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Damage evolution of cross-ply ceramic-matrix composites under stress-rupture and cyclic loading at elevated temperatures in oxidizing atmosphere

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A R T I C L E I N F O

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A B S T R A C T

In this paper, the damage evolution of cross-ply ceramic-matrix composites (CMCs) under stress-rupture and cyclic loading at elevated temperatures in oxidizing atmosphere has been investigated. The Budiansky-Hutchinson-Evans shear-lag model was used to describe the micro stress field of the damaged composite considering matrix multicracking, interface debonding/wear/oxidation/slip. The damage parameters of fatigue hysteresis dissipated energy, fatigue hysteresis modulus, fatigue peak strain and fatigue hysteresis width have been used to monitor the damage evolution inside of CMCs. The relationship between damage parameters and internal damage of matrix multicracking, interface debonding/slip has been established. The experimental fatigue hysteresis loops, interface slip lengths, fatigue hysteresis dissipated energy and peak strain of cross-ply SiC/MAS composite under cyclic fatigue and stress-rupture at 566 °C and 1093 °C in air have been predicted.

1. Introduction

Ceramic-matrix composites (CMCs) have been considered for long duration applications in aeroengine, i.e., the combustion liners, turbine vanes and shroud applications, a strong understanding of their behavior under conditions of sustained load at elevated temperatures is needed [1].

Many researchers performed the experimental and theoretical investigations on stress-rupture and cyclic fatigue behavior of fiber-reinforced CMCs. Mehrman et al. [2] investaged the effect of hold times on the fatigue behavior of an oxide/oxide CMCs at elevated temperature in air and in steam environment. The fatigue life with hold times exceeded those produced in creep but was shorter than those obtained in fatigue, and the presence of steam significantly degraded the material performance. Morscher [3] investigated the tensile creep and rupture of 2D SiC/SiC composite at 1315 °C in air. It was found that the creep and rupture of SiC/SiC depend on the creep resistance of fiber-type and the matrix-type. Gowayed et al. [4] investigated the accumulation strain evolution under dwell-fatigue of 2D SiC/SiC with and without holes for different stress ratio, stress levels, testing temperatures and hold times. It was found that the accumulation strain increases with increasing of time and stress levels. During stress upture and cyclic fatigue loading, the damage evolution inside of the composites should be monitored to predict the lifetime. Li [5] established the relationship between the damage evolution inside of CMCs under cyclic fatigue loading, i.e., matrix multicracking, interface denonding/wear/slip, and fibers fracture with the hysteresis dissipated energy-based damage parameter, and developed an approach to predict the fatigue life of CMCs using the hysteresis dissipated energy-based damage parameter.

The objective of this paper is to investigate the damage evolution of cross-ply CMCs under stress-rupture and cyclic loading at elevated temperature in air. The damage parameters of fatigue hysteresis dissipated energy, fatigue hysteresis modulus, fatigue peak strain and fatigue hysteresis width have been used to monitor the damage evolution inside of CMCs. The experimental fatigue hysteresis loops, interface slip lengths, fatigue hysteresis dissipated energy, and the peak strain of cross-ply SiC/MAS composite under stress-rupture and cyclic loading at 566 °C and 1093 °C in air have been predicted.

2. Material and experimental procedures

The Nicalon™ SiC (Nippon Carbon Co., Ltd., Tokyo, Japan) fiber-reinforced barium-stuffed magnesium aluminosilicate (MAS) cordierite matrix composite (SiC/MAS CMCs) was received from Corning (Corning, NY USA) [6], and manufactured using the hot pressing method at temperature above 1200 °C. The volume fraction of fibers was approximately 40%.

The tension-tension fatigue tests were conducted on a MTS servo hydraulic load-frame (MTS Systems Corp., Minneapolis MN, USA), and performed by Grant [6]. The longitudinal deformation was measured with the aid of an MTS 632.65B-03 extensometer with quartz rods. The rods were sharpened and the extensometer calibrated to a gauge length of 25.4 mm prior to each of the tests. The tension-

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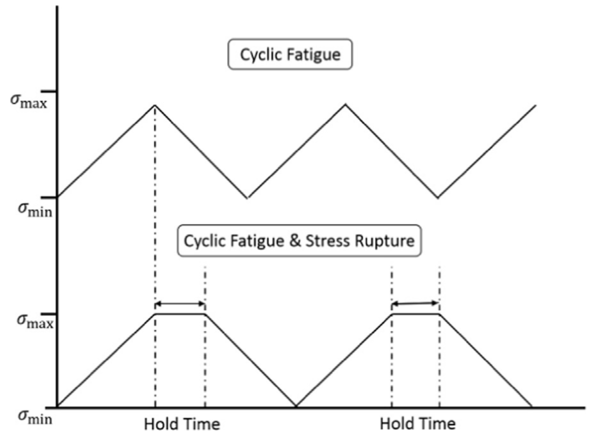

Fig. 1. The schematic of cyclic fatigue loading and cyclic fatigue loading with hold time.

tension fatigue tests at 566 °C and 1093 °C in air were performed under the load control at a triangular waveform with the loading frequency of 1 Hz and the fatigue load ratio, i.e., minimum to maximum stress, of 0.1, and the maximum number of applied cycles was defined to be 1,000,000 applied cycles.

Under cyclic loading at elevated temperature, there are two types of loading sequences considered as shown in Fig. 1, including:

(1) Case 1: cyclic fatigue loading without stress-rupture time, and the interface debonding and frictional slipping between the fiber and the matrix is mainly affected by interface wear.
(2) Case 2: dwell-fatigue loading with stress-rupture time, and the interface debonding and frictional slipping between the fiber and the matrix is mainly affected by the interface oxidation.

At 566 °C in air, the tensile strength was about 292 MPa, as shown in Fig. 2, and the fatigue peak stress was 138 MPa with the stress-rupture time of t=0, 1, 10 and 100 s. At 1093 °C in air, the tensile strength was about 209 MPa, as shown in Fig. 2, and the fatigue peak stress was 103 MPa with the stress-rupture time of t=0, 1 and 10 s.

3. Theoretical analysis

Under stress-rupture and cyclic fatigue loading at elevated temperature, the interface wear and interface oxidation affect the damage evolution inside of CMCs [7–15]. When the interface oxidation region and the interface wear region is less than the matrix crack spacing, upon unloading, the interface debonded region can be divided into three regions, i.e., interface counter-slip region with the interface shear stress of τ_f ($x\in[0, \xi]$), interface counter-slip region with the interface shear stress of $\tau_i(N)$ ($x\in[\xi, y]$) and interface slip region with the interface shear stress of $\tau_i(N)$ ($x\in[y, l_d]$), in which y denotes the interface counter-slip length. Upon reloading, the interface debonded region can be divided into four regions, i.e., interface new-slip region with the interface shear stress of τ_f ($x\in[0, z]$), the interface counter-slip region with the interface shear stress of τ_f ($x\in[z, \xi]$), the interface counter-slip region with the interface shear stress of $\tau_i(N)$ ($x\in[\xi, y]$) and the interface slip region with the interface shear stress of $\tau_i(N)$ ($x\in[y, l_d]$). The unloading and reloading stress–strain relationship is determined by Eq. (1).

$$\begin{aligned}\varepsilon_{c\text{pu}} = &\frac{2\sigma l_d}{V_f E_f l_c} + \frac{2\tau_f}{r_f E_f l_c}\xi^2 + \frac{4\tau_f}{r_f E_f l_c}\xi(l_d - \xi) + \frac{4\tau_i(N)}{r_f E_f l_c}(y - \xi)^2\\ &- \frac{2\tau_i(N)}{r_f E_f l_c}(2y - \xi - l_d)^2\\ &+ \frac{2\sigma_{fo}}{E_f l_c}\left(\frac{l_c}{2} - l_d\right) + \frac{2r_f}{\rho E_f l_c}\left[\frac{V_m}{V_f}\sigma_{mo} + \frac{2\tau_f}{r_f}\xi + \frac{2\tau_i(N)}{r_f}(2y - \xi - l_d)\right]\\ &\times\left[1 - \exp\left(-\rho\frac{l_c/2 - l_d}{r_f}\right)\right] - (\alpha_c - \alpha_f)\Delta T\end{aligned} \tag{1a}$$

$$\begin{aligned}\varepsilon_{c\text{pr}} = &\frac{2\sigma}{V_f E_f l_c} l_d - \frac{4\tau_f}{r_f E_f l_c}z^2 + \frac{2\tau_f}{r_f E_f l_c}(2z - \xi)^2 - \frac{4\tau_f}{r_f E_f l_c}(2z - \xi)(l_d - \xi)\\ &+ \frac{4\tau_i(N)}{r_f E_f l_c}(y - \xi)^2\\ &- \frac{2\tau_i(N)}{r_f E_f l_c}(2y - \xi - l_d)^2 + \frac{2\sigma_{fo}}{E_f l_c}\left(\frac{l_c}{2} - l_d\right)\\ &+ \frac{2r_f}{\rho E_f l_c}\left[\frac{V_m}{V_f}\sigma_{mo} - \frac{2\tau_f}{r_f}(2z - \xi) + \frac{2\tau_i(N)}{r_f}(2y - \xi - l_d)\right]\\ &\times\left[1 - \exp\left(-\rho\frac{l_c/2 - l_d}{r_f}\right)\right] - (\alpha_c - \alpha_f)\Delta T\end{aligned} \tag{1b}$$

where V_f and V_m denote the fiber and matrix volume fraction, respectively; E_f denotes the fiber elastic modulus; ξ denotes the interface oxidation length; l_d denotes the interface debonded length; l_c denotes the matrix crack spacing; r_f denotes the fiber radius; α_f and α_c denote the fiber and composite thermal expansion coefficient, respectively; and ΔT denotes the temperature difference between the fabricated temperature T_0 and testing temperature T_1 ($\Delta T=T_1-T_0$).

When the interface oxidation region and the interface wear region is equal to matrix crack spacing, the unloading and reloading stress–strain relationship is determined by Eq. (2).

$$\begin{aligned}\varepsilon_{c\text{fu}} = &\frac{\sigma}{V_f E_f} - \frac{2\tau_f}{r_f E_f l_c}\xi^2 + \frac{2\tau_f}{r_f E_f}\xi + \frac{4\tau_i(N)}{r_f E_f l_c}(y - \xi)^2\\ &- \frac{2\tau_i(N)}{r_f E_f l_c}\left(2y - \xi - \frac{l_c}{2}\right)^2 - (\alpha_c - \alpha_f)\Delta T\end{aligned} \tag{2a}$$

$$\begin{aligned}\varepsilon_{c\text{fr}} = &\frac{\sigma}{V_f E_f} - \frac{4\tau_f}{r_f E_f l_c}z^2 + \frac{2\tau_f}{r_f E_f l_c}(2z - \xi)^2 - \frac{4\tau_f}{r_f E_f l_c}(2z - \xi)(y - \xi)\\ &+ \frac{4\tau_i(N)}{r_f E_f l_c}(y - \xi)^2\\ &- \frac{4\tau_f}{r_f E_f l_c}(2z - \xi)\left(\frac{l_c}{2} - y\right) - \frac{2\tau_i(N)}{r_f E_f l_c}\left(2y - \xi - \frac{l_c}{2}\right)^2 - (\alpha_c - \alpha_f)\Delta T\end{aligned} \tag{2b}$$

The area associated with the stress/strain hysteresis loops is the dissipated energy during corresponding cycle, which is defined by Eq. (3).

$$U_e = \int_{\sigma_{\min}}^{\sigma_{\max}} [\varepsilon_{c\text{unload}}(\sigma) - \varepsilon_{c\text{reload}}(\sigma)]d\sigma \tag{3}$$

where ε_{c_unload} and ε_{c_reload} denote unloading and reloading strain, respectively. Substituting unloading and reloading strains of Eqs. (1) and (2) corresponding to interface partially and completely debonding into Eq. (3), the hysteresis dissipated energy U_e can be obtained.

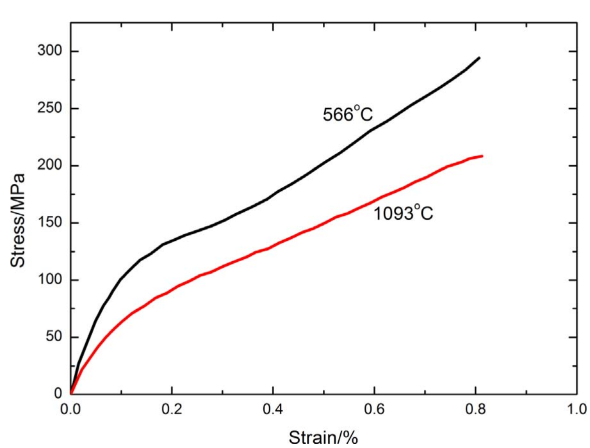

Fig. 2. The tensile stress–strain curves of cross-ply SiC/MAS composite at 566 °C and 1093 °C in air.