Journal of Materials Science & Technology xxx (2017) xxx–xxx

Contents lists available at ScienceDirect

Journal of Materials Science & Technology

journal homepage: www.jmst.org

Effect of matrix sub-layer interfacial fracture on residual strength improvement of the fatigued carbon/carbon composites

Jing Cheng, He-jun Li*, Shou-yang Zhang, Li-zhen Xue

State Key Laboratory of Solidification Processing, Northwestern Polytechnical University, Xi'an 710072, China

ARTICLE INFO

Article history:
Received 11 January 2017
Received in revised form 22 February 2017
Accepted 19 April 2017
Available online xxx

Keywords:
Carbon/carbon composites
Fatigue behavior
Interface
Strengthening mechanism

ABSTRACT

In the present study, flexural behavior of carbon fiber reinforced pyrolytic carbon matrix composites (C/C composites) before and after fatigue tests had been studied. The results showed that the residual flexural strengths of the samples had been improved after fatigue tests, and the fracture mechanisms of the original and post-fatigue specimens had some differences. Fracture mechanism of the original specimens could be described as fiber/matrix interfacial de-bonding, and the dominant damage of the post-fatigue specimens could be regarded as pyrolytic carbon sub-layers' step-delamination. The degradation of matrix sub-layer interfacial bonding strength was beneficial to improve the mechanical properties of C/C composites.

© 2017 Published by Elsevier Ltd on behalf of The editorial office of Journal of Materials Science & Technology.

1. Introduction

Carbon fiber reinforced pyrolytic carbon composite, often known as carbon/carbon composite (C/C), exhibits many superior properties at elevated temperature [1–4]. Their unique combination of mechanical and thermal properties makes them as an ideal choice for aerospace structural applications at temperature till 3000 °C either in vacuum or in inert environment [5]. As structural applications, the mechanical performance of C/C composites under fatigue loading needs to be paid more attentions, in order to assess the durability of the composites under sustained loading. It is reported by many scholars that the fatigue behavior of C/Cs has two distinguishing features: firstly, the C/Cs have a high fatigue limit (often 80%–90% of the static strength), and secondly, residual strengths of C/Cs after modest cyclic loading are higher than their static strengths [6–9].

Since the C/C is composed of carbon fibers, carbon matrix and various kinds of interfaces existed between different phases, the micro-structure of C/C is very complicated [10,11]. The interfaces existed inside C/Cs not only refer to the interfaces between fiber and matrix, but also include the interfaces between different carbon matrix sub-layers [12–15]. Since the interfacial bonding strength is lower than both of carbon fiber tensile fracture strength and pyrolytic carbon fracture strength [16], the cyclic loading during fatigue firstly exceeds the interfacial bonding strength and leads micro-cracks to generate and propagate along the interfaces. The interfacial damage accumulation was improved with the increase of fatigue cycles [10]. In other words, structural evolution of the C/Cs under fatigue loading is a process of interfacial damage accumulation.

It is accepted by many scholars that the appropriate interface damage modes make a positive contribution to the residual strength improvement of C/C composites after cyclic loading [17,18]. According to the former studies, foci are mainly concentrated on the fiber/matrix interfacial property evolution during cyclic loading. In the present study, a pyrolytic carbon matrix sub-layer fracture mechanism has been observed both in the post-fatigue 1D- and 2.5D- C/C composites. The matrix sub-layer fracture mechanism has been proved to play an important role in the residual strength improvement of the C/C samples.

2. Experimental

2.1. Raw materials

Materials used in this study were two kinds of C/C composites, the first kind was a unidirectional (1D) C/C composite and the second one was a needled no-woven carbon fiber felt reinforced (2.5D) C/C composite. The 1D-C/C samples were reinforced by polyacrylonitrile-based carbon fibers (Toray, T300-3K) aligned in only one direction. The fiber volume fraction of the unidirectional carbon fiber preforms was 39%–41%. The carbon fiber preforms

* Corresponding author.
E-mail address: lihejun@nwpu.edu.cn (H.-j. Li).

http://dx.doi.org/10.1016/j.jmst.2017.05.010
1005-0302/© 2017 Published by Elsevier Ltd on behalf of The editorial office of Journal of Materials Science & Technology.

of the 2.5D-C/C samples were fabricated by repeatedly stacked non-woven fabrics and carbon fiber webs with a needle-punching technique step by step. Then, the 1D- and 2.5D- carbon fiber preforms were densified in the processes of isothermal chemical vapor infiltration (ICVI). Carbon fiber of the unidirectional C/C composite was constrained by a model during the matrix infiltration process. The ICVI processes were conducted repeatedly to obtain the 1D-C/C samples with densities of 1.60 (±0.01) g/cm^3 and the 2.5D-C/C samples with densities of 1.80 ± 0.01 g/cm^3.

2.2. Mechanical tests

Static flexural strengths of the two kinds of as-prepared composites were measured by three-point bending tests in an ambient environment. Three-point bending tests in this study were carried out in accordance with Q/GB 95-92 [19]. The tests were carried out with a 40 mm span and a crosshead speed of 0.5 mm/min by means of an electrohydraulic-servo testing machine (Instron Co. Ltd, 8872). The number of specimens used in every bending test was eight. Specimens for bending tests were prepared using a diamond saw. The dimension of the specimens used in bending tests was 4 mm × 8 mm × 55 mm.

The flexural strength (σ, MPa) was given as follows:

$$\sigma = 3PL/2bh^2, \tag{1}$$

where P is the maximum load (N), L is the span length of flexural test (mm), b is the width of specimen (mm), h is the thickness of specimen (mm).

Testing machine used for the flexural fatigue tests was as same as the static flexural tests, as well as the testing mode and span length. The applied stress ratio (R: the ratio of minimum to maximum applied stress) of the fatigue tests was 0.1, the flexural fatigue loading frequency was 15 Hz, and the fatigue loading waveform was a sine wave. In this study, fatigue limits of the composites were determined as the value where the specimen was not failure even though the stress was repeated for 10^6 times. The residual flexural strengths were determined for surviving specimens after fatigue test using the mode as same as the static flexural test.

Values of fracture toughness were determined using single-edge-notched bend (SENB) test. All of the specimens (original and post-fatigue specimens) were notched in the center using a diamond blade with a thickness of 0.2 mm, and the notch depth was approximately 1 mm. The geometry of the specimen used in the fracture toughness test is shown in Fig. 1. The SENB tests were performed using Instron Co. Ltd, 8872 under a cross-head speed of 0.5 mm/min with a 32 mm span length. The number of specimens used in every SENB test was three. The fracture toughness values of the used C/Cs were determined from the net fracture stress, the dimension of SENB specimen and the pre-crack depth by use of the following equations [20]

$$KIC = 0.316f(\xi)(P_{IC}S/BW^{3/2}), \tag{2}$$

$$f(\xi) = \frac{3\xi^{1/2}\left[1.99 - \xi\left(1-\xi\right)\left(2.15 - 3.93\xi + 2.72\xi^2\right)\right]}{2\left(1+2\xi\right)\left(1-\xi\right)^{3/2}}, \tag{3}$$

where P_{IC} is fracture bending stress, S is span length, B is specimen width, W is specimen height, ξ is ratio of notch depth and specimen height.

2.3. Microstructure characterization

The pyrolytic carbon of the as-prepared composites was identified under a polarized light microscope (PLM, Leica DMLP optical microscope), and the micro-structures of the specimens before and after mechanical tests were observed using a scanning electron microscope (SEM, JSM-6360).

3. Results and discussion

Since the texture type of pyrolytic carbon matrix influenced the mechanical behavior of the C/C composite [21], the pyrolytic carbon of the two samples had been observed under PLM, and the micrographs of the 1D- and 2.5D- C/C samples are shown in Fig. 2(a) and (b), respectively. As indicated in Fig. 2(a), pyrolytic carbon matrix of the 1D-C/C exhibited binary layer textures (marked as A and B), the texture A refers to rough laminar (RL) texture, and the texture B is smooth laminar (SL) texture. As obviously observed in this micrograph, the thickness of the B layer is very thin. On the other hand, Fig. 2(b) reveals that all the pyrolytic carbon matrix of the 2.5D-C/C is typical RL texture.

Static flexural properties of the two samples were measured, and then the specimens were loaded on cyclic flexural loading under different maximum applied stresses, until the specimens failed or the number of cyclic loading was up to 10^6. The relationship between the maximum fatigue stresses and the number of cycles to failure (S-N curve) of the two samples is shown in Fig. 3. The vertical axes of the graphs in Fig. 3 were normalized by the average static flexural strength of the two samples, respectively. As shown in Fig. 3, the flexural fatigue limits of the 1D- and 2.5D-C/C samples were determined to be approximate 73% and 78% of their static flexural strength, respectively.

Then, the two C/C samples were loaded at the stress level of their flexural fatigue limit with 10^6 fatigue cycles, respectively. At the end of each fatigue test, no significant macroscopic damage, such as delamination or fracture, was evident. Then, the average residual flexural strengths and fracture toughness of the post-fatigue samples were tested. Fig. 4(a) shows the static and residual flexural test results of the 1D- and 2.5D-C/C composites. As shown in the graph, the average flexural strengths of the original and post-fatigue 1D-C/C specimens are 200.9 MPa and 230.7 MPa, respectively. And the average flexural strengths of the 2.5D-C/C composite before and after fatigue test are 165.7 MPa and 182.3 MPa, respectively. As illustrated in Fig. 4(b), the average fracture toughnesses of the original and post-fatigue 1D-C/C specimens are 2.03 MPa m$^{-1/2}$ and 2.37 MPa m$^{-1/2}$, respectively. Moreover, the average fracture toughnesses of the original and post-fatigue 2.5D-C/C specimens are 1.72 MPa m$^{-1/2}$ and 1.86 MPa m$^{-1/2}$, respectively. The results imply that both of the flexural strength and fracture toughness of the samples have been improved after fatigue tests.

The typical flexural stress-strain curves of the 1D- and 2.5D-C/C samples before and after fatigue test are shown in Fig. 4(c) and (d), respectively. As indicated in Fig. 4(c), the flexural failure modes of the original and post-fatigue 1D-C/C samples are similar. Both of the original and post-fatigue specimens, the flexural stress-strain curves could be divided into three stages. The first stage ended until the flexural stress was improved to the maximum value. And in the second stage, the stress-strain curve is shown as a fluctuated shape. In the last stage, the stress-strain curve shows a catastrophical decrease. The detailed study about the flexural behavior of the 1D-C/C sample was reported in Ref. [22]. According to the former literature, the energy absorption of material during mechanical tests could be estimated by the area under the stress-strain curve [23,24], and the absorption energy increased with the improving value of area under the stress-strain curve. Based on this theory, since the area under flexural stress-strain curve of the post-fatigue specimen is larger than that of the original specimen, the post-fatigue specimen needs to absorb more energy until ultimate failure during the flexural test. In conclu-

Download English Version:

https://daneshyari.com/en/article/7951962

Download Persian Version:

https://daneshyari.com/article/7951962

Daneshyari.com