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Mechanical behavior and properties of fiber reinforced ceramic matrix composites for high temperature use

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Abstract : Ceramics can keep their mechanical characteristics up to 2 000 °C or higher. In this paper, A model to predict ultimate strength of continuous fiber-reinforced brittle matrix composites is developed. A statistical theory for the strength of a uni-axially fiber-reinforced brittle matrix composite is presented. Also a semi-empirical frictional heating method for estimating in-situ interfacial shear in fiber-reinforced ceramic matrix composites was improved. Local uneven fiber packing variation as well as uneven micro-damage during fatigue can be expected to have effects on the composites: generation of frictional heating, thermal gradients, and residual stresses around local fiber breaks. This study examined those engineering interests by the finite element method.

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Introduction

It is not possible to efficiently design optimal composites for a designated use and in result to select the constituents of a desired composite a priori without understanding strengthening as well as failure mechanism. In the last two decades ceramic matrix composites are designed for extremely high temperature use such as jet propulsion engine components. Traditionally ceramics are well known to be one of most stable materials in all environmental conditions as well as their brittleness. Ceramics reinforced by advanced fibers (ceramic matrix composite, CMC) can overcome their fatal brittle weakness where interfacial characteristic between fiber and matrix plays a key role. As part of these efforts, investigating tensile strength and interfacial shear of composites is important. In this paper we consider Weibull statistics, geometric distribution, and mechanical properties of constituent materials.

Continuous fibers and whiskers have been

commonly utilized to reinforce the materials. Many advanced fibers show strong brittleness which leads to a wide strength spectrum which can be commonly represented by a Weibull distribution. The major purpose in nonceramic matrix composites is to have the fiber share greater proportion of the applied load. This load sharing depends on the ratio of fiber and matrix elastic moduli. In polymeric matrix composites, this ratio is very high, while in ceramic matrix composites, it is rather low and can be as low as unity. In the ceramic matrix composites, the purpose of reinforcing is mainly to enhance toughness by utilizing fiber/matrix interfacial characteristics. Thus, the mechanism of ultimate tensile strength should be understood on the basis of frictional resistance along the interface as well as fracture strength of constituents.

Once that the composite is under the uniaxial tensile load, the composite stress increases linearly with the corresponding fiber stress. When the composite strain exceeds to the fracture strain of the matrix at any applied load, multiple

matrix cracks occur and thus intact fibers which bridge the matrix cracks should bear higher load. After the matrix cracking process ceases , additional external load directly transfers to the fiber. During the process of load increase ,there gradually form statistical fiber breaks. The broken fibers that still bridge the crack planes by the friction force formed between matrix and fibers ,resist to be pulled out of the matrix. Finally ,when it is impossible for the small number of unbroken fibers and frictionally resisting fibers to endure the additional increase of external load ,the catastrophic failure occurs in the composite specimen.

This study has two purposes: the first is to build a method to predict ultimate tensile strength of continuous fiber-reinforced brittle matrix composites and the second is to investigate an interfacial shear estimation method influenced by uneven fiber packing in composites.

1 Method of simulation

1.1 CMC under uniaxial monotonic tensile loading

The brittle matrix of this study is uniaxially reinforced by continuous fibers as shown in Fig. 1. Once that the composite is under a uniaxial tensile load (σ), the composite stress (σ) increases linearly with the corresponding fiber stress (σ). When the composite strain exceeds to the fracture strain of the matrix at any applied load, a matrix crack occurs and thus intact fibers which bridge the matrix cracks should bear higher load. Stress level at the matrix crack surface should be zero and load transferring from the intact fibers to the matrix through interfaces causes the matrix stress recovery in a distance depending on interfacial properties and external load. So the recovered matrix stress can make another matrix crack and the scenario repeats until the matrix stress is anywhere below the matrix strength^[1] (see Fig. 2). After the matrix cracking process ceases ,additional external load directly transfers to the fiber. During the process of load increase ,the fibers gradually break at some locations in a random fashion. The fracture

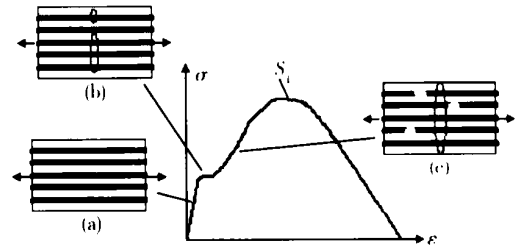


Fig. 1 Schematic of mechanical behavior of a ceramic matrix composite subjected to a uniaxial tensile loading where S_T denotes a tensile strength of the composite: (a) a linear response region; (b) a matrix cracking region; and (c) a region of fiber breaks and pull-out.

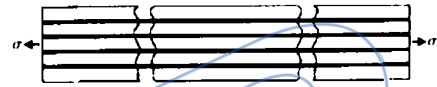


Fig. 2 Matrix cracking model of CMC of the fiber depends on the size of surface flaw and its flaw distribution follows Weibull type statistics^[2]. The broken fibers that still bridge the crack planes by the friction force formed between matrix and fibers ,resist to be pulled out of the matrix. Finally ,when it is impossible for the small number of unbroken fibers and frictionally resisting fibers to endure the additional increase of external load ,the catastrophic failure occurs in the composite specimen.

1.2 Problem formulation

As stated before ,the tensile strength of advanced fibers shows the Weibull statistical distribution where the failure of materials depends on flaw distribution and size. In the brittle material of Weibull statistics under the uniform load , the probability of failure at a stress σ is generally expressed by

$$P = 1 - \exp \left\{ - \frac{1}{\alpha} \left(\frac{\sigma - \sigma_0}{\sigma_0} \right)^m \right\} \quad (1)$$

where m is the shape factor (Weibull modulus) , α is the dimension (length ,area ,or volume) factor relating to the characteristics of flaws ,and α and σ_0 are normalizing parameters. And the material has no probability to fail at a stress less than σ_0 . $\alpha \sigma_0^m$ is a constant depending on the specimen and determined by experiment^[3]. In the a-

bove equation, since a fiber fracture depends on the surface flaw rather than the defects in the volume, $\mathbf{A}$ should be considered as the area, A .

After the matrix cracking occurs, frictional slip zone along the interface of fiber and matrix forms in the matrix crack wake. The fiber stress, σ , which would be a maximum on the matrix crack surface ($z = 0$) decreases at a rate of $4\pi d_f$ (say κ) along the interface and becomes a plane strain stress at the end of the slip zone where load transfer between fiber and matrix happens^[1]. The stress τ is defined as the shear stress working along the slip zone caused by the friction force. Most of fiber break locations would be clustered in the vicinity of matrix crack because the fracture of fibers depends on the local stress. For simple calculation, we may neglect fiber fractures outside the slip zone and the slip zone length is obtained as $l = \sigma_f / \kappa$; σ_f is the maximum fiber stress (i.e., stress at the matrix crack plane) and thus local fiber stress may be expressed as $\sigma_f = \kappa(l - z)$. Neglecting the lower fracture limit stress σ_0 , integration of Eq. (1) over the debond area results in the failure probability of one specimen as Eq. (2)

$$P = 1 - \exp\left\{-\frac{\pi d_f^2 \tau^m}{2A_0 \sigma_f^m (m+1)} \left(\frac{4l}{d_f}\right)^{m+1}\right\} \quad (2)$$

However Eq. (2) does not provide the failure sites in the specimen. In order to include the sites information the probability density function may be derived by following Oh and Finnie^[4]. A fiber is divided into $2N$ elements of $\delta z = l/N$ each located at z_i from the matrix crack plane ($z = 0$). Let $G_{\delta i}(\sigma)$ denote the probability that the element δz_i fractures when subjected to the stress less than σ . A local stress σ is a function of the reference stress σ_f and z_i . Thus $G_{\delta i}(\sigma)$ may be expressed by $G_{\delta i}(\sigma_f, z_i)$. The probability that the element survives at the stress less than σ_f is $[1 - G_{\delta i}(\sigma_f, z_i)]$. The probability that all elements survive is then the product of all individual survival probabilities. Therefore, provided that it has survived when the reference stress is less than σ_f , the probability $P_{\delta i}$ that the element at z_i fractures when the reference stress is be-

tween σ_f and $\sigma_f + \delta\sigma_f$ can be obtained. Also evaluating Weibull distribution Eq. (1) with $\sigma_f = 0$ for $G_{\delta i}(\sigma_f, z_i)$, the probability that the element at z_i breaks when subjected to a local uniform stress σ is given by

$$G_{\delta i}(\sigma_f, z_i) = 1 - \exp\left\{1 - \frac{\pi d_f \delta z_i}{A_0} \left(\frac{\sigma}{\sigma_0}\right)^m\right\} \quad (3)$$

Substituting Eq. (3) into $P_{\delta i}$ and performing the differentiation, when the number of elements $N \rightarrow \infty$ the probability density function $p(\sigma_f, z)$ is derived as

$$p(\sigma_f, z) \delta\sigma_f \delta z = \frac{\pi d_f}{A_0} \frac{m \sigma_f^{m-1}}{\sigma_f^m} \times \exp\left\{-\frac{2\pi d_f}{A_0} \int_0^l \left(\frac{\sigma}{\sigma_0}\right)^m dz\right\} \delta\sigma_f \delta z \quad (4)$$

Considering the local stress σ as $\kappa(l - z)$ and integrating Eq. (4) multiplied by the weighting factor z , the average fracture distance from the matrix crack plane may be estimated by

$$\langle h \rangle = \frac{S_0}{\kappa(m+1)} \gamma\left(\frac{m+2}{m+1}, t_f\right) \quad (5)$$

where a new variable $t = (\sigma/S_0)^{m+1}$ is introduced (t_f is associated with σ_f) and S_0 is defined by

$$S_0 = \left\{ \frac{2A_0 \sigma_f^m \tau(m+1)}{\pi d_f^2} \right\}^{\frac{1}{m+1}} = \left\{ \frac{2H(\hat{S}_f(H))^m \tau}{d_f \ln 2} \right\}^{\frac{1}{m+1}}$$

where H is the fiber specimen length and $\hat{S}_f(H)$ is the stress at which 50 % of specimens break. It is meaningful to note that $v_f S_0$ (v_f is defined as the fiber volume fraction) would be a composite tensile strength when m goes to infinity (shown later in this communication). Interestingly observed from the above, the dimension change of fiber diameter d_f to d_f/M , keeping $\hat{S}_f(H)$ constant, should cause the increase of tensile strength by $M^{1/(m+1)}$. Likewise, the increase of interfacial shear τ to $M\tau$ would improve the same amount as fiber diameter d_f to d_f/M by $M^{1/(m+1)}$. However the upgrading S_f to MS_f would lead to the increase of $M^{m/(m+1)}$ times, about the same order of fiber strength enhanced; so this is the most efficient way to improve the composite tensile strength. As the dimensionless load t_m goes to infinity, all of fibers would break and the incomplete Gamma function $\gamma(\cdot, \infty)$ reaches to the Gamma function $\Gamma(\cdot)$. When the

tensile strength of a material is deterministic ($m \rightarrow \infty$), the probability density function of strength is Dirac delta and then $\langle h \rangle = 0$, indicating that all fibers break at the matrix crack plane.

After the occurrence of the matrix cracking, the fiber failure probability $P(\sigma_f)$ at σ_f is determined by Eq. (2) and the rest of fibers, $1 - P(\sigma_f)$, would survive. Since the fiber broken at a distance h resists the friction force of $\pi \tau_d h$ to pull out of the matrix, the composite stress of fiber volume fraction v_f , neglecting the load bearing from the unbroken matrix ligaments, may be estimated by

$$\begin{aligned} \sigma_c &= \sigma(\text{from intact fibers}) + \sigma(\text{from friction}) \\ &= \sigma_f v_f (1 - P(\sigma_f)) + k v_f \langle h \rangle P(\sigma_f) \end{aligned} \quad (6)$$

where, since the load needed to pull out the broken fiber out of the matrix is proportional to the failure distance, the expected friction force for all of the failed fibers is expressed in terms of the average failure distance.

2 Interfacial shear measurement

Micro-frictional heating phenomenon of FRCMCs under cyclic loading condition was reported^[7] and well known. The frictional heat of this phenomenon depends on the range, magnitude, and direction of fatigue stress^[8], the loading frequency, the matrix crack spacing^[8,12], the broken fibers^[10], and the material properties of the composite system. And because the frictional heat of the specimen is dissipated into environments, the interfacial shear can be derived by considering energy balance.

2.1 Friction work rate of a composite under fatigue loading

Cho and coauthors^[7-11] proposed an interfacial shear measurement which utilizes the frictional heating phenomenon of FRCMC during cyclic loading. In unidirectionally fiber reinforced CMCs, a major cause of frictional heating is dissipation energy due to micro-frictional sliding between reinforcing fibers and matrix along the debonded interface. The frictional sliding is clas-

sified as micro (or partial) and gross slip. In the beginning stage, the most of slip is partial one of which average sliding length is less than a half distance of matrix crack spacing. After assuming only defects of matrix cracks and corresponding interfacial debonding in FRCMCs without broken fibers, the friction work of one load cycle generated in FRCMCs subjected to cyclic loading is estimated by^[8]

$$W_{\text{unfrac}} = \frac{\pi d_f^3 C^2 \Delta \sigma^2}{96 v_f E_f \tau_d} \quad (7)$$

where d_f is the fiber diameter, $\Delta \sigma$ the fatigue stress range, v_f the fiber volume fraction, E_f the elastic modulus of a fiber, τ_d the interfacial shear stress, and C stands for $(1 - v_f) E_m / E_f$. Multiplying Eq. (7) by the fatigue frequency f and dividing by the volume of the gage section of FRCMC cylinder model, $\pi D_c^2 l / 4 = \pi d_f^2 l / 4 v_f$, the energy dissipation rate per unit volume of FRCMC is calculated as

$$\dot{W}_{\text{unfrac}} = \frac{f C^2 d_f \Delta \sigma^2}{24 E_f l \tau_d} \quad (8)$$

where l is the mean matrix crack spacing.

On the other hand, beside of matrix cracks, if there exist fiber failures, then Eq. (7) and (8) must be modified accordingly. In order to include fiber break in the model, the cylindrical composite of Fig. 3 was utilized and axi-symmetrically modeled. All of applied loads should be supported by fibers only at the matrix crack plane, so the fibers should bear much higher load near matrix crack. Fibers would fail at some locations in the crack wake after the matrix cracking occurs. In order to predict fiber fracture locations at a given applied load, weakest link theory can be adopted^[14].

Since fiber fractures strongly depend on local stress and fiber local stress recovers the stress in plane strain at the end of slip zone, fracture sites are clustered around the matrix crack plane where the fiber stress is maximum and most of fractures should occur in the slip zones. In the case of occurrence of broken fibers in the composite, micro and gross frictional slip must be considered to calculate frictional energy dissipation.

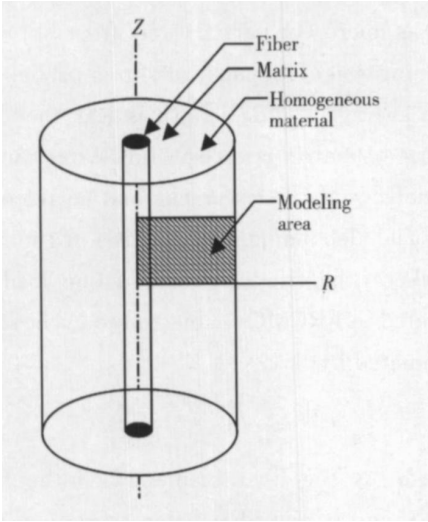


Fig. 3 Fiber-reinforced composite cylinder model

pation during cyclic loading. In micro slip case there are four slip zones^[10]. Hence energy dissipation has the same of unbroken fiber case. If matrix crack opening is less than a critical amount, say δ_m to induce gross frictional slip, the cell of Fig. 3 remains in microslip phase throughout the cycle. In the other hand, the case of the matrix crack opening exceeding the critical amount is followed by gross slip phase after microslip. During the ensuing gross slip phase, the only relative motion occurs in the fractured fiber by an amount, say δ and results in a frictional work of $2p\tau_d f h \delta$ where p is contact pressure on the debonded interface, h the distance of fiber failure from matrix crack surface. Thus rather than Eq. (7), the frictional work per unit volume done by a fractured cell can be represented by

$$w_{\text{frac}} = \frac{16\pi(1+\mu)\tau_d^2 h^3}{3(1+\mu)^2 E_f} + 2\pi\tau_d f h \delta \quad (9)$$

where $\mu = \nu_f E_f / (1 - \nu_f) E_m = \nu_f / C$ is defined. The microslip limit, δ_m , can be derived as

$$\delta_m = \frac{8\tau_d h^2}{(1+\mu) E_f d_f} \quad (10)$$

Total crack opening amount of a cylinder composite model with broken fiber at a distance h from the crack surface is derived by

$$U_t = \frac{u}{1+\mu} + l \left\{ \frac{\mu}{1+\mu} \sqrt{\frac{8u(1+\mu)\tau_d}{E_f d_f}} - \frac{8\tau_d h \mu}{E_f d_f (1+\mu)} \right\} + \frac{8\tau_d h^2 \mu (1-\mu)}{E_f d_f (1+\mu)} \quad (11)$$

where u is defined as

$$u = \frac{d_f \Delta \sigma}{8(1+\mu)\tau_d E_f \nu_f^2}$$

Thus the amount of the gross slip distance must be expressed by

$$\delta = U_t - \delta_m \quad (12)$$

Multiplying Eq. (9) by load frequency and dividing it by unit cell volume $\pi d_f^2 l / 4 \nu_f$, the energy dissipation rate per unit volume is estimated by

$$\dot{w}_{\text{frac}} = \frac{dw_{\text{frac}}}{dt} = \frac{4f \nu_f w_{\text{frac}}}{\pi d_f^2} \quad (13)$$

2.2 Heat energy balance

The mechanical work against microfriction force along the debonded interfaces of Eq. (7) or (9) is dissipated into the surrounding medium in the form of heat. As the saturation of matrix cracking is achieved under a constant fatigue load, the frictional work rate and/or the temperature distribution of the composite remain constant to be a steady state condition. Considering the heat energy forms of convection, conduction, and radiation for heat transfer, the steady state heat energy dissipation rate through the specimen into environments is estimated as follows

$$\dot{q} = \dot{q}_{\text{conv}} + \dot{q}_{\text{rad}} + \dot{q}_{\text{cond}} = \frac{2L_{\text{surf}}}{V_{\text{gage}}} \int_{\text{inter}}^{\text{end}} \times [h((T_s - T_a) + \beta(T_s^4 - T_a^4))] dz + \frac{2kA_{\text{cond}}}{V_{\text{gage}}} \frac{dT_s}{dz} \quad (14)$$

where L_{surf} is the peripheral length of the gage section, V_{gage} the gage section volume, β the radiation coefficient, β Stefan-Boltzman constant ($5.67 \times 10^{-8} \text{ W/m}^2 \cdot \text{K}^4$), T_s the surface temperature, T_a the atmospheric temperature, k the heat conductivity, h the heat convection coefficient, A_{cond} the gage cross section area, and z is a coordinate along fiber direction. The surface temperature T_s may depend on measuring points, $T_s(z)$.

If we neglect the other forms of energy transmission except heat transfer, by equating Eqs. (8) or (13) as Eq. (14) with the measured temperature of the specimen test spots, we can calculate the in-situ interfacial shear τ_d of FRC-MCs.

3 Uneven fiber packing model

In order to estimate the temperature distribution of a FRCMC specimen under fatigue ,the finite element method was utilized for the Nicalon SiC/ CAS- II composite specimen which is uniaxially reinforced by 35 % fiber volume fraction. The loading condition was 25 Hz loading frequency ,the fatigue stress $\Delta\sigma=170\text{ MPa}$ ($\sigma_{\max}=180\text{ MPa}$) ,the mean crack spacing of the specimen under study is used as $l=198\text{ }\mu\text{m}^{[9,14]}$. The mechanical properties of the composite were used as listed in Table 1 for the analysis.

Table 1 SiC/ CAS- II composite material properties^[8]

Composite Material	ρ (kg/ m ³)	C (J/ kg · K)	K (W/ m · K)	α (°Q)	E / (GPa)
Axial	2 712.5	0.72 *	5.16	4.45E-6	131.25
Trans.	2 712.5	0.72 *	1.0 *	5.92E-6	118.40

* Assumed values

From the symmetry of the tensile specimen as shown in Fig. 4 ,only one eighth of the whole specimen was analyzed by employing the transient state finite element technique. The frictional work rate of Eq. (13) was uniformly distributed over the problem domain and the surrounding was kept at a constant temperature of 20 °C through the analysis. The experimental equation $h=1.42(T_s-T_a)/L^{1/4}$ was adopted for the heat convection coefficient^[13].

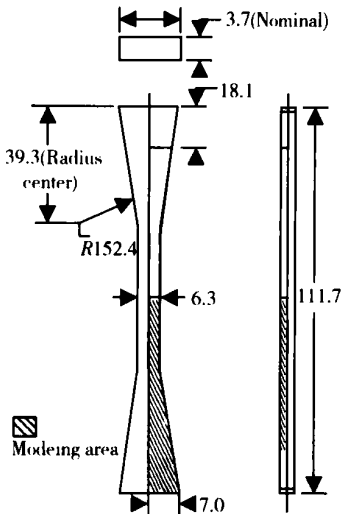


Fig.4 Specimen shape and modeling area

Besides ,in order to include fiber break in the model ,the one filament cylindrical composite of Fig.3 was utilized by embedding one filament cylindrical model into orthotropic homogeneity equivalent to the original FRCMC in terms of orthotropic mechanical properties and then axisymmetrically modeled. Thus the resultant computational model including broken fiber can be considered as a compound. In the analysis ,the fiber volume fraction v_f of 0.35 ,the fatigue loading frequency of $f=50\text{ Hz}$,the fatigue stress range $\Delta\sigma=310\text{ MPa}$ ($\sigma_{\max}=320\text{ MPa}$) and the interfacial shear of $4.2\text{ MPa}^{[8]}$ were used for the frictional work rate of a fiber broken cell^[10] to examine the influence of the fiber break on the residual stresses.

The composite compound as described above was modeled by axisymmetric 2-D finite element model consisting of 1 248 axisymmetric finite elements and 1 320 nodes. Heat sources for the finite element computation were calculated from Eq. (9) and were uniformly distributed over the frictional sliding areas. The corresponding thermal stress calculation was post-processed ,following the temperature distribution prediction a priori.

4 Results and discussion

4.1 CMC under uniaxial monotonic tensile loading

Since the composite starts to fail catastrophically when the bridging fibers cannot endure additional increase of the external load ,in this stage the maximum fiber stress σ increases in a unstable fashion. And thus the condition for the catastrophic failure of the composite specimen can be stated by $d\sigma/d\delta=0$.

Thus ,substitution of Eqs. (2) ,(5) and (6) into the instability condition results in a nonlinear equation.

$$1-(m+2)t_f-t_f e^{-t_f}+t_f^{\frac{-m}{m+1}}\gamma\left(\frac{m+2}{m+1},t_f\right)=0$$

(15)

Figure 5 shows the solution behavior with the increase of Weibull modulus. Let a solution of Eq. (15) be T_f and then $\sigma_f=T_f^{1/(m+1)}S_0$. Sub-

stituting Eqs. (2) and (5) into Eq. (6) and rearranging, the final equation to determine the ultimate tensile strength of the composite system, S_T , is given by

$$\frac{S_T}{v_f S_0} = \frac{1}{m+1} \gamma \left(\frac{m+2}{m+1}, T_f \right) + \left\{ \frac{1}{T_f^{m+1}} - \frac{1}{m+1} \gamma \left(\frac{m+2}{m+1}, T_f \right) \right\} e^{-T_f} \quad (16)$$

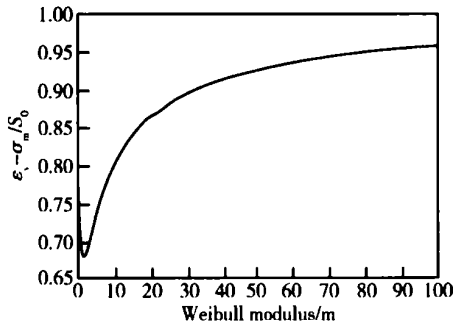


Fig. 5 The root behavior of Eq. (15) is shown against Weibull modulus

In this model the fiber fractures are assumed to occur after matrix cracking and thus the predicted tensile strength of Eq. (16) should exceed the composite stress that corresponds to the matrix cracking stress S_{mu} for validity. In the undamaged composite subjected to the tensile load, the strains of the composite, fiber and matrix are the same ($\epsilon_c = \epsilon_f = \epsilon_m$) until the first matrix crack occurs. When the composite strain becomes equal to the failure strain of the matrix, the composite stress is $\sigma_c = E_c S_{mu} / E_m = S_{mu} \{ 1 + v_f (E_f / E_m - 1) \}$. And thus in any case, this is the minimum tensile strength.

4.2 Interfacial shear measurement influenced by fiber packing

Figure 6 shows that the steady state temperature distribution along the centerline of FRCMC double sided specimen of Fig. 4 under the tensile fatigue loading which was assumed only matrix cracking without a fiber break.

The typical temperature distribution along the axial direction was simulated to be quadratic by using FEM. The temperature distribution^[8] was assumed to be constant over the gage section. Thus in this study, a quadratic curve fitted temperature distribution was used to estimate

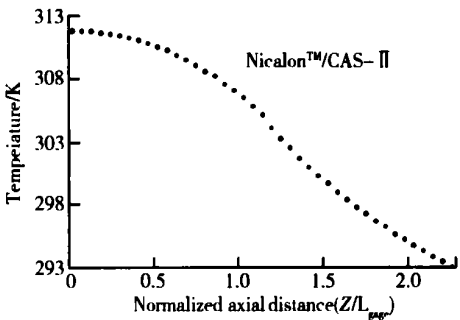


Fig. 6 Temperature distribution along symmetric plane of a double edge specimen

the frictional heat dissipation amount of Eq. (14). The dissipation should be equal to the mechanical work rate of Eq. (13) to give 5.25 MPa for the interfacial shear stress value. Note that the reported shear of the same system was 5.3 MPa which was evaluated on the basis of the experimental hysteresis loop^[8].

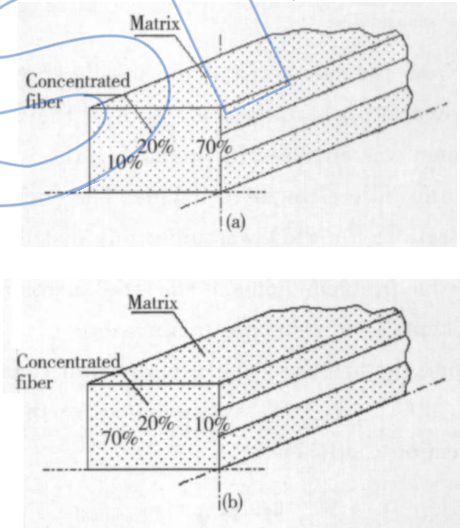


Fig. 7 Fiber concentrated specimens : (a) case 1 ; (b) case 2.

When fiber bundles in CMCs are bias-packed during manufacturing, the frictional heating may be somewhat different due to packing variation. To examine this, two extreme examples of Fig. 7 were numerically simulated by using the finite element method. The computational results for surface temperature are shown in Fig. 8. The fiber packing of Fig. 7 (a) shows 0.5 % deviation of interfacial shear compared to the perfect uniform packing and Fig. 7 (b) case leads the interfacial shear to 5.1 % different from the perfect fiber packing one. Thus, the in-

situ interfacial shear is reasonably estimated with the frictional heating method despite of uneven fiber packed CMCs.

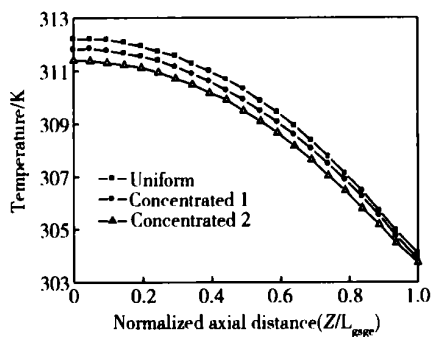


Fig. 8 Temperature differences due to fiber packing variations

5 Conclusions

The unidirectional continuous fiber reinforced brittle matrix composites when subjected to a uniaxial tensile load are considered to analyze the tensile strength. The matrix is assumed to be homogeneous and isotropic elastic materials and the fiber to have Weibull statistical characteristics for its tensile strength. The analysis provides new physical insights for the tensile strength of the brittle matrix composites. In particular, it shows how the role of interfacial friction force as well as mean fiber strength influences on the composite strength. The equation for predicting the tensile strength is derived. The derived equation is very sensitive to the fiber strength data. And thus in order to use the equation a caution should be placed to have the accurate data for a composite. The model of this research would give a upper limit of the strength.

From this study we also conclude that the second order temperature distribution over the gage section should improve the interfacial shear estimation accuracy from the fractional heating measurements. As a result, two temperature measuring points over the gage section should be required for the improved calculation. However fiber packing variations in CMCs do not effect on estimated interfacial shear by using the method. In the case of fiber failures in CMCs, the thermal stress has an effect on interfacial shear level. The

influence may be adverse in the mechanical capability of CMCs for some thermal expansion mismatch between fiber and matrix. However Nicalon/CAS-II of this study showed a little effect of order of 1 MPa at most.

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