

Analysis and Modeling of the Creep Behavior of the Thermostable PMR-15 Polyimide

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ABSTRACT: To establish the constitutive equations for carbon/PMR-15 composites and to determine the contribution of the matrix to the mechanical response, an analysis of the behavior of the PMR-15 polyimide alone has been performed under thermomechanical loading conditions corresponding to military aeroengine applications. In particular, the creep behavior of the material has been studied for several combinations of stress (0.3 to 0.7 σ_r), temperature (250 to 300°C), and time (5 and 1300 h). The creep compliance has been modeled in the field where the behavior is linear, using the Burgers model, together with a retardation time distribution for the viscoelastic region. The viscoelastic behavior is described satisfactorily by the Kohlrausch function, and combined with the Maxwell relation, enables the creep compliance of PMR-15 polyimide to be modeled in the temperature range considered for short loading times. For long times at 250°C and 0.3 σ_r , the polymer undergoes slow plastic deformation, but nevertheless, conserves its overall viscoelastic properties. There is no equivalence between time and temperature. The master curve, constructed graphically on the basis of linear viscoelastic strains, short times, and a reference temperature of 250°C, deviates quite significantly from the experimental curve at longer times. In the latter case, the creep compliance curve can be correctly fitted to a simple power law. © 1998 John Wiley & Sons, Inc. *J Appl Polym Sci* 69: 1983–1991, 1998

Key words: thermostable polymers; PMR-15 polyimide; thermomechanical behavior; creep; modeling

INTRODUCTION

The PMR-15 polyimide matrix resin has been designed to withstand the thermomechanical loading of composites usable over a wide range of temperatures. For example, carbon woven fabric/PMR-15 resin composites are employed for certain warm components in one of SNECMA's turbojet engines.

To analyze or predict the behavior of the composite material, it is necessary to know the behavior of each of the constituent components. In the case of components subjected to mechanical loading at high temperatures, it therefore appeared useful to determine the mechanical behavior of the thermoset resin under these conditions and, hence, to evaluate the contribution of the matrix to the overall response of the composite. In fact, the PMR-15 resin matrix in the composite is considered to be thermostable up to about 300°C, and the composite is designed for potential applications at temperatures between 250 and 300°C. To ensure that the polymer is capable of withstanding the applied

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loads over long periods without change of shape, it is therefore necessary to know its deformation capacity and limiting strain behavior. A convenient method of studying the behavior of PMR-15 polyimide is to apply constant tensile stresses for varying times at suitably chosen temperatures, and to analyze the resulting strains.

Because of the high temperatures that this type of resin can withstand, a simple characterization is not sufficient, and a thorough and detailed analysis of the behavior is necessary. Furthermore, because of the complexity of its structure, the behavior of PMR-15 resin might be expected to differ from that of other usual thermosets already studied. To determine the behavior of PMR-15 polyimide in the temperature range from 250 to 300°C, it was therefore decided to perform tests under different conditions of stress and time. The stress levels of interest could be accurately chosen based on tests previously performed on both the polymer alone and the composite, particularly in fatigue.^{1,2} For each stress, the loading time was taken as the mean time during which a real component is maintained under load in a military aircraft engine.

In the first part of the article, the creep results obtained for different stresses and temperatures are described and analyzed to determine the limits of the field in which the compliance can be described by a linear behavior model. This linear range ensures the integrity of the mechanical properties of the matrix when the composite is subjected to repeated loading. The viscoelastic behavior of the material is modeled by representing it as an assembly of analog elements. A simple model is first of all employed to represent the creep compliance at each temperature. Two other models are then used to precisely describe the short time behavior. The coefficients involved in these models are determined by graphical and numerical analysis.

In the second part of the article, the equivalence between time and temperature is studied, based on the results of short time tests. A master curve is constructed and compared to a creep curve established experimentally for longer times. The residual mechanical properties after creep are determined to study the aging behavior of the polymer. A creep compliance model is also developed to describe the long-term behavior.

TEST MATERIAL AND EXPERIMENTAL METHODS

Manufacture of the Polyimide

The test material was manufactured from a methanol solution of the PMR-15 reactive monomers,

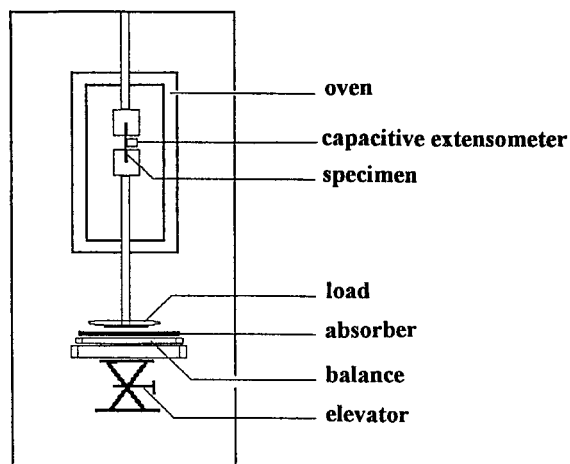


Figure 1 Schematic representation of the equipment for creep testing.

prepared by the STRUCTIL company (Vert-le-Petit, France). Resin sheets were produced by compression molding in a two-step process.³ These sheets were cured at 315°C under 6 MPa for 4 h. Standard ISO flat dumbbell test pieces were taken from these sheets, the profiles being produced by grinding. The thickness of the test pieces was equal to that of the molded sheets, i.e., between 2.5 and 3.5 mm. The specimen gage length was 30 mm long and 5 mm wide, enabling an extensometer to be attached. All the specimens were postcured for 16 h at 315°C after machining to reproduce the usual manufacturing conditions of composites. No particular further treatment was employed, the test pieces being stored at ambient temperature in the laboratory atmosphere (60% RH).

Equipment and Test Conditions

The creep tests were performed on a rig comprising a PYROX (Rambouillet, France) oven and an MTS (Minneapolis, USA) capacitive extensometer, together with a balance on which is placed the weight that applies the load to the specimen (Fig. 1).

The imposed load can be freed either rapidly or gradually, partly or completely. The specimen is mounted separately on a marble bed to ensure perfect alignment of the specimen and straining system. The minimum specimen clamping stress is determined experimentally so as to avoid squashing the head while ensuring the absence of slipping in the grips during application of the load. The axial capacitive extensometer is

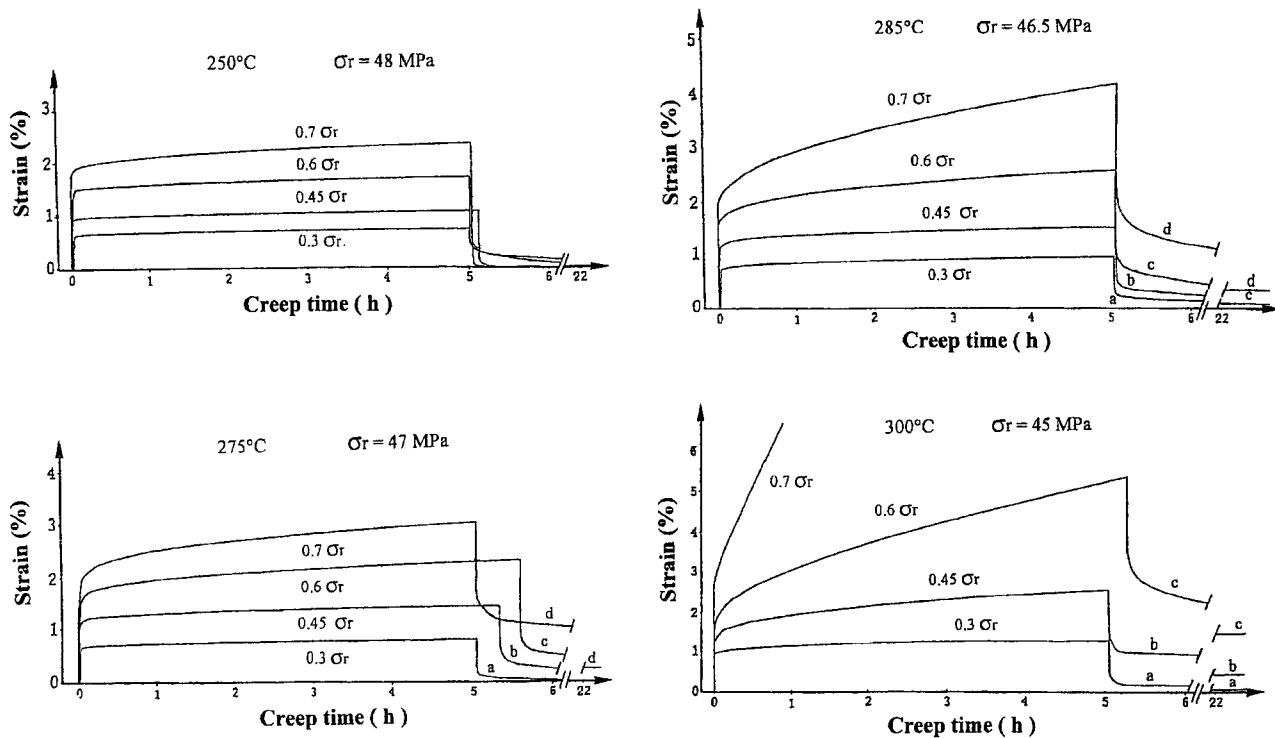


Figure 2 Strain and recovery of the PMR-15 polyimide postcured 16 h at 315°C (four levels of stress and four temperatures).

attached with variable pressure clips adjusted so that the knife edges do not damage the specimen surface, particularly at high temperature. The temperature control system is commanded by a thermocouple positioned directly on the test piece. The strain zero is set after a period of at least 10 min to allow the temperature of the specimen and the equipment to stabilize.

The damage observed in woven $[\pm 45]$ carbon/PMR-15 composites in fatigue² at 290 and 315°C becomes large when the loading cycle has a high maximum stress equals to $0.8 \sigma_r$, σ_r being the static tensile failure stress. If a constant stress of this level is applied to the polymer, creep rupture can occur rapidly. A maximum creep stress of $0.7 \sigma_r$ was therefore chosen. Based on creep tests already performed on amorphous polymers,^{4,5} it was decided that $0.3 \sigma_r$ would be a suitable minimum creep stress to apply. Two intermediate levels, corresponding to $0.45 \sigma_r$ and $0.6 \sigma_r$, were also selected. Apart from the two extreme temperatures of 250 and 300°C, two other values were chosen, 275 and 285°C, nearer to the maximum limit. The time during which the imposed load was applied was limited to 5 h, based on the average duration of loading of a real military aircraft engine component.

Prior to creep testing, the tensile failure stress σ_r was determined at the four temperatures. Over the temperature range considered, σ_r was found to be virtually constant at 47 MPa, demonstrating the remarkable thermal stability of PMR-15 polyimide compared to the properties of other resins commonly used in aeronautical engineering applications.³

ANALYSIS OF THE CREEP CURVES

Figure 2 shows the creep curves obtained for the four stress levels at each of the four temperatures with a single sample for each testing condition. However, several tests were doubled to confirm some experimental curves. The stresses are calculated in terms of the ultimate stress at each temperature. After 5 h of creep loading, or sometimes slightly more, the specimens were unloaded instantaneously to study the recovery behavior. All the creep curves have the same general form. The instantaneous elastic strain on loading is followed by a period of slow linear deformation between 1 and 5 h. These two regions are separated by a transition zone that is more or less pronounced depending on the stress and temperature. All the

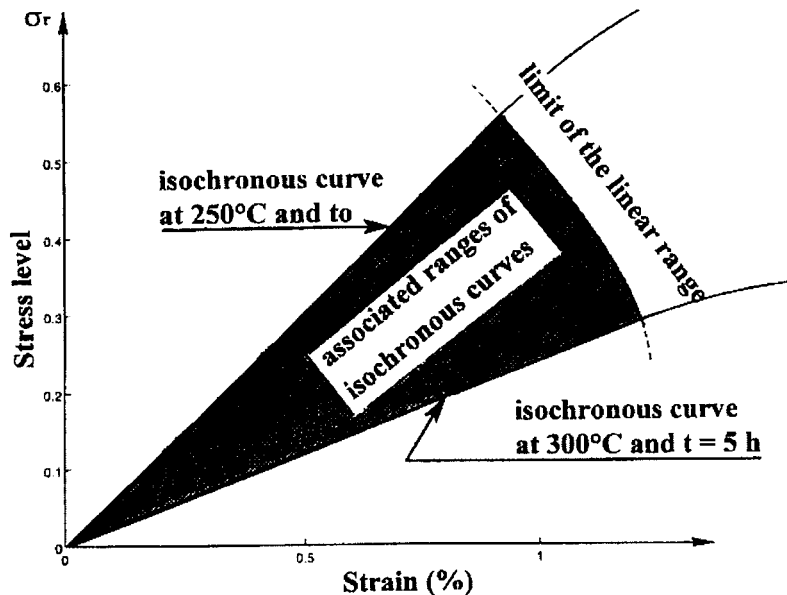


Figure 3 Complete mapping of isochronous stress–strain curves over both fixed time and temperature ranges (situation between 250 and 300°C to and $t = 5$ h).

recovery curves are also similar in form, with an instantaneous contraction of amplitude equal to or slightly less than the initial elastic strain, followed by a period of slow recovery, continued for 16 h after unloading. These two regions are separated by a transition zone of variable amplitude. A number of features are noteworthy: (a) at the three lowest temperatures, the strain after 5 h loading is no more than a few percent, whatever the applied stress level. (b) After 5 h at 300°C, the strain is significantly greater for a stress of $0.6 \sigma_r$, while rupture occurs after 1 h under a stress of $0.7 \sigma_r$. (c) On removal of the load, the material recovers completely at 250°C, whatever the stress level applied initially. (d) Only an applied stress of $0.3 \sigma_r$ ensures total recovery at each of the four temperatures. (e) The residual strain, when it exists 16 h after unloading, increases with temperature and initial stress level (compare the residual strains marked a, b, c and d in Fig. 2). After recovery for 1 week, the observed residual strains show no further change, and can, therefore, be considered to represent permanent deformations.

Because of the low residual strain levels, the differences in specimen section before and after testing are too small to be measured, even for the test piece, which broke at 300°C. For the sake of the present study, the stress levels will, therefore, be considered to be directly proportional to the applied loads.

The linearity of the creep behavior is expressed

by the proportionality between the stress and strain parameters, according to Boltzmann's superposition principle.⁶ This principle applies to relatively few polymers, because the observed creep strains are often large at the stress levels employed in the present study. Furthermore, these strains increase as the temperature rises. In contrast, in a thermoset resin such as PMR-15, loaded in its glass state after a postcure of 16 h at 315°C, the strains obtained are small. To define this linear range, it is convenient to plot isochronous curves from the raw creep data. Thus, the creep strains at arbitrary constant time values for each temperature can be plotted on stress-strain coordinates (Fig. 3). The isochronous curves are found to be linear up to certain stress and strain levels, whose average limiting values for the four temperatures are represented by the dotted line in Figure 3. The overall range of linearity under the conditions studied is represented by the shaded region. The linear range for PMR-15 resin is quite extensive, because the largest strains ($\approx 1\%$) correspond to high stress levels (up to $0.55 \sigma_r$). Indeed, although a maximum stress level of $0.7 \sigma_r$ was employed to determine the strain capacity of the material, this is extremely severe, and real composite components are typically loaded to no more than 0.4 to $0.5 \sigma_r$. At loading levels of $0.5 \sigma_r$ in quasi-isotropic composites, it is possible that the range of linear behavior of the matrix will never be exceeded.^{7,8}

In the framework of the present study, only the linear creep behavior of the resin has been modeled. Modeling of the non-linear creep behavior of PMR-15 polyimide is being undertaken in the case of $[\pm 45]$ oriented composites.⁹

Apart from modeling considerations, particularly with regard to creep compliance, the region of linearity is perfectly suited for studying the equivalence between time and temperature and its extrapolation for long times.

MODELING OF CREEP COMPLIANCE

All the creep curves obtained have the same form, which can be decomposed and represented approximately by common rheological models. Although the latter take no account of the polymer structure, they have the advantage of ideally describing a differential equation and are sufficient to define the linear viscoelastic behavior of polymers in general. During the description of the creep curves given above, it was shown that they can be broken down into three parts. Among the behavior types and corresponding models proposed in the literature,⁶ the three regions observed appear to best match the Burgers model. This one has been applied and a further refinement has been made by a more detailed consideration of the viscoelastic response of the material, to obtain a precise expression for the creep compliance.

Application of the Burgers Model

The Burgers model is a series combination of the Maxwell and Kelvin-Voigt models, the total strain ε_T being given by the general equation:

$$\varepsilon_T = \varepsilon_1 + \varepsilon_2 + \varepsilon_3$$

$$\varepsilon_T = \frac{\sigma}{E_1} + \frac{\sigma}{E_2} \left[1 - \exp\left(-\frac{E_2 t}{\eta_2}\right) \right] + \frac{\sigma t}{\eta_3}$$

where ε_1 and ε_3 are respectively the elastic and viscous strains represented by Maxwell model and ε_2 is the viscoelastic strain represented by the Kelvin-Voigt model. E_1 and E_2 are elastic moduli, η_2 and η_3 are viscosities, σ is the applied stress, and t is the creep time.

Applied to this study, the equation simplifies in two ways: 1) because the study of the PMR-15 resin is confined to the domain of linear behavior,

it is possible to define the proportionality between stress σ and strain ε by the creep compliance J . 2) The strain ε_2 includes both an elastic part and a viscous part associated with a retardation time $\tau = \eta_2/E_2$.

In the linear range, whatever the stress, the creep compliance as a function of time is then given by the expression:

$$J_t = J_1 + J_2 + J_3$$

$$J_t = \frac{1}{E_1} + \frac{1}{E_2} \left[1 - \exp\left(-\frac{t}{\tau}\right) \right] + \frac{t}{\eta_3}$$

However, a single retardation time is not sufficient to give an accurate description of the calculated viscoelastic compliance J_2 . This deficiency leads to select a generalized Kelvin-Voigt model describing compliance J_2 more precisely.

Models Corresponding to the Viscoelastic Compliance

If the stress is decomposed into n increments ($\sigma = \sigma_1 + \sigma_2 + \dots + \sigma_n$), the viscoelastic compliance J_2 represents the corresponding sum of compliance increments (sum of the different lattice movements), and can be simulated by a Kelvin-Voigt series model (generalized K-V model):

$$J_2 = \sum_{i=1}^{i=n} \frac{1}{E_i} \left[1 - \exp\left(-\frac{t}{\tau}\right) \right]$$

More generally, the sum of the discrete elements can be replaced by a continuous function of the retardation times $P(\tau)$. The compliance J_2 is then given in the form of an integral. To facilitate the description of the deformation at short times, it is convenient to represent this integral in logarithmic form. The creep compliance $J(t)$ gives:¹⁰

$$J_{(t)} = J_1 + J_2 + J_3 = \frac{1}{E_1} + \int_{-\infty}^{+\infty} P(\log \tau)$$

$$\times \left[1 - \exp\left(-\frac{t}{\tau}\right) \right] d(\log \tau) + \frac{t}{\eta_3}$$

This continuous J_2 function is very difficult to determine directly by mathematical treatment.¹¹ The simplest and also the most useful approximation is that given by Alfrey¹² as early as 1945. The

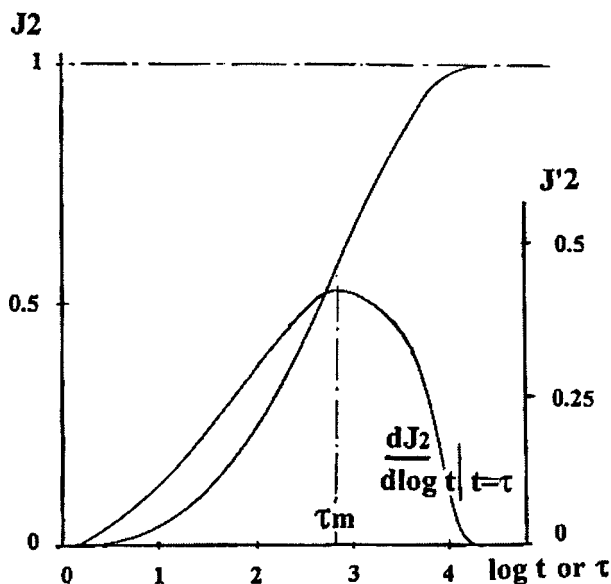


Figure 4 Normalized plot of J_2 viscoelastic compliance at 285°C and its J_2' derivative curve (retardation spectrum).

retardation time τ of the elastic part, retarded by the damping part and represented by the Kelvin-Voigt element, is considered to correspond at time t to the deformation of the macromolecular lattice represented by the experimental creep curve. If it is assumed, to a first approximation, that the retardation time is equal to the experimental time, the experimental curve then represents the geometrical locus of the retardation times for the series of Kelvin-Voigt elements. In fact, the derivative of the experimental J_2 compliance curve plotted on logarithmic time coordinates has a bell shape, called the retardation spectrum (Fig. 4).

Among the large number of representations of the retardation spectrum found in the literature,⁶ the most commonly used is the Gauss curve.¹¹ A Gaussian distribution of the retardation times was therefore applied in the present study but did not give satisfactory results. There are several possible reasons for this, including the irregular, albeit unimodal, profile of the retardation spectrum, its dissymmetrical form, skewed towards shorter times, and the lack of precision in the tails, particularly for the initial stages of creep. In contrast, the Kohlrausch function¹³ gives a suitable fit for the PMR-15 polyimide creep compliance data. This function, also known as the KWW function¹⁴ (Kohlrausch, Williams, Watts) is defined by the equation:

$$J_2 = \frac{1}{E_2} \left\{ 1 - \exp \left[- \left(\frac{t}{\tau_m} \right)^b \right] \right\} \quad \text{or}$$

$$b(\text{Ln } t - \text{Ln } \tau_m) = \text{Ln}[-\text{Ln}(1 - J_2 E_2)]$$

where b is the Kohlrausch coefficient and can take values between 0 and 1.

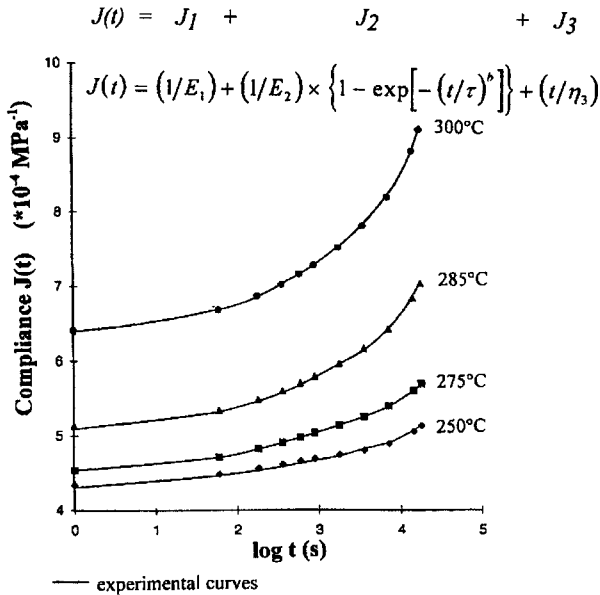
The integral of the viscoelastic compliance J_2 given above is here characterized by the KWW function, which is statistical in nature and generally represents a sigmoidal curve, such as either the experimental curve or a Gaussian distribution. The second term of the function is called the extended exponential.⁵ The double logarithmic representation as a function of time enables the value of the coefficient b to be determined for any value of $J_2 E_2 < 1$. There is a certain analogy between the parameters characterizing the Gaussian spectrum and those of the KWW function. If the retardation spectrum were perfectly Gaussian, the mean retardation time τ_m and the coefficient b determined by linear regression of the KWW function would give a symmetrical sigmoidal curve.¹⁵ These two parameters, b and τ_m , in the KWW function are adjusted based on the experimental data, and render the Gaussian curve asymmetrical, giving a better fit of the experimental results, particularly for short times. The experimental and theoretical compliance curves for the four temperatures in the linear range are represented in Figure 5. For short creep times (< 5 h), good agreement is obtained between the calculated and experimental compliances.

In short, the method the best adapted for representing the creep compliance of PMR-15 polyimide, based on application of the Burgers model to short times, is to graphically divide the experimental curve and to linearize the KWW function. The first step enables the parameters contained in J_1 and J_3 to be determined, while those involved in J_2 are defined by the second operation. The complete relation used to describe the creep compliance of PMR-15 resin for times less than 5 hours at high temperature is thus:

$$J_{(t)} = J_1 + J_2 + J_3$$

$$J_{(t)} = \frac{1}{E_1} + \frac{1}{E_2} \left\{ 1 - \exp \left[- \left(\frac{t}{\tau_m} \right)^b \right] \right\} + \frac{t}{\eta_3}$$

It should be noted that, in extreme cases of high stresses and temperatures, for which it becomes



- 300°C: $J(t) = 6.37 \times 10^{-4} + 1.4 \times 10^{-4} \times \left\{ 1 - \exp\left[-(t/1054)^{0.49}\right] \right\} + 7.7 \times 10^{-9} t$
- ▲ 285°C: $J(t) = 5.09 \times 10^{-4} + 1.06 \times 10^{-4} \times \left\{ 1 - \exp\left[-(t/967)^{0.48}\right] \right\} + 5.1 \times 10^{-9} t$
- 275°C: $J(t) = 4.5 \times 10^{-4} + 0.76 \times 10^{-4} \times \left\{ 1 - \exp\left[-(t/713)^{0.45}\right] \right\} + 2.5 \times 10^{-9} t$
- ◆ 250°C: $J(t) = 4.3 \times 10^{-4} + 0.44 \times 10^{-4} \times \left\{ 1 - \exp\left[-(t/235)^{0.44}\right] \right\} + 2.17 \times 10^{-9} t$

Figure 5 Experimental and calculated creep compliance of the polyimide PMR-15 for four temperatures.

sometimes difficult to define an oblique asymptote, and therefore impossible to suitably divide the experimental curve, the Burgers model is no longer applicable. The purely viscous domain no longer exists and the expression for $J(t)$ reduces to the terms J_1 and J_2 . In this case, knowing the strain ϵ_1 accurately, the parameters b and τ_m can be obtained only by numerical analysis of the modulus E_2 , after linear regression.

EQUIVALENCE BETWEEN TIME AND TEMPERATURE

The WLF theory¹⁶ (William, Landel, and Ferry) shows that, in the linear viscoelastic domain, a viscoelastic curve determined at a temperature T_1 can be transposed to another temperature T_2 by

an appropriate multiplication of the time scale. This translation is performed by applying a shift factor a_T at a reference temperature T_r , with a logarithmic time scale:

$$\log a_{T(T_r)} = \frac{-C_1(T - T_r)}{C_2 + (T - T_r)}$$

This known expression is suitable for a number of linear amorphous polymers at a reference temperature close to T_g , with values of the “universal” constants C_1 and C_2 determined from studies on polystyrene¹⁶ (17.4 and 51.6, respectively). However, although the compliance curves for PMR-15 polyimide were determined in the linear range close to T_g , this relation is not suitable for constructing a master curve using one of the experimental temperatures (250 to 300°C) as the reference. Nevertheless, it is indeed possible to translate the compliance curves with respect to one of them chosen as the reference. Thus, Figure 6 shows the master plot obtained by horizontally shifting each of the other curves by an appropriate factor onto the 250°C reference curve. The time during which the curves superimpose increases the closer the temperature to the reference value. This behavior was compared with the results of a long time creep test (1277 h), performed at 250°C as the master curve, and with a stress level of 0.3 σ_r to obtain the same compliance in the initial stages of creep. This loading time corresponds to about half the typical lifetime of composite components in a military aircraft engine. Regarding the temperature, 250°C can be considered to represent that of the permanent service regime. In fact, composite engine components are not always at 250°C during each aircraft mission, but if the cu-

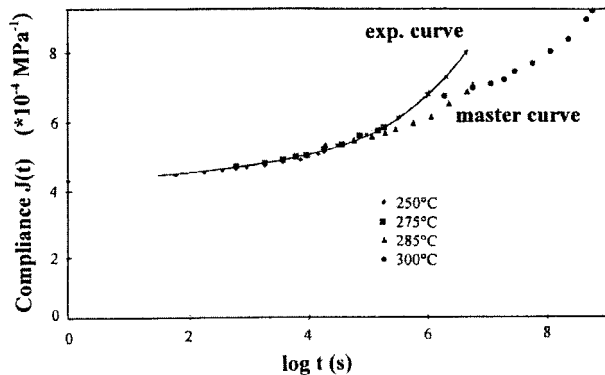


Figure 6 Master creep compliance curve at a reference temperature of 250°C. Comparison of experimental and master curves.

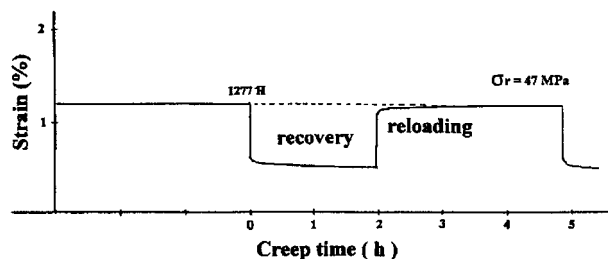


Figure 7 Strain evolution after unloading at 1277 h and then reloading (250°C/0.3 σ_r).

mulative time at this temperature is considered throughout the life of the engine, the duration amounts only to a few hundred hours. The conditions applied in the present study are thus particularly severe.

The experimental compliance curve obtained is superimposed on the master curve (Fig. 6). After about 30 h, the two curves diverge. Therefore, for longer creep times, the equivalence between time and temperature is not respected and the compliance model given above is not suitable. McCrum¹⁷ points out that the master curve should be constructed by both horizontal and vertical shifts to allow for aging of the polymer. However, because aging of polymers generally causes embrittlement, the experimental curve should fall below the master curve, whereas the situation can be seen to be the opposite for PMR-15. The material becomes more compliant with time. Nevertheless, PMR-15 might effectively be expected to be embrittled by aging at 250°C. To examine this possibility, additional tests were performed on the specimen crept for 1277 h at 250°C. The aim of these tests was to determine the behavior of the resin after unloading then reloading to the same stress level of 0.3 σ_r , and to analyze the residual rupture properties.

Figure 7 shows that the residual strain represents a large proportion of the total creep strain. After relaxation for a week, this residual strain was found to be permanent, and this was confirmed by a similar test on another specimen. After reloading to the same stress, the profile of the curve and the strain values ε_1 , ε_2 , and ε_3 are identical to those observed during the initial creep. The reloading is applied on a new gauge length (initial length + permanent deformation). This signifies that, when the permanent plastic strain is not considered, the viscoelastic behavior is identical to that in a short creep test. Moreover, this behavior is linear, because the previous maximum strain level (dashed line in Fig. 7) is reat-

tained with the previous permanent deformation, in accordance with Boltzmann's superposition principle.¹⁸ The material thus undergoes slow plastic deformation but conserves its viscoelastic properties.

No residual strain could be detected after initial unloading in the short time (5 h) creep test carried out at 250°C and 0.3 σ_r (Fig. 2), and this was true for all the other stress levels studied. At 275°C, measurable residual strain after recovery was observed only at a stress of 0.7 σ_r , or after longer loading times at lower stresses (Fig. 2). Thus, at 250°C and 0.3 σ_r , the time at which viscoplasticity occurs must correspond to the point where the experimental curve diverges from the master curve constructed solely on a linear viscoelastic basis, i.e., about 30 h (Fig. 6). After 1277 h of creep at 250°C, the ultimate tensile stress of the polymer σ_r is 47 MPa, equivalent to the value determined before creep. Similar values were obtained on specimens aged at 250°C without load for the same time (1277 h).

Thus, isothermal aging at 250°C, either without a load or under a stress of 0.3 σ_r , does not deteriorate the intrinsic properties of the polymer. Under these loading conditions, the polymer can thus be considered to be mechanically stable at 250°C. The study of the aging behavior of PMR-15 resin is presently being continued at temperatures up to 300°C,¹⁹ and among other things, should indicate the maximum temperature at which the material remains usable.

Although the deformation ε remains small at long loading times, the residual plastic strain after creep is proportionally quite large. The behavior of the polymer is therefore elastoviscoplastic in nature. Under the loading conditions shown in Figure 7, the behavior at long times can be described by the relation:

$$\varepsilon = \varepsilon_1 + \varepsilon_3 t^n.$$

The coefficients for the viscoplastic part can be determined from points taken from the experimental curve and over a logarithmic scale giving the equation of the calculated creep compliance:

$$J(t) = 4.3 \times 10^{-4} + 4.5 \times 10^{-6} t^{0.29}.$$

CONCLUSION

To determine the constitutive equations for carbon/PMR-15 composites and evaluate the contri-

bution of the mechanical response of the matrix, it appeared necessary to establish the behavior of the resin alone, under thermomechanical loading conditions representative of those encountered by military aero-engine components.

After having determined the ultimate tensile strength σ_r of PMR-15 over the temperature range 250–300°C, an analysis of the creep behavior was carried out in the following conditions: (a) short time loading (5 h) at stresses from 0.3 to 0.7 σ_r in the temperature range mentioned; (b) long time loading (1300 h) at 250°C and 0.3 σ_r .

Both of these represent severe conditions compared to those experienced by real engine components.

In all the creep test configurations (time, stress, and temperature), the strains remained small. The range in which the creep behavior is linear extends to relatively quite large stress values, and it was for this reason that modeling of the creep compliance was limited in this one. Inside this range, all the creep curves have the same form, which can be readily decomposed into elastic, viscoelastic, and viscous parts. The creep compliance was thus represented by a series of rheological elements and the model was successively refined using several methods. The initial Burgers model was improved by association with the generalized Kelvin-Voigt model, in which the distribution of retardation times is represented by the Kohlrausch function. The extended exponential term in this function has the advantage of correctly describing the non-Gaussian experimental retardation spectrum, particularly at short creep times. The Kohlrausch function provides a satisfactory representation of the viscoelastic compliance, and combined with the Maxwell relation, closely models the overall creep compliance at all temperatures.

It is not possible to establish an equivalence between time and temperature using WLF-type shift factors. The master curve constructed graphically, based on linear viscoelastic strains and a reference temperature of 250°C, diverges gradually but significantly from the experimental curve determined from long-time creep data (1300 h). In the conditions employed for the long-time creep loading (250°C and 0.3 σ_r), the polymer undergoes slow plastic strain but conserves its viscoelastic properties, leading to an overall long time behavior that is elastoviscoplastic in nature and can be correctly fitted to a simple power law.

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