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Micromechanics and Creep Behavior of Fiber-Reinforced Polyether-Ether-Ketone Composites

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ABSTRACT: A micromechanics formulation of composite properties in terms of the constituent properties using Eshelby's inclusion principle and Mori-Tanaka's mean field method. A linear four-element Burger mechanical model has been employed for application to viscoelastic polymers. The influence of volume fractions of inclusions on the overall creep strain of a polyether-ether-ketone (PEEK)-matrix composite is investigated. The creep rate of the PEEK-matrix which depends linearly on the creep strain and the primary creep, resulting from the viscous flow of creep deformation, is also considered in addition to the usual steady state, or secondary creep. In order to examine the applicability of the model to pure PEEK resin and carbon/PEEK, a series of experiments on PEEK 150P and unidirectional carbon fiber-reinforced PEEK composites has been carried out, which included constant rate of tensile tests and constant load tensile creep tests. It is found that the pure PEEK resin and its composite satisfies the four element Burger model, and that the model agrees with the experimental results extremely well. Also, it is demonstrated that this simple, albeit micromechanic theory is capable of predicting the volume fraction dependence of the time dependent creep, with a characteristic consistent with the known elastic behavior.

1. INTRODUCTION

THIS PAPER IS concerned with the quantitative determination of the time-dependent creep of a PEEK-matrix composite, where unidirectional carbon fiber (AS4) are homogeneously dispersed in the PEEK matrix. The fibers are assumed to remain elastic and the matrix linear elastic and isotropic. The matrix and fibers will be taken to be perfectly bonded together, without any void nucleation or growth. Here, the composite as a whole is transversely isotropic, and its overall creep behavior will be effected by the volume function of the fibers.

The method to be presented is based on the theoretical framework recently established by Zhu and Weng [1] for the creep anisotropy of a metal-matrix com-

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posite containing different volume fractions of aligned spheroidal inclusions, which is further rooted in Eshelby's [2] and Mori-Tanaka's [3] theories.

When such a composite is subjected to a constant, external stress, the initial response is elastic, and its overall elastic strain can be determined from its effective moduli. Mori-Tanaka's method in conjunction with Eshelby's [2] solution of an ellipsoidal inclusion has been adopted by Taya and Chow [4], Taya and Mura [5], Weng [6] and Benveniste [7], among others, to study various elastic properties of an inhomogeneous medium with some success. Pertinent to the present investigation are the finding of Weng [6] and Zhao et al. [8], in which the effective bulk and shear moduli of a particle-reinforced composite and the five effective moduli of a fiber (circular)-reinforced composite have been proved, respectively, to coincide with the Hashin and Shtrikman [9] bounds, and the Hill [10] and Hashin [11] bounds. These findings provide us with some basis of confidence, in that the initial elastic response of the composite and the average stress of the matrix are both reasonably estimated by this method. The initial creep rate of the PEEK matrix then can be determined from its average stress at a given volume fraction of fibers.

Using Eshelby's [2] equivalent-inclusive principle and Mori-Tanaka's [3] mean-field method, the composite is subsequently replaced by the comparison material filled with equivalent transformation strains. This approach allows one to find the average stress of the matrix in terms of the macroscopic stress, and then by appealing to the constitutive equation of the PEEK matrix, the overall stress-strain relation of the two-phase system can be easily determined.

In this paper the constitutive equations of the PEEK matrix will be briefly recapitulated first, and then the initial stress state of the PEEK matrix, suitable for calculation of the initial creep rate, will be derived. The micromechanical principles involved in assessing the stress variation of the matrix following an incremental creep will be established subsequently, and the overall creep will be presented at the end. This result is then checked by comparison with the results of a series of experiments on carbon fiber-reinforced PEEK composite specimens.

2. CONSTITUTIVE EQUATION OF THE PEEK MATRIX

For linearly viscoelastic materials, the total strain of material at any constant temperature and at any time is composed of three independent component parts, i.e., (a) the elastic strain (b) the transient creep strain and (c) the steady state creep strain. We have

$$\epsilon = \epsilon_1 + \epsilon_2 + \epsilon_3 \quad (2.1)$$

where ϵ is the total strain, ϵ_1 the elastic strain, ϵ_3 the transient creep strain, and ϵ_2 is the steady state creep strain. Thus, we use the four-parameter Burger model to represent the deformation:

$$\dot{\epsilon}_1 = \frac{\dot{\sigma}}{E_1} \quad (2.2a)$$

$$\dot{\epsilon}_2 = \frac{\sigma}{\eta_1} \tag{2.2b}$$

$$\dot{\epsilon}_3 = \frac{1}{\eta_2}(\sigma - E_2\epsilon_3) \tag{2.2c}$$

From Equation (2), we can get

$$\epsilon(t) = \epsilon_1 + \epsilon_2 + \epsilon_3 = \frac{\sigma}{E_1} + \frac{\sigma}{\eta_1}t + \frac{\sigma}{E_2}(1 - e^{-(E_2/\eta_2)t}) \tag{2.3}$$

$$\dot{\epsilon}(t) = \frac{\sigma}{\eta_1} + \frac{\sigma}{\eta_2}e^{-(E_2/\eta_2)t} \tag{2.4}$$

in which $\epsilon(t)$ and $\dot{\epsilon}(t)$ are the total creep strain and total creep strain rate induced by an axial loading, σ . Since the stress field in a composite is always multi-dimensional due to the existence of fibers, Equation (2.3) is extended to give a 3-D relation between the stress and creep strain field as follows.

Since creep deformation shows some similarities to plastic deformation of metals, the elastic (time-independent) part of the strain is obtained from Hooke’s law, while the time-dependent strain may be obtained from the following assumptions:

- (a) Incompressibility of the material.
- (b) Coaxiality of principle directions of stresses and strain rate.

The matrix phase in a composite system often exists in a triaxial stress state. To accommodate this, Equation (2.4) can be generalized to in terms of Von Mises’s effective stress σ_e :

$$\dot{\epsilon}_e(t) = \frac{\sigma_e}{\eta_1} + \frac{\sigma_e}{\eta_2}e^{-(E_2/\eta_2)t} \tag{2.5}$$

where

$$\sigma_e = \left(\frac{3}{2} \sigma'_{ij} \sigma'_{ij} \right)^{1/2} \tag{2.6}$$

and σ'_{ij} is defined the deviatoric stress. Moreover, if we also adopt the Prandtl-Reuss flow rule, the creep strain rate can be expressed as

$$\dot{\epsilon}_{ij}(t) = \frac{3}{2} \frac{\epsilon_e}{\sigma_e} \sigma'_{ij} \tag{2.7}$$

This set of constitutive equations allows us to calculate the creep strain, $\dot{\epsilon}_{ij}(t)$, at a given σ_{ij} and time t in the PEEK matrix.

3. THE INITIAL STRESS OF THE PEEK MATRIX

The average stress of the matrix containing homogeneously dispersed circular fibers will be derived by employing Mori-Tanaka's [3] idea.

We shall refer to the matrix in the composite system as phase 0 and the inclusions as phase 1. The elastic moduli tensor of the γ -th phase will be denoted by \mathbf{c}_γ and its volume-fraction by f_γ . Taking the matrix phase to be elastically isotropic and the inclusion to be transversely isotropic, we write, in Hill's [10] shorthand notations,

$$\begin{aligned} \mathbf{C}_0 &= (3\kappa^0, 2\mu^0) \\ \mathbf{C}_1 &= (E_{11}, E_{22}, G_{12}, \nu_{12}, \nu_{23}) \end{aligned} \quad (3.1)$$

where κ^0 , and μ^0 are the elastic bulk modulus and shear modulus, respectively, of the matrix phase, and E_{11} , E_{22} , G_{12} , ν_{12} , ν_{23} are the Young's moduli, shear moduli and Poisson's ratios, respectively, of the carbon fiber.

The composite is now subjected to a boundary traction giving rise to a uniform stress $\bar{\sigma}$. To facilitate the analysis, we also introduce an identically shaped comparison material with the property of the matrix moduli. It is subjected to the same boundary condition. The strain in the comparison material is then given by

$$\epsilon^0 = \mathbf{C}_0^{-1} \bar{\sigma} \quad (3.2)$$

The average stress and strain of the matrix in the composite system usually differ from $\bar{\sigma}$ and ϵ^0 due to the presence of inclusions. Denoting the differences by $\tilde{\sigma}$ and $\tilde{\epsilon}$, respectively, we may write its average stress as

$$\underline{\sigma}^{(0)} = \bar{\sigma} + \tilde{\sigma} = \mathbf{C}_0(\epsilon^0 + \tilde{\epsilon}) \quad (3.3)$$

Those of the inclusions further differ from the values of the surrounding matrix by the perturbed σ^{pt} and ϵ^{pt} , respectively, such that their mean stress can be written in terms of their own elastic moduli \mathbf{C}_1 , as well as in terms of Eshelby's [2] equivalence principle,

$$\begin{aligned} \underline{\sigma}^{(1)} &= \bar{\sigma} + \tilde{\sigma} + \sigma^{pt} = \mathbf{C}_1(\epsilon^0 + \tilde{\epsilon} + \epsilon^{pt}) \\ &= \mathbf{C}_0(\epsilon^0 + \tilde{\epsilon} + \epsilon^{pt} - \epsilon^*) \end{aligned} \quad (3.4)$$

where ϵ^* is Eshelby's equivalent transformation strain (or Mura's eigenstrain [18]), introduced into the regions of comparison material corresponding to the sites of the composite occupied by the inclusive to yield the same $\underline{\sigma}^{(1)}$. The perturbed strain ϵ^{pt} is related to ϵ^* through Eshelby's \mathbf{s} -tensor, as

$$\epsilon^{pt} = \mathbf{S} \epsilon^* \quad (3.5)$$

The components of the **S**-tensor depend on the secant Poisson ratio of the matrix ν_0 and the aspect ratio (the length/diameter ratio) of the inclusion. Since $\bar{\sigma} = \Sigma f_r \underline{\sigma}^{(r)}$, one has $\bar{\sigma} = -f_r \underline{\sigma}^{pr}$, or

$$\underline{\bar{\epsilon}} = -f_i(\underline{\epsilon}^{pr} - \underline{\epsilon}^*) = -f_i(\mathbf{S} - \mathbf{I})\underline{\epsilon}^* \tag{3.6}$$

where **I** is the fourth-order identity tensor. Substituting Equations (3.5) and (3.6) into the last of (3.4) yields the equation for $\underline{\epsilon}^*$ as

$$\underline{\epsilon}^* = [\mathbf{A}^{-1} - f_0\mathbf{S} - f_i\mathbf{I}]^{-1}\underline{\epsilon}^0 \tag{3.7}$$

where

$$\mathbf{A}^{-1} = [\mathbf{I} - \mathbf{C}_0^{-1}\mathbf{C}_1]^{-1} \tag{3.8}$$

and **I** is the symmetric fourth order identity tensor

$$I_{ijkl} = \frac{1}{2}(\delta_{ik}\delta_{jl} + \delta_{il}\delta_{jk}) \tag{3.8a}$$

Once $\underline{\epsilon}^*$ is known, the initial strain of the composite can be evaluated from $\underline{\bar{\epsilon}} = \Sigma f_r \underline{\bar{\epsilon}}^{(r)}$, leading to

$$\underline{\bar{\epsilon}} = \underline{\epsilon}^0 + f_i \underline{\epsilon}^* \tag{3.9}$$

The effective elastic moduli of the composite then follow from $\bar{\sigma} = \mathbf{C}\bar{\epsilon}$ as

$$\mathbf{C} = \mathbf{C}_0[\mathbf{I} + f_i(\mathbf{A}^{-1} - f_0\mathbf{S} - f_i\mathbf{I})^{-1}]^{-1} \tag{3.10}$$

Now returning to Equation (3.3) for the initial mean stress of the matrix, we find

$$\underline{\sigma}^{(0)} = \mathbf{C}_0(\mathbf{A}^{-1} - \mathbf{S})(\mathbf{A}^{-1} - f_0\mathbf{S} - f_i\mathbf{I})^{-1}\mathbf{C}_0^{-1}\bar{\sigma} \tag{3.11}$$

for the calculation of the initial creep rate.

4. THE MICROMECHANICS OF CREEP BEHAVIOR IN CARBON FIBER-REINFORCED PEEK COMPOSITES

Following Eshelby's [2] equivalent-inclusion principle and Mori-Tanaka's [3] mean stress by the comparison material filled with equivalent transformation strains, from section 3, we have the stress distribution of the PEEK matrix $\underline{\sigma}^{(0)}$ and the strain of the matrix $\underline{\epsilon}^{(0)}$ under a given stress $\bar{\sigma}$, individually as

$$\underline{\sigma}^{(0)} = \mathbf{C}_0(\mathbf{A}^{-1} - \mathbf{S})(\mathbf{A}^{-1} - f_0\mathbf{S} - f_i\mathbf{I})^{-1}\mathbf{C}_0^{-1}\bar{\sigma} \tag{4.1}$$

Substitution Equation (4.1) into Equation (2.6) leads to the creep rate of the

PEEK matrix. Under a given $\bar{\sigma}$, the initial creep rate of the PEEK matrix then can be calculated, and this, over a time increment dt , generates an incremental creep strain,

$$d\epsilon^c = \dot{\epsilon}^c dt \quad (4.2)$$

Such is also the case for any subsequent time increment. Our concern now is the nature of stress redistribution between inclusions and the PEEK matrix following an incremental creep, with $\bar{\sigma}$ kept constant.

In general, an incremental creep in the PEEK matrix is also accompanied by a stress reduction, which has to be carried over by the elastic inclusions to maintain a constant $\bar{\sigma}$. Here, we use the approximate method suggested previously by the writers (Zhu and Weng, [1,12,13]), which was originally formulated for the self-consistent determination of the time-dependent creep of polycrystals [14]. This model takes the constraint power of the matrix to be elastic and, therefore, is more suitable for the small creep strain range.

To estimate the average stress variations in both phases, we first bring the inclusions out and let the matrix creep homogeneously over a short time, dt . The inclusions are now subjected to a boundary traction corresponding to a homogeneous stress, $c_1 d\epsilon^c$, so that they fit perfectly into the deformed holes. A negative traction, $-c_1 d\epsilon^c$, is now applied to the bonded interfaces to remove the unwanted traction discontinuity. This is equivalent to having an Eshelby's "stress-free" strain,

$$d\epsilon^* = -d\epsilon^c \quad (4.3)$$

in the context of the inhomogeneity/transformation problem (see also Mura's eigenstrain [18]). Then, by means of Mori and Tanaka's [3] concept of mean-field approximation, the average stress variation in the PEEK matrix can be written as

$$d\bar{\sigma}^{(0)} = d\bar{\sigma} = C_0 d\bar{\epsilon} \quad (4.4)$$

in parallel to Equation (3.3). Similarly, in parallel to Equation (3.4), the mean stress variation of the inclusions can be cast into

$$\begin{aligned} d\bar{\sigma}^{(0)} &= d\bar{\sigma} + d\bar{\sigma}^{pr} = C_1 (d\bar{\epsilon} + d\epsilon^{pr} - d\epsilon^*) \\ &= C_0 (d\bar{\epsilon} + d\epsilon^{pr} - d\epsilon^* - d\epsilon^{**}) \end{aligned} \quad (4.5)$$

noting that a stress-free transformation strain, $d\epsilon^*$, already exists in the inclusion. Strain $d\epsilon^{**}$ is Eshelby's equivalent transformation strain, and together with $d\epsilon^*$, they form the total transformation strain, such that

$$d\epsilon^{pr} = S(d\epsilon^* + d\epsilon^{**}) \quad (4.6)$$

The boundary condition $d\bar{\sigma} = \Sigma f_i d\sigma^{(r)} = 0$ further leads to

$$d\bar{\sigma} = -f_1 d\sigma^{pr} \tag{4.7}$$

or

$$d\bar{\epsilon} = -f_1 [d\epsilon^{pr} - (d\epsilon^* + d\epsilon^{**})] = -f_1 (\mathbf{S} - \mathbf{I})(d\epsilon^* + d\epsilon^{**}) \tag{4.8}$$

To find the overall strain variation over dt , we recall that a uniform creep strain, $d\epsilon^c$, already took place before the inclusions were placed back into the deformed holes. It follows that

$$d\bar{\epsilon}^c = d\epsilon^c + f_0 d\bar{\epsilon} + f_1 (d\bar{\epsilon} + d\epsilon^{pr}) = d\epsilon^c + f_1 (d\epsilon^* + d\epsilon^{**}) \tag{4.9}$$

where the strain increment, $d\epsilon^c$, is the inclusion's deformation of the taken out state. Meanwhile, the composite now has the uniform deformation strain $d\epsilon^c$. Therefore, we need to add $d\epsilon^c$ to form the total strain increment, $d\bar{\epsilon}^c$. Using Equations (4.5), (4.6) and (4.8), $d\epsilon^* + d\epsilon^{**}$ then can be derived as

$$d\epsilon^* + d\epsilon^{**} = \underline{B}_1 d\epsilon^c \tag{4.10}$$

where

$$\underline{B}_1 = (\mathbf{I} - f_0 \underline{A} \mathbf{S} - f_1 \underline{A})^{-1} (\mathbf{I} - \underline{A}) \tag{4.10a}$$

$$\underline{A} = (\mathbf{I} - \mathbf{C}_0^{-1} \mathbf{C}_1) \tag{4.10b}$$

Substituting Equations (4.3) and (4.10) into Equation (4.9) leads to

$$d\bar{\epsilon} = (\mathbf{I} - f_1 \underline{B}_1) d\epsilon^c \tag{4.11}$$

Therefore, the stress increment of the matrix $d\sigma^{(0)}$ and the strain increment of the matrix $d\epsilon^{(0)}$ then can be derived as

$$d\sigma^{(0)} = f_1 \mathbf{C}_0 (\mathbf{S} - \mathbf{I}) \underline{B}_1 d\epsilon^c \tag{4.12}$$

$$d\epsilon^{(0)} = f_1 (\mathbf{S} - \mathbf{I}) \underline{B}_1 d\epsilon^c + d\epsilon^c \tag{4.13}$$

The total creep strain of the composite system, $\bar{\epsilon}$, and the PEEK matrix, $\bar{\epsilon}^{(0)}$, after a time increment can be written as

$$\bar{\epsilon}(t + dt) = \bar{\epsilon}(t) + d\bar{\epsilon}(t) \tag{4.14}$$

$$\bar{\epsilon}^{(0)}(t + dt) = \bar{\epsilon}^{(0)}(t) + d\bar{\epsilon}^{(0)}(t) \tag{4.15}$$

The total stress of the PEEK matrix after a time increment is

$$\sigma^{(0)}(t + dt) = \sigma^{(0)}(t) + d\sigma^{(0)}(t) \quad (4.16)$$

Therefore, by using the small time increment approach, we can predict the creep history of the carbon fiber-reinforced PEEK composites.

5. EXPERIMENTS

PEEK 150P polymers used to determine APC-2 PEEK matrix properties and APC-2 PEEK matrix composites were selected for this study because these types of thermoplastics have been increasingly used in commercial, military, and aerospace applications. These materials also illustrate the behavior of semicrystalline thermoplastics and their composites. It is recognized that the grade of PEEK used in APC-2 may differ from PEEK 150P in molecular weight distribution and additive content. However, according to the Fiberite Corporation, PEEK 150P is the commercial grade PEEK closest to the APC-2 matrix material.

The tensile and creep test specimens were cut out of plates manufactured using the procedures described below. PEEK 150P, in powder form, was dried in an oven at 150°C for three hours. The powder was then poured into a steel mold. The PEEK was then processed as follows:

1. The mold was heated from room temperature to 395°C (~1.5 hours).
2. The mold was kept at 395°C for 25 minutes and was then cooled by air to 360°C (~30 minutes).
3. A 3 kg/cm² pressure was applied while the mold was kept at 360°C for 25 minutes.
4. The pressure was maintained and the mold was cooled to 200°C (~1.5 hours).
5. The pressure was released and the mold was cooled to room temperature (~3 hours).

In general terms, increasing the cooling rate increases the residual stress state by driving the viscous to elastic transition temperature to higher values. In this situation, the volumetric change associated with crystallization contributes more to the state of residual stress. In contrast, residual stress predicted by Chapman et al., [15] for extremely slow cooling rates (0.3°C/sec for 40 plies of laminates) are almost completely relaxed out. In our experiment, we have controlled the cooling rate is 0.32°C/sec for 4 plies of laminates which is corresponded to the extremely slow-cooling-rate condition, and therefore, the residual stresses could be ignored. It is known that the mechanical properties of semicrystalline thermoplastics depend strongly on the crystalline structure of the polymer. However, within the ranges studied, the shear strengths and moduli of the polymer and the fracture toughness of the polymer and the composite seem to be insensitive to the thermal history employed during processing and to depend mainly on the crystallinity. Therefore, within accuracies required for our analyses, the mechanical

properties of PEEK 150P polymer and APC-2 composite can be related to crystallinity, this crystallinity being obtained by DSC in this study.

A small sample taken from the plate (~ 10 mg) was enclosed in an aluminum sample pan and placed into the DSC cell. The heat flow in the cell as a function of temperature was measured at a constant heat rate of $36^\circ\text{F}/\text{minute}$. The crystallinity was determined from the expression

$$C = \frac{H_t}{H_{ult}}$$

where H_t is the net amount of heat absorbed on heating from the glass transition temperature (293°F) to 720°F as measured by the DSC. H_{ult} the ultimate heat of crystallization of 100 percent crystalline polymer, was taken to be 130 J/g [16].

In evaluating the crystallinities of the APC-2 composite, adjustments were made for the fiber content using the expression

$$C = \frac{C_{measured}}{WP}$$

where WP is the weight percent of the polymer in the composite. Since the density of the polymer depends on the crystallinity, an iterative procedure was used to calculate the weight percent polymer in the composite.

The PEEK 150P plates thus produced had a nominal thickness of about 0.5 mm. It was very important to adhere strictly to the above procedure. Even small variations in the process resulted in plates which had excessive void contents or rough surfaces. The PEEK 150P plates manufactured in the above manner had an average crystallinity of 32 percent. The APC-2 composite plates used in this study were provided by ICI. The procedures was essentially the same as that used for the PEEK 150P polymers and their thickness was about 0.5 mm.

Two types of tests were performed: uniaxial tensile tests under a constant strain rate and tensile creep tests under a constant load. The mechanical properties of the PEEK 150P polymer were measured by tensile test (ASTM D638 type I). The creep test of carbon/PEEK used followed the ASTM D3090 standard test. All the tensile and creep tests were performed in a laboratory environment of 50% relative humidity (RH) and at 25°C .

6. RESULTS AND DISCUSSION

It is now of interest to examine the influence of the f_i of inclusions on the overall creep as predicted by the theory. To this end, we assumed in our calculations that the inclusions and the matrix took the properties of carbon fiber and the PEEK, respectively, at room temperature. This is a 3-D problem, and the aspect ratio l/d of the inclusion is assumed as ∞ to account for long fibers. The elastic constants of both phases are listed below.

Based on several creep curves tested at different loads, the creep properties of PEEK 150P at room temperature have been found to be $E_1 = 4.45 \text{ GPa}$, $\eta_1 =$

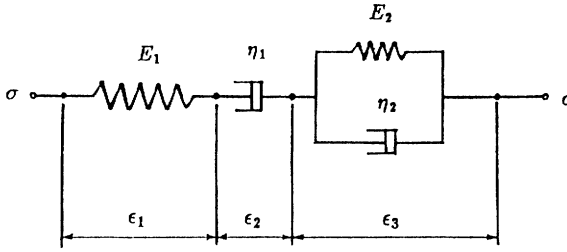


Figure 1. Linear four-element Burger's model.

2.609×10^4 GPa·hr, $E_2 = 154.809$ GPa, $\eta_2 = 3.901 \times 10^2$ GPa·hr and $\nu = 0.3$. The mechanical properties of carbon fibers [17] were $E_{11} = 235$ GPa, $E_{22} = 13.25$ GPa, $G_{12} = 27.58$ GPa, $\nu_{12} = 0.2$, $\nu_{23} = 0.5$. The quantitative accuracy of the present theory can to some extent be assessed by a comparison between its predicted results and the experimental results. Such a comparison is shown in Figures 2–5. The volume fractions of 0%, 17.7%, 21% and 44% of carbon fiber-reinforced PEEK composites are under two different constant creep test loads, where the solid line represent the theoretical predictions, and the symbols represent the experimental results. The results show that the unidirectional carbon fibers had the reinforced effect.

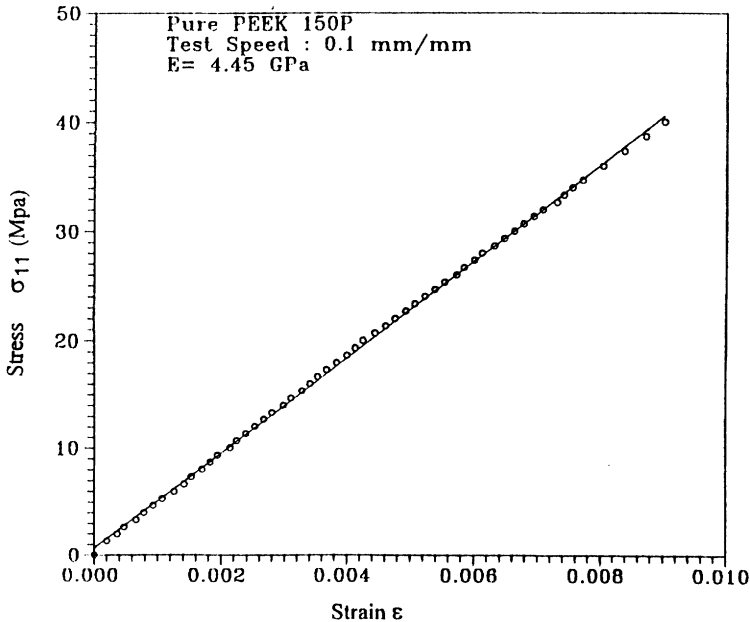


Figure 2. The stress σ_{11} versus strain ϵ curve.

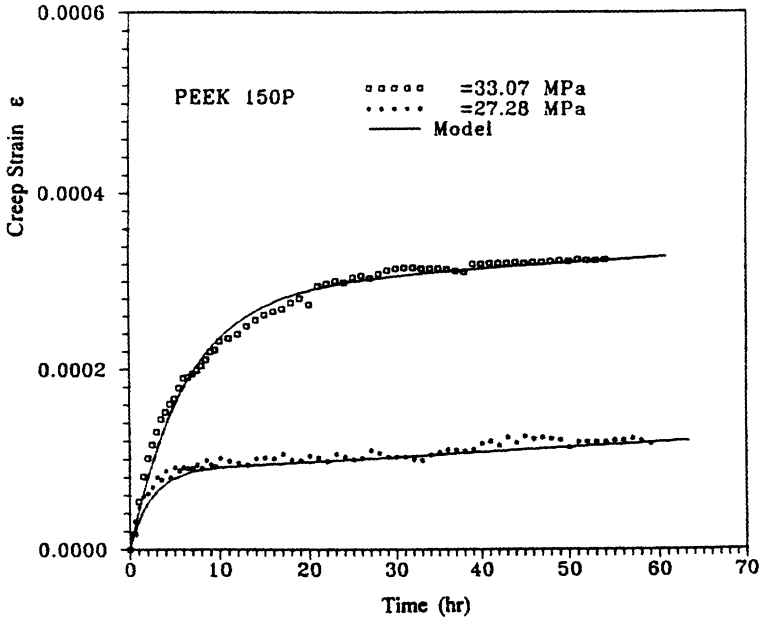


Figure 3. The creep strain ϵ versus time curve of $V_f = 0.00$.

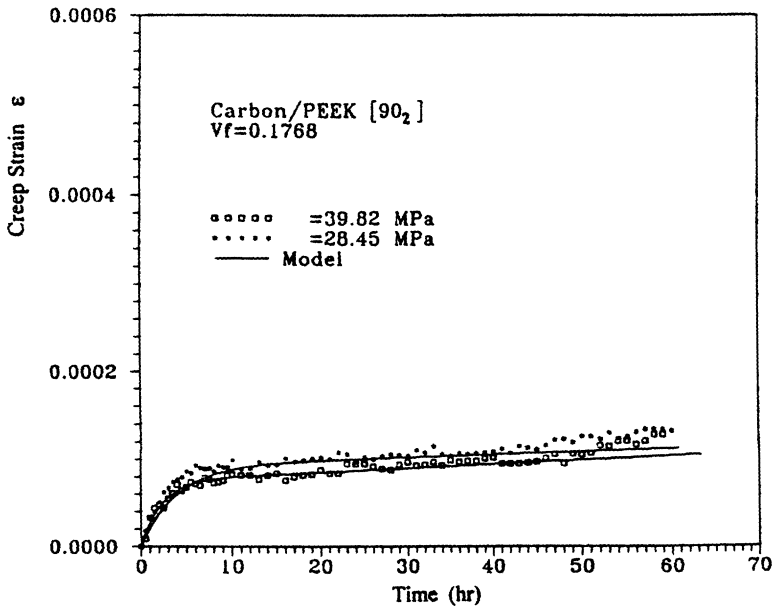


Figure 4. The creep strain ϵ versus time curve of $V_f = 0.17$.

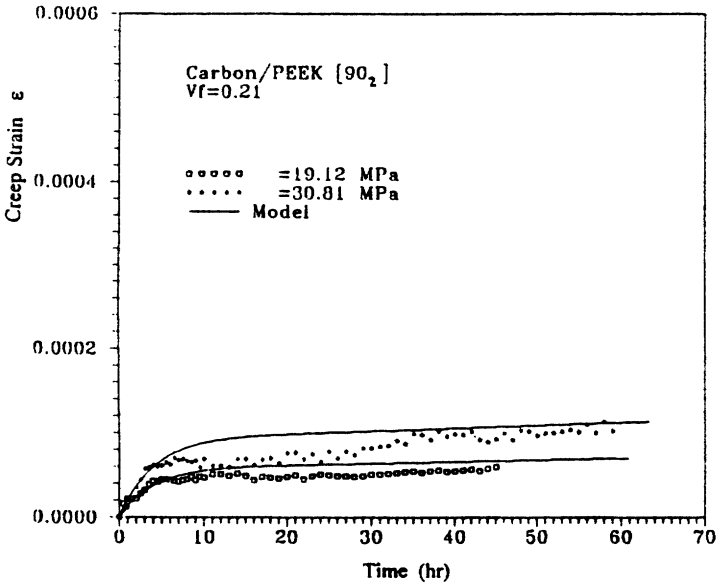


Figure 5. The creep strain ϵ versus time curve of $V_f = 0.21$.

7. SUMMARY

This paper has used the experimental method to study the creep behavior of a carbon fiber-reinforced PEEK composite. The specimens made of PEEK 150P and different volume fractions of carbon fibers underwent uniaxial tensile tests and constant-loading creep tests. The Burger model was employed to describe the linear viscoelastic stress-strain and strain rate relationship of PEEK.

In addition, this paper has been concerned from the micromechanics point of view, with the quantitative determination of the time-dependent creep of PEEK-matrix composites, where carbon fibers are homogeneously dispersed in the PEEK matrix. The fibers were assumed to remain elastic, but at an elevated temperature the PEEK matrix could undergo both primary (transient) and secondary (steady-state) creeps. The creep rate of PEEK usually depended linearly on the stress. The matrix and fibers were taken to be perfectly bonded together, without any void nucleation or growth. Here the composite as a whole was transversely isotropic, and its overall creep behavior was affected by the volume fraction of the fibers. The micromechanical principles involved in assessing the stress variation of the matrix following an incremental creep were subsequently established, and the theoretical predictions of the overall creep test were in very good agreement with the experimental results.

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