A MECHANICS MODEL FOR THE DEFORMATION OF SWIRL-MAT COMPOSITES¹

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ABSTRACT

Swirl-mat polymeric composites are considered as representative materials for application in the automotive industry. This article summarizes a three-year long effort to characterize, model, and predict the deformation of these composites under stress, temperature, cyclic loadings, and the ingress of fluids. The effort involved several hundred characterization tests of various kinds and the formulation of a constitutive model based upon fundamental principles of irreversible thermodynamics and continuum mechanics [1]. The above-mentioned experiments and model development were performed interactively. Subsequently, the model was employed in a predictive manner to anticipate experimental data collected under complex load and temperature histories, as well as to predict long-term behavior from short-term data.

INTRODUCTION

The research summarized herein was conducted as a part of a larger project that aimed at the establishment of design criteria for swirl-mat polymeric composites, for application in the automotive industry. Additional tasks performed under the project are reported in companion articles presented in this conference.

We refer the reader to those articles regarding basic information on the material, its processing, and its potential application in the automotive industry. Suffice to say that, for the purpose of the present article, the material exhibits an overall scatter of $\pm 10\%$ in properties within a plaque and contains a void content $V_v = 5\%$.

The above scatter, which reflects sample-to-sample variability, must be accounted for prior to model development,

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because it can easily mask the effects of stress, temperature, and fluids on the response of the composite. That scatter is to be addressed in the context of design criteria as well.

Preliminary observations were performed to assess the basic features of the material's behavior. These observations utilized various NDE and microscopic techniques, including C-scan, SEM, confocal microscopy, in-situ optical microscopy under loading and unloading, and SPATE. It was concluded that failure progression in the material at hand occurs in the form of a multitude of distributed micro-cracks, that these micro-cracks occur between the individual fibers within the fiber strands² as well as the interfaces between strands and surrounding matrix. In addition, upon unloading, the above micro-cracks may not completely close back to the original configurations due to interfacial asperities, resulting in a permanent deformation.

The above observations indicated that for swirl-mat composites irreversibility due to fracturing should not be attributed to the propagation of dominant cracks. Instead, this phenomenon should be modeled within the framework of continuum damage theory.

Preliminary mechanical tests demonstrated the existence of creep under all stress levels, suggesting a response commensurate with viscoelastic behavior. Finally, it was discovered that it was possible to reduce much of the data scatter caused by sample-to-sample variability upon scaling deformations by the distinct initial stiffnesses of the individual coupons.

CREEP RESPONSE

The characterization of creep deformation was achieved by the application of constant stresses σ_0 for fixed durations t_0 and their subsequent removals, as sketched in Fig. 1a. The stress magnitudes σ_0 ranged from 14 and 124 MPa (2 and 18 ksi at equal intervals of 2 ksi), and the load durations t_0 were 1,

² Each strand consists of approximately 200 individual fibers.

10, 30, 100, 1000, and 10,000 minutes. Replicate tests were performed in most cases. Note, however, that all the foregoing ranges of t_0 are much shorter than typical failure times t_f . Upon unloading, recovery strains were recorded for time durations t_r , with $3t_0 \leq t_r \leq 10t_0$. Typical strain records are sketched in Fig. 1b, where ϵ_p denotes non-recoverable, permanent strain.

Figure 1: Schematic of stress input (a) for a creep test and the associated strain response (b).

The initial stiffness was determined during the loading stage, employing strains recorded at stress levels of 6 and 28 MPa (0.8 and 4 ksi), while "damaged stiffnesses" were determined during unloading, employing strains corresponding to 85% and 15% of σ_0 .

It was possible to express the creep compliance $D(t)$ in power-law form, namely $D(t) = D_0 + D_1 t^n$ resulting in

$$\epsilon(t) = (D_0 + D_1 t^n) \sigma_0 \quad 0 \leq t \leq t_0 \quad (1)$$

while, during recovery,

$$\epsilon_r(t) = D_1 [t^n - (t - t_0)^n] \sigma_0, \quad t > t_0. \quad (2)$$

Note that $\epsilon_r(t)$ necessitates a data fit that involves two parameters only (D_1 and n), whereas creep incorporated a third parameter D_0 . It is therefore advantageous to utilize recovery strains to guide data fitting.

The linear expressions (1) and (2) were noted to apply for stresses $0 < \sigma_0 < \sigma_d = 36$ MPa (5.2 ksi). At higher stress levels, strains were no longer directly proportional to stress due to the onset of irreversible damage, and their correlation was amplified by a stress-dependent factor $K_\sigma = D_d/D_0$, where D_d is the inverse of the stiffness evaluated during unloading. It was possible to express K_σ as follows:

$$K_\sigma(\sigma) = 1 + K_d \langle \sigma - \sigma_d \rangle \quad \sigma > \sigma_d \quad (3)$$

here $\langle \cdot \rangle$ are the McAulay brackets. Obviously, K_σ quantifies the amount of stress-induced damage in the composite. The strain-stress relation (1) is thus modified to read

$$\epsilon(t) = (D_0 + D_1 t^n) K_\sigma(\sigma) \sigma \quad 0 < t < t_0, \quad \sigma > \sigma_d \quad (4)$$

The parameters in Equations 1 and 3 are

$$8.4 \times 10^{-11} < D_0 < 11 \times 10^{-11} \text{ Pa}^{-1}$$

$$(5.8 \times 10^{-7} < D_0 < 7.4 \times 10^{-7} \text{ psi}^{-1})$$

$$1.0 \times 10^{-11} < D_1 < 2 \times 10^{-11} \text{ Pa}^{-1} (\text{min})^{-n}$$

$$(0.7 \times 10^{-7} < D_1 < 1.4 \times 10^{-7} \text{ psi}^{-1} (\text{min})^{-n})$$

$$0.08 < n < 0.09$$

$$K_d = 3.5 \times 10^{-3} \text{ MPa}^{-1} (2.4 \times 10^{-2} \text{ ksi}^{-1})$$

with time in minutes. The above ranges for D_0 , D_1 and n reflect the sample-to-sample variability.

It was also noted that the non-linear range ($\sigma > \sigma_d$) is associated with non-recoverable permanent strains ϵ_p . As may well be expected from the increased randomness of the aforementioned interfacial asperities, which are the underlying causes for ϵ_p , the records of ϵ_p exhibit much larger scatter than all other experimental data. Nevertheless, it was noted that ϵ_p correlated reasonably well with the largest strain $\epsilon_{\max}(t)$ which corresponded to Eq.(4).

The validity of the above results was verified by performing two-step stress loadings and comparing the resulting strain data with model predictions [2]. Good agreement was achieved as indicated in Figs. 2a and 2b.

Figure 2: Comparison between the model (---) and experimental data (o) for the case of a two-step stress history where a step stress σ_2 was applied after the unloading of the step stress σ_1 .

It should be noted that while it was possible to quantify damage through $K_{\sigma}(\sigma)$ for the short-term characterization tests, of durations t_0 , employed in the foregoing experimental program, it is obvious that damage continues to evolve with time t for $t \gg t_0$. It turned out that it was possible to express the evolutionary relation through a Kachanov-Rabotnov equation of the form [3]:

$$\dot{D}_d/D_0 = \left(\frac{D_d}{D_0} \right)^2 \left[\frac{\langle \sigma - \sigma_c \rangle}{A} \cdot \frac{D_d}{D_0} \right]^{\chi} \quad (5)$$

where the parameters A and χ are evaluated from time-to-failure data. Typically, for swirl mat composites, $\chi \approx 7$ and $A=260$ MPa (hours) $^{1/\chi}$, [3]. Consequently, the deformation process incorporates two time-dependent phenomena. The first, governed by polymeric creep is quantified by the exponent n in the power law expression (1), while the second—dominated by the evolution of damage—is related by the rate equation (5) with the initial values prescribed in (3). The strain-stress relation that accounts for both phenomena reads

$$\varepsilon(t) = \int_0^t D(t-t) \frac{d}{dt} (w(t)\sigma(t)) dt \quad (6)$$

where

$$w(t) = D_d(t)/D_0.$$

A typical result is shown in Fig. 3 where the creep response under a step stress $\sigma=83$ MPa (12 ksi) is predicted in the presence and absence of growing damage.

Figure 3: Model predictions for creep-damage behaviour up to failure (—) as compared with predicted viscoelastic creep alone (---) for 12 ksi at 75 °F.

CYCLIC LOADING

The deformation of swirl-mat polymeric composites under cyclic loading was investigated by subjecting test coupons to "saw tooth" patterns of stress histories, as sketched in Fig. 4a, and recording the corresponding strains, as shown schematically in Fig. 4b.

The test program involved values of $\sigma_{\max}=28, 56, 84$, and 104 MPa (4, 8, 12, and 16 ksi) as well as $R = \sigma_{\min}/\sigma_{\max}$ ratios of 0, 0.1, and 0.3. At least three

replicate specimens were tested at each pair of values of $\sigma_{\max}$ and R .

Figure 4: Schematic diagrams for cyclic testing (a) "saw tooth" stress input (b) strain response.

The cyclic loads were applied at a frequency of 1 Hz³ and tests were terminated after 2000 cycles. The cyclic stress-strain data were reduced to determine the increases in compliance (= reductions in stiffness) with ongoing cycling. Compliances were evaluated as the ratios of $\Delta\varepsilon/\Delta\sigma$ during the down-loading stages of the cyclic inputs, where $\Delta\sigma = 0.5\sigma_{\max}$ and $\Delta\varepsilon = \varepsilon(0.85\sigma_{\max}) - \varepsilon(0.35\sigma_{\max})$, and recorded at each individual cycle during the first ten cycles, then at ten-cycle intervals during the first one hundred cycles, and subsequently at hundred-cycle intervals. Results are plotted in Figs. 5 and 6 for $R=0.1$ and 0.3, respectively.

Figure 5: Compliance ratio versus cycles for $R=0.1$.

³ It was noted that lower frequencies, as well as intermittent pauses between sequential loadings, did not affect the response in any noticeable manner [4].

Figure 6: Compliance ratio versus cycles for R=0.3.

The foregoing results for D_u/D_0 vs. cycles provide a measure for damage growth with number of cycles (i). They can be related by the empirical expression as follows:

$$w_i = D_u^{(i)} / D_0 = 1 + (\Delta\omega + \hat{\Delta}\omega \cdot i^m) H(\sigma - \sigma_d) \quad (7)$$

where

$$\Delta\omega = (-0.00030 * \sigma_{\max}^2 + 0.01420 * \sigma_{\max} - 0.05372)$$

$$*(R+0.9)$$

and

$$\hat{\Delta}\omega = (-8.60592E-05 * \sigma_{\max}^2 + 3.03826E-03 * \sigma_{\max} - 1.20423)$$

$$*(1.08 - 0.8 * R).$$

Note that expression (5) differs from Eq. (7), since the latter applies to damage induced by cyclic loading whereas the former pertains to damage growth under sustained loads. This disparity resembles that which occurs in the case of failure driven by a dominant crack, where fracture mechanics applies to growth under sustained loading while the empirical Paris' law may relate the evolution under cyclic loading. Due to the exceedingly higher complexity that occurs in swirl mat composites, where damage develops in the form of a multitude of microcracks, the relationship between sustained and cyclic loading induced failures is more remote than for failure that is driven by a single dominant crack.

The strain due to cyclic load is evaluated by substituting expressions (1), namely $D(t) = D_0 + D_1 t^n$, and (7) into Eq. (6). Upon accounting for the cyclic time dependence of $\sigma(t)$ in the saw tooth pattern shown in Fig. 5a and performing the integration indicated in Eq. (6), it is possible to predict the strain outputs. Omitting the details, the resulting predictions are compared against recorded values in Figs. 7 and 8. These figures exhibit the maximal and minimal strains vs. number of cycles. Note the good correlation between data and model predictions.

Figure 7: Recorded and predicted strains during cyclic loading for R=0.1.

Figure 8: Recorded and predicted strains during cyclic loading for R=0.3.

TEMPERATURE EFFECTS

The effects of temperature on the deformation of swirl-mat polymeric composites were evaluated from short-term creep and recovery data of the kind exhibited in Figs. 1a and 1b. Experiments were conducted at stress levels of $\sigma_0 = 28, 56, 69$, and 84 MPa (4, 8, 10, and 12 ksi) and at temperatures $T = 75^\circ\text{F}$, 135°F , 190°F , and 250°F . In all circumstances, load was maintained for a duration $t_0 = 30$ min. with recovery strains recorded during subsequent periods of $3t_0 \leq t_r \leq 10t_0$. At least three replicate tests were performed at each of the above levels of σ_0 and T .

To separate the effects of stress from those of temperature, the creep and recovery strains of all specimens were recorded at first under $\sigma_0 = 28 \text{ MPa}$ (4 ksi) and at $T \approx 75^\circ\text{F}$ (room temperature), with $t_0 = 30$ minutes, in order to establish their "linear range signatures". Subsequently, the coupons were divided into four groups, denoted by $j = 1, 2, 3, 4$ with each group subjected to preconditioning stresses of $\sigma_o = (\sigma_{oj} + 7 \text{ MPa (1ksi)})$ for a five minute duration, where $\{\sigma_{01}, \sigma_{02}, \sigma_{03}, \sigma_{04}\} = \{28, 56, 69, 84 \text{ MPa (4, 8, 10, 12 ksi)}\}$. All the latter tests were performed at room temperature.

The reason for the above "stress conditioning" is that in this manner it was ascertained that the stress induced levels of internal damage due to preconditioning, exceeded those that corresponded to σ_{oj} alone in subsequent tests at elevated temperatures.

At this stage, each of the aforementioned groups of specimens were subdivided into four distinct sub-groups, each consisting of at least three coupons, which were tested under stresses of σ_{oj} ($j = 1, 2, 3, 4$) at temperatures of 75°F , 135°F , 190°F , and 250°F . All stresses were maintained for a duration of 30 min. with subsequent recovery strains recorded until their leveling off with time.

Typical results are shown in Figs. 9, 10, and 11. It turned out that the data shown in Figs. 9-11 could be reduced by a format that relates stress and temperature effects on creep in a product form.

Figure 9: Creep response as a function of stress at 135°F .

Accordingly

$$\varepsilon(t; \sigma; T) = \{D_o(\sigma; T) + D_1(\sigma)[t/a_T(T)]^n\} \sigma \quad (8)$$

where

$$D_o(\sigma; T) = F(\sigma)R(T)D_o^{75^\circ\text{F}}, \quad D_1(\sigma) = F(\sigma)D_1^{75^\circ\text{F}}$$

with

$$F(\sigma) = 1 + k_d \langle \sigma_o - \sigma_d \rangle$$

and

$$R(T) = 1 + 0.0025(T - 75^\circ\text{F})$$

In Eq. (8) $a_T(T)$ is a well known time-temperature "shift factor" function. For the material at hand, it was found that, to within a reasonable degree of approximation,

$$\text{Log}[a_T(T)] = 6.1695 - 0.0988T + 0.0002T^2 \quad (9)$$

Figure 10: Creep response as a function of stress at 190°F .

Figure 11: Creep response as a function of stress at 250°F .

SUMMARY

This article presented a methodical approach which, employing an exhaustive experimental program in tandem with basic principles of mechanics and thermodynamics, resulted in the construction of a predictive constitutive model for the mechanical response of swirl-mat polymeric composites.

While transitions to other materials will certainly involve variations in details, the methodology should nevertheless remain valid for composite materials that exhibit similar characteristic features in their thermomechanical response. In this regard it may be of interest to point out the methodology presented herein is currently applied successfully to random-mat, chopped fiber composites.

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