The loading direction is shown by the white arrows in the upper left corner of the images. For all specimens, several cracks can be observed on the side surfaces of specimens. However, unlike homogenous brittle materials, some cracks are tortuous due to the hindrance of fiber bundles to crack propagation. For specimens from quasi-static, room temperature (293 K) experiments, most of cracks show relative large angle to the loading direction, shown in Fig. 12(a). At high strain rate, the angles of cracks to the loading direction are much smaller than those at quasi-static strain rate, shown in Fig. 12(b). Similar phenomena are also observed by Liu and Li [8,9] in 2D and 3D C/SiC composites. In case of the strain rate of 0.0001/s, the cracks formed at high temperature show smaller angle to the loading direction compared with those at room temperature, as shown in Fig. 12(c and d). Considering a macro crack formed by the coalescence of interfacial cracks (cracks along interface) and fiber raptures (cracks perpendicular to the axis of fiber), more fiber raptures are expected for the formation of a macro crack with smaller angle to the loading direction. It has been proposed that the interface strength between fiber and matrix influences the failure behavior of fiber-reinforced composites [36,37]. Higher interfacial strength tends to degrade the possibility of horizontal interfacial cracking and makes cracks likely to penetrate into the fiber, thus cracks with small angle to the loading direction are formed [38,39]. Experimental results presented by Li [40] showed that enhanced interfacial strength was expected at high loading rate. Moreover, in case of high temperature experiments the release of TRS is also favorable to interfacial bonding strength [6]. Thus cracks achieved at high strain rates

![Fig. 12 – The ruptured scratches of the 2D-C/SiC composites under quasi-static and dynamic compression tests at elevated temperatures (a–d). Shown in (e) is a sketch of specimen geometry and fiber orientation.](image-url)
or at high temperatures display smaller angles with respect to the loading direction than those at room temperature under quasi-static loading.

Fig. 13 presents the fracture surfaces of the recovered specimens tested at 0.0001/s and at 293 K. For the specimen compressed at room temperature, the failure surface is fea-

Fig. 13 – SEM micrographs of a fractured C/SiC specimen under quasi-static compression: (a) fracture surface, (b–d) fiber bundle fractured.

Fig. 14 – SEM micrographs of a fractured C/SiC specimen under dynamic compression: (a) fracture surface, (b–d) fiber bundle splitting.
tured by fiber bundle breakage with extensive interfacial failure within bundle. The rough fracture surface of fiber bundles reflects the random breakage of carbon fibers. Fracture surfaces of the recovered specimen tested at the strain rate 1500/s and at 293 K are shown in Fig. 14. Fiber failure is now characterized by fiber bundle multiple splitting (breaking) with abrupt fracture surfaces. Different from that under quasi-static loading, the split fibers at high strain rate remain in bundles without extensive debonding, which also can be identified by the cracks penetrating straightly across the carbon fiber bundles. At high strain rate, in bundle failure seems to be suppressed and thus fiber bundles exhibit better integrity, which suggests stronger interfacial strength than that at low strain rate.

4. Conclusion

The mechanical behavior of two-dimensional (2D) carbon fiber reinforced silicon carbide (C/SiC) matrix composites was investigated at temperatures ranging from 293 to 1273 K under both quasi-static and dynamic uniaxial compression. The results show that the compressive behavior of 2D C/SiC composites in the air is a combined effect of applied loading rate and high temperature. The amount of oxidation is determined by the applied loading rate as well as the testing temperature. The interfacial bonding strength between fiber and matrix can be enhanced due to the release of thermal residual stress below the preparation temperature. With the competitive effects of thermal residual stress release and oxidation of carbon fibers, there is a gradual increase in the compressive strength of 2D C/SiC composites with temperature up to 873 K, and the compressive strength decreases as temperature further increases from 873 K. The reason is that at higher temperatures the strength degradation due to the oxidation dominates the process. The compressive strength of 2D C/SiC composites is oxidation sensitive at low strain rates due to relatively long loading time. In comparison, the effect of oxidation can be neglected at high strain rates where the loading time is in regime of tens of microseconds. The strain rate sensitivity factor of compressive strength of 2D C/SiC composites increases slightly as the temperature rises up to 873 K. At temperatures above 873 K, the compressive strength of the 2D C/SiC composites shows remarkable dependence on temperature. Strain rate has a significant effect on the compressive strength of 2D C/SiC composites at temperatures higher than 873 K. Post-deformation observation illustrates the remarkable differences in fracture surface between room temperature quasi-static and dynamic experiments and higher temperature quasi-static experiments, which is believed to be caused by the enhanced interfacial strength due to either high strain rate or the release of thermal residual stress at high temperature.

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