**OAK RIDGE
NATIONAL
LABORATORY**

RECEIVED
NOV 29 1999
OSTI

**Some Aspects of the Deformation
Response of Swirl-Mat Composites**

M. Elahi and Y. J. Weitsman

**MANAGED AND OPERATED BY
LOCKHEED MARTIN ENERGY RESEARCH CORPORATION
FOR THE UNITED STATES
DEPARTMENT OF ENERGY**

ORNL-27 (3-96)

This report has been reproduced from the best available copy.

Reports are available to the public from the following source.

National Technical Information Service
5285 Port Royal Road
Springfield, VA 22161
Telephone 703-605-6000 (1-800-553-6847)
TDD 703-487-4639
Fax 703-605-6900
E-mail orders@ntis.fedworld.gov
Web site <http://www.ntis.gov/ordering.htm>

Reports are available to U.S. Department of Energy (DOE) employees, DOE contractors, Energy Technology Data Exchange (ETDE) representatives, and International Nuclear Information System (INIS) representatives from the following source.

Office of Scientific and Technical Information
P.O. Box 62
Oak Ridge, TN 37831
Telephone 423-576-8401
Fax 423-576-5728
E-mail reports@adonis.osti.gov
Web site <http://www.osti.gov/products/sources.html>

Reports produced after January 1, 1996, are generally available via the DOE Information Bridge.

Web site <http://www.doe.gov/bridge>

This report was prepared as an account of work sponsored by an agency of the United States Government. Neither the United States Government nor any agency thereof, nor any of their employees, makes any warranty, express or implied, or assumes any legal liability or responsibility for the accuracy, completeness, or usefulness of any information, apparatus, product, or process disclosed, or represents that its use would not infringe privately owned rights. Reference herein to any specific commercial product, process, or service by trade name, trademark, manufacturer, or otherwise, does not necessarily constitute or imply its endorsement, recommendation, or favoring by the United States Government or any agency thereof. The views and opinions of authors expressed herein do not necessarily state or reflect those of the United States Government or any agency thereof.

DISCLAIMER

Portions of this document may be illegible in electronic image products. Images are produced from the best available original document.

Engineering Technology Division

**SOME ASPECTS OF THE DEFORMATION RESPONSE
OF SWIRL-MAT COMPOSITES**

M. Elahi and Y. J. Weitsman*

*MAES Department, University of Tennessee, Knoxville.

October 1999

Prepared by the
OAK RIDGE NATIONAL LABORATORY
Oak Ridge, Tennessee 37831-6285
managed by
LOCKHEED MARTIN ENERGY RESEARCH CORP.
for the
U.S. DEPARTMENT OF ENERGY
under contract DE-AC05-96OR22464

CONTENTS

	Page
LIST OF FIGURES.....	v
PREFACE.....	vii
ABSTRACT.....	1
1. INTRODUCTION.....	1
2. STATIC RESPONSE.....	1
3. SOME ASPECTS OF MODELING AND RESPONSE OF DAMAGE COUPLED WITH VISCOELASTIC CREEP.....	3
4. SOME OBSERVATIONS REGARDING CREEP RUPTURE AND DAMAGE GROWTH.....	4
5. CYCLIC LOADING.....	6
6. TEMPERATURE EFFECTS.....	9
7. THE ABSORPTION OF DISTILLED WATER AND ITS EFFECTS ON DEFORMATION.....	10
7.1 ABSORPTION OF DISTILLED WATER.....	10
7.2 EFFECTS OF DISTILLED WATER ON DEFORMATION	10
8. CONCLUDING REMARKS.....	11
9. REFERENCES.....	11

LIST OF FIGURES

Figure		Page
1	Photograph of the "dry" random glass fiber swirl-mat preform	12
2	Confocal laser scanning images of swirl-mat coupons	12
3	Relation between stiffness and strength over a temperature range of $-40^{\circ}\text{C} \leq T \leq 121^{\circ}\text{C}$	13
4	The stress-strain response of a multiply gaged specimen under monotonically increasing load	13
5	Creep rupture data for a multiply gaged specimen under $\sigma = 130$ MPa at room temperature	14
6	Creep-to-failure response under gage 2 in Fig. 5	14
7	Schematic diagram for cyclic testing	15
8	Compliance enhancement at $R = 0$ and various stresses	15
9	Compliance enhancement at $R = 0.1$ and various stresses	16
10	Compliance enhancement at $R = 0.3$ and various stresses	16
11	Factor Δw_0 vs stress	17
12	Factor Δw_1 vs stress	17
13	Factor m vs stress	18
14	Unscaled fatigue deformation data at $R = 0$ and various stress levels	18
15	Unscaled fatigue deformation data at $R = 0.1$ and various stress levels	19
16	Unscaled fatigue deformation data at $R = 0.3$ and various stress levels	19
17	Scaled fatigue deformation data at $R = 0$ and various stress levels	20
18	Scaled fatigue deformation data at $R = 0.1$ and various stress levels	20
19	Scaled fatigue deformation data at $R = 0.3$ and various stress levels	21
20	Creep strain at various stress levels at temperature $T = 57^{\circ}\text{C}$	21
21	Creep strain at various stress levels at temperature $T = 88^{\circ}\text{C}$	22
22	Creep strain at various stress levels at temperature $T = 121^{\circ}\text{C}$	22
23	Factor $F(T)$ vs temperature T	23
24	The shift factor function a_T vs temperature T	23
25	Temperature history vs time	24
26	Creep response of a multiply gaged coupon at 23°C under $\sigma_0 = 28$ MPa	24
27	Data and model predictions under both stress and fluctuating temperature at a selected location	25
28	Weight gain of distilled water in swirl-mat composites at various temperatures	25
29	Approximate values of the diffusion coefficient D vs the inverse of temperature (T in degrees Kelvin)	26
30	Creep data for dry and wet specimens (submerged in 22°C distilled water, weight gain = 2.2%) at 28 MPa	26
31	Creep data for dry and wet specimens (submerged in 52°C distilled water, weight gain = 2.4%) at 28 MPa	27
32	Creep data for dry and wet specimens (submerged in 82°C distilled water, weight gain = 4.9%) at 28 MPa	27
33	Wet/dry compliance ratios vs immersion temperature in distilled water	28

PREFACE

This report presents a portion of a larger investigation undertaken at Oak Ridge National Laboratory (ORNL) on the durability of polymeric composites for automotive structural applications. The ORNL project, which is entitled "Durability of Lightweight Composite Structures for Automotive Applications," is sponsored by the U.S. Department of Energy Office of Advanced Automotive Technologies.

The report supplements previously published results on the response of swirl-mat composites that appeared in ORNL/TM-13281 titled *The Visco-Damage Mechanical Response of Swirl-Mat Composites* by L. V. Smith and Y. J. Weitsman. Additional background information related to this report can be found in Refs. 1-4 listed in Sect. 9.

The main purpose of this report is to demonstrate the applicability of a theoretically based methodology to model and predict the response of a class of randomly reinforced polymeric composites under consideration for automotive applications. Theoretical modeling and experimental investigations progressed interactively throughout this investigation, demonstrating the essential need for the systematic establishment of a comprehensive database conjointly with theory.

SOME ASPECTS OF THE DEFORMATION RESPONSE OF SWIRL-MAT COMPOSITE

M. Elahi and Y.J. Weitsman

ABSTRACT

This report concerns the mechanical response of random glass fiber strand swirl-mat/urethane matrix composite under static and cyclic loads as well as under elevated temperatures and exposure to distilled water. The article presents an extensive amount of experimental data as well as predictions based on a coupled damage/viscoelastic constitutive formulation generated to model the specific behavior of the material at hand. Damage evolution relations are derived from an empirical relationship. This work extends previously published results. It is shown that the current model has the capability to predict long-term response on the basis of short-term data and account for time-varying stresses and temperatures.

1. INTRODUCTION

A swirl-mat polymeric composite consists of preforms of glass-fiber strands that are swirled in a random circular pattern, which are injected with urethane resin by a rapid SRIM process. A typical duration of the SRIM process is about 20 s. A photo of the "dry" preform is shown in Fig. 1. This composite is a model material for application in the automotive industry.

This report presents information regarding the room-temperature response of the above composite under static and cyclic loading, as well as its static response at elevated temperatures and due to immersion in distilled water. Also included are an outline of a proposed model, which couples viscoelasticity with continuum damage theory to account for observed behavior, and comparisons between model predictions and experimental data.

An extensive program of nondestructive evaluation (NDE) was undertaken to inspect as-received and mechanically loaded coupons. This program, which is detailed in earlier works,^{1,2} consisted of c-scan, SPATE, scanning electron microscope (SEM), confocal microscopy, and optical microscopy. Those inspections showed conclusively that swirl-mat polymeric composites contain a multitude of randomly distributed initial micro-cracks that are located between the individual fibers within the fiber strands, as well as between the strands and surrounding urethane matrix.

In response to mechanical loading, the above micro-cracks grow in a distributed fashion within the fiber strands and along the strand/matrix interfaces. A typical observation regarding the distributed manner of the aforementioned "damage" can be seen in the confocal laser scanning microscope images shown in Fig. 2.

2. STATIC RESPONSE

The swirl-mat composite at hand exhibits a high degree of scatter in mechanical properties as well as a certain degree of orthotropy. In the roll direction (0°) the stiffness and strength were noted to vary from 7.6 to 11.4 GPa and from 115 to 194 MPa, respectively, while transversely to the roll direction (90°) these quantities were noted to vary between 10.5 and 12.1 GPa and from 125 to 219 MPa, respectively.

These variations are attributed to the inherent random nature of the swirl-mat preform that is compounded by the rapidity of the resin injection and curing process, which results in nonuniform wetting of the glass fibers. The ensuing material inhomogeneity is typified by a nonuniform spatial distribution of void volume fraction that may reach the value of $V_v \sim 5\%$ in certain locations.

Nevertheless, it is interesting to note the existence of an obvious correlation between stiffness and strength which persists over the temperature range of $-40^{\circ}\text{C} \leq T \leq 121^{\circ}\text{C}$ (Ref. 3), as shown in Fig. 3. This rather remarkable relationship between stiffness and strength suggests that both may be governed by the same controlling factors, namely, spatial variability in fiber content and orientation, and a localized level of pre-existing internal damage, such as voids and disbands. This notion is also supported by experimental observations of varying stress-strain response recorded at several stations along a single coupon subjected to a monotonically increasing load, as shown in Fig. 4. Note that failure occurs under the strain gage associated with the location of lowest modulus. A better insight into the failure process is provided in Fig. 5. This figure exhibits strain-to-failure (creep rupture) data recorded in a multigaged coupon subjected to a step load of 138 MPa at room temperature. Note the wide variation in initial stiffnesses ($8.36 \leq E \leq 10.21$ GPa) recorded at several stations along the specimen, with failure again occurring at the location of lowest stiffness (8.36 GPa). It is illuminating to notice the sharp upswing in strain prior to failure at the aforementioned location, with a trace of such upswing noted at a nearby station. These observations suggest that the failure region is localized about the failed cross-section. It is also worth noting that similar localized upswings in strain, upon the approach of failure, were noted earlier under cyclic loading.^{3,*}

To recapitulate on some previous works,^{1,2} it is mentioned that the experimental characterization program consisted of the application of step stresses of magnitudes σ_o , followed by their removal after time-intervals t_o , namely

$$\sigma(t) = \sigma_o [H(t) - H(t - t_o)] \quad , \quad (1)$$

with $\sigma_o = 28, 41, 55, 69, 83, 97$, and 110 MPa for durations of $t_o = 1, 10, 30, 60, 100, 1000$ min and, at some of the lower stress values up to $10,000$ min. Creep and recovery strains were recorded during $0 < t < t_o$ and $t_o < t < kt_o$ ($4 < k < 10$), respectively. Replicate specimens were employed in many circumstances.² Note that the above durations t_o are much shorter than the failure time t_f , that, on the average, was tentatively related to stress by the empirical relation $\sigma = 16.6t_f^{-0.008}$ (Ref. 3), with t_f in hours.

The experimental creep data can be expressed in the following power-law form:

$$\varepsilon(t) = \begin{cases} K_{\sigma}(\sigma)(D_o + D_1 t^n)\sigma & 0 < t < t_o \\ K_{\sigma}(\sigma)D_1[(t^n - (t - t_o)^n)]\sigma + \varepsilon_p & t_o < t < kt_o \end{cases} \quad (2)$$

In Eq. (2) $t_o \ll t_f$, and time is in minutes. The values of D_o ranged over $8.12 \times 10^{-5} < D_o < 11.02 \times 10^{-5} \text{ MPa}^{-1}$ and those of D_1 between $5.16 \times 10^{-6} < D_1 < 11.79 \times 10^{-6} \text{ MPa}^{-1} \text{ min}^{-n}$, averaging at $D_o = 9.12 \times 10^{-5} \text{ MPa}^{-1}$ and $D_1 = 8.00 \times 10^{-6} \text{ MPa}^{-1} \text{ min}^{-n}$. This scatter is commensurate with the variations in recorded stiffnesses as noted earlier. Only small variations were noted in the power n about its average value of 0.086. The quantity $K_{\sigma}(\sigma)$, which expresses scaled, stress-induced level of damage is given by²

$$K_{\sigma}(\sigma) = 1 + k_{\sigma} \langle \sigma - \hat{\sigma} \rangle \quad , \quad (3)$$

where $k_{\sigma} = 0.0035 \text{ MPa}^{-1}$, $\hat{\sigma} = 36 \text{ MPa}$, and $\langle \rangle$ are the McAuly brackets. Namely $\langle x \rangle = x$, for $x > 0$ and $\langle x \rangle = 0$, for $x < 0$.

Finally, ε_p denotes a damage-induced permanent strain, which is due to the internal roughness of opposing micro-crack surfaces that prevents their complete closure upon load removal. Such roughness was observed by means of optical microscopy.^{1,2} As may be expected, the scatter in ε_p is much larger than that of all other measured properties due to the highly random nature of the abovementioned roughness.

*In that case strains were recorded by means of a single extensometer. Upswings in strain data just prior to failure were noted in those specimens that failed within the extensometer's span.

Nevertheless, it was possible to assess ε_p by correlating it with the maximum strain $\varepsilon_{\max}$ experienced by the material during its loading history, according to²

$$\varepsilon_p = K_\varepsilon (\varepsilon_{\max} - \varepsilon_d) , \quad (4)$$

where $K_\varepsilon = 0.051$ and $\varepsilon_d = 4100 \mu\varepsilon$. For the stress histories given by Eq. (1), $\varepsilon_{\max} = \varepsilon(t_o^-)$. The factor $K_\sigma(\sigma)$ was determined individually for each test coupon by evaluating its initial compliance D_o within the linear range, namely under stress $\sigma_o < \hat{\sigma}$, subsequently reloading the coupon to $\sigma_o > \hat{\sigma}$ and evaluating its unloading compliance D_u . In this manner $K_\sigma(\sigma) = \frac{D_u(\sigma, \sigma > \hat{\sigma})}{D_o(\sigma, \sigma < \hat{\sigma})}$ accounts for damage-induced enhancement of compliance (= stiffness reduction).

The power-law form in Eq. (2), which is limited to times $t \ll t_f$, does not predict the time of failure t_f . To extend the range of validity of the foregoing form it is necessary to consider $K_\sigma(\sigma, t)$, thus accounting for the time-dependent evolution of damage caused by the growth of internal micro-cracks. The representation of such growth under sustained constant loading, which was indeed observed through optical microscopy, will be addressed in Sect. 3.

3. SOME ASPECTS OF MODELING AND RESPONSE OF DAMAGE COUPLED WITH VISCOELASTIC CREEP

As a background for the subsequent sections, this subsection provides an outline of a model based on fundamental principles of continuum mechanics and irreversible thermodynamics that accounts for a viscoelastic response coupled with distributed damage. Attention is focused on mechanisms typical of swirl-mat and random chopped-mat polymeric composites. Accordingly, it is assumed that molecular-level motions within the polymeric phase dominate the creep, while damage is associated with micro-cracks at the fiber/matrix interfaces. In view of the disparate geometric scales involving the above phenomena, it is considered that molecular motion may occur even at constant levels of damage. Consequently, the internal variables that represent molecular motions can vary with time even at fixed values of the internal variables that account for damage. The complete formulation is given elsewhere.^{4,5}

Let Φ, S, T , and q_i denote the Gibbs free energy, entropy, temperature, and heat flux, respectively; then the reduced entropy inequality reads

$$-\dot{\Phi} - \varepsilon_{ij} \dot{\sigma}_{ij} - s\dot{T} - \frac{q_i T_{,i}}{T} \geq 0 . \quad (5)$$

Let γ_r ($r = 1, \dots, N$) denote scalar valued internal variables that represent the degrees of freedom of polymeric molecular motion and ω_{abcd} , chosen as a fourth-rank symmetric tensor,^{*} account for the continuously distributed damage. A suitable form for the aforementioned coupled response can be obtained by letting $\Phi = \Phi(\sigma_{ij}, \gamma_r, \omega_{abcd}, T)$ (Refs. 4 and 5). A systematic manipulation of Eq. (5) involves the following steps. Time is introduced into the response through the assumption that "resistance to molecular motion" (conjugate forces Γ_r) is caused by a viscouslike internal mechanism within the polymer, namely

$$\Gamma_r = -\frac{\partial \Phi}{\partial \gamma_r} = a_{rq} \dot{\gamma}_q . \quad (6)$$

Damage is introduced into the response through the assumption that the undamaged compliances S_{ijkl}^o are enhanced and modified by an appropriate tensor valued function of ω_{abcd} . Thus,

^{*}In Ref. 5, ω is taken as a second-rank symmetric tensor.

$$S_{ijkl} = S_{ijps}^0 P_{pskl}(\omega_{abcd}) \quad (7)$$

It can be shown^{4,5} that for isotropic materials and a one-dimensional case Eqs. (5)–(7) can lead to the following strain-stress expression:

$$\epsilon(t) = \int_{0^-}^t D(t-\tau) \frac{d}{d\tau} \left(\frac{\sigma}{1-\omega} \right) d\tau \quad (8)$$

In Eq. (8) $D(t)$ denotes the linear viscoelastic compliance, and $1/(1-\omega)$ is the substantially reduced form of $P_{pskl}(\omega_{abcd})$ in Eq. (7). Note that $\sigma/(1-\omega)$ is the familiar “effective stress” encountered in continuum damage mechanics. However, since damage grows with time, the integrand in Eq. (8) involves two time-dependent variables, $\sigma(\tau)$ and $\omega(\tau)$. In the context of the previous section it is possible to identify the short time value of $1-\omega$ with D_0/D_u .

It is still necessary to complement the above considerations with a damage growth law expressing $\dot{\omega}_{abcd}$, which in the present case would require that $1-\omega = K_G^{-1}(\sigma; t)$. Though there exists a theoretical framework for relating damage growth rate $\dot{\omega}_{abcd}$ by means of a scheme somewhat analogous to viscoplasticity,⁶ this formulation involves an additional set of functions and parameters that become intractable for the material at hand with its highly scattered data. It was therefore expedient to complement expression (8) with the empirical Kachanov-Robotnov expression:

$$\dot{\omega} = \left[\frac{\langle \sigma - \hat{\sigma} \rangle}{A(1-\omega)} \right]^\chi \quad (9)$$

Equation (9) contains two new material constants, A and χ . The threshold value of $\hat{\sigma}$ is 36 MPa as before.

The evolution of $\omega = \omega(t; \sigma)$ is now obtained by solving the differential Eq. (9), with the initial condition $1-\omega(0) = K_G^{-1}(\sigma)$. Subsequently, it is possible to evaluate $\epsilon(t)$ up to failure time t_f and compare it with long-term creep data. The time t_f is another experimental quantity corresponding to failure at $\omega = \omega_f < 1$ (Ref. 6). It is thus possible to “tune” the quantities A and χ in Eq. (9), as well as t_f with creep to failure data. Preliminary estimates suggested that typical values for the swirl-mat composite at hand may be $A = 250$ and $\chi = 7$ and $\omega_f = 0.4$. Obviously, uncertainties remain because failure times are widely scattered.

4. SOME OBSERVATIONS REGARDING CREEP RUPTURE AND DAMAGE GROWTH

While creep rupture per se falls outside the scope of this investigation,* this report presents additional, detailed data that were collected from multiply gaged samples in order to explore some features of damage growth in the composite considered herein.

The experiments utilized several 29-cm-long straight edge coupons and 20-cm-long, dog-bone shaped samples, with strains recorded between five and seven equally spaced locations over the coupons' lengths, with an additional strain recorded for each sample by means of an extensometer.

Typical results are shown in Fig. 5, where numbers 2 through 5 refer to the strain gage location, starting from the top of the test specimen. The numbers on the right side provide the value of initial stiffness, as

* Creep-rupture experiments were performed elsewhere at ORNL under this project.

recorded within the linear stress range. Note the wide spread in strain data, which is consistent with the results shown in Fig. 4. More significantly, the location of failure is associated with the position of lowest stiffness, with localization ranging within approximately 2 cm about the failed cross section. It can also be seen that strain rises sharply with time prior to failure at $t = t_f = 686$ min.

Focusing attention on the strain recorded at the failed location, it was attempted to relate its time dependence to growing damage, in addition to the effect of stress. As a first step the above strain was related by means of the power-law given in Eq. (2), which should well fit the short-time data. ** A very good fit, up to $t^* = 140$ min, was indeed established by utilizing $D_o(\sigma) = 9.18 \times 10^{-5} \text{ MPa}^{-1}$, $D_1(\sigma) = 1.30 \times 10^{-5} \text{ MPa}^{-1} \text{ min}^{-n}$ and $n = 0.12$. Beyond t^* creep data exceeded values predicted by the power-law expression, as shown in Fig. 6. The excessive strain was attributed to damage growth.

Let $g(t)$ denote the time-dependent portion of damage growth. Thus, for $0 < t < t_f$ one has

$$\epsilon[t; \sigma, g(t)]/\sigma = D_o(\sigma) + D_1(\sigma)t^n + D_o(\sigma)g(t) + D_1(\sigma) \int_0^t (t-\tau)^n g'(\tau) d\tau, \quad (10)$$

where $g(0) = 0$.

The last two terms on the right-hand side of Eq. (10) should account for the excessive strain noted in Fig. 6.

Inspection of Fig. 6 suggests that $g(t)$ should contain a combination of slowly varying and rapidly varying functions of time. This motivates the following representation:

$$g(t) = \hat{A}t^p + \hat{B}t^q, \quad (11)$$

with $q \gg p$.

Substitution of Eq. (11) into the last two terms on the right side of Eq. (10) yields, after some manipulations,[†]

$$\begin{aligned} D_o(\sigma)g(t) + D_1(\sigma) \int_0^t (t-\tau)^n g'(\tau) d\tau \equiv D_o(\sigma) \left[A \left(\frac{t}{t_f} \right)^p + B \left(\frac{t}{t_f} \right)^q \right] \\ + D_1(\sigma) t_f^n \left[A \frac{\Gamma(n+1)}{1+n\Psi(p+1)} \left(\frac{t}{t_f} \right)^{n+p} + B \frac{\Gamma(n+1)}{1+n\Psi(q+1)} \left(\frac{t}{t_f} \right)^{n+q} \right]. \quad (12) \end{aligned}$$

In Eq. (12) Γ and Ψ denote the gamma and digamma functions, respectively.

A reasonable data fit for the effect of time-dependent damage growth was achieved by selecting, $p = 2$ and $q = 40$, $A = 0.12$, and $B = 0.26$. The resulting fit is shown in Fig. 6.

The strain data exhibited in Fig. 6 suggest that for $0 < t < t^* \sim 140$ min, creep is dominated by visco-elastic polymeric motions as modeled by the power-law expression. Subsequently, deformation is enhanced by time-dependent microcrack growth that encounters several arrest mechanisms, as attested by the varying slopes in the strain-time data. The final stage is associated with an extremely rapid damage growth, which corresponds to crack coalescence and fiber breakage.

As remarked earlier, traces of the phenomenon shown in Fig. 6 are noted at the neighboring gages (3 and 4) as may be seen in Fig. 5. Furthermore, note that at those neighboring locations, strains are predicted by the power-law expression (2) for times nearly up to failure. Consequently, it was possible to conclude that failure is localized within a region of ~ 2 cm about the failed cross section.

** Recall that Eq. (2) reflects the experimental observation that for short times D_o and D_1 do not depend on load duration.

[†] These manipulations include a two-term Taylor series expansion of $\Gamma(p+n)$ about $\Gamma(p)$ and similarly for $\Gamma(q+n)$.

5. CYCLIC LOADING

This section expands a previous work¹ to a wider range of R ratios $R = \sigma_{\min}/\sigma_{\max}$, amplitudes $\sigma_{\max}$, and number of cycles. Although a similar methodology is employed here, the present formulation differs in detail.

Consider a "sawtooth" stress input as shown schematically in Fig. 7(a), with a resulting strain output sketched in Fig. 7(b). In the current study we had $\sigma_{\max} = 55, 83$ and 110 MPa, with $R = 0, 0.1$, and 0.3 .

Also, the cycle duration was $2a = 1$ s.

Damage η_i was associated with the cyclically recorded compliance enhancement, S_i , which was evaluated during the unloading stages according to

$$\eta_i = \frac{S_i}{D_o} = \frac{1}{D_o} \frac{\left[\epsilon_i(0.85\sigma_{\max}) - \epsilon_i(0.35\sigma_{\max}) \right]}{0.5\sigma_{\max}} \quad (13)$$

In the above expression D_o is the initial compliance in the linear range of response, and $i = 0, 1, \dots, N = 2000$ denotes the cycle number.*

The experimental program utilized at least three replicate specimens for each set of R and $\sigma_{\max}$ values, for a total of more than 27 coupons. Coupon dimensions were $20 \times 2.5 \times 0.3$ cm. Prior to cyclic testing all coupons were subjected to step loading within the linear range in accordance with Eq. (1) and with $t_o = 30$ min, followed by 90 min of recovery time. Creep and recovery data were recorded to determine the specific parameters D_o and D_1 for each specimen. A common value of $n = 0.086$ was employed throughout, and strains were measured by extensometry.

It was possible to express η_i by the following empirical, power-law relation:

$$\eta_i = 1 + \Delta w_o + \Delta w_1 i^m \quad i = 0, 1, \dots, N, \quad (14)$$

where Δw_o , which depends on $\sigma_{\max}$ alone, accounts for the damage caused during the first loading, while Δw_1 and m , that depend on both $\sigma_{\max}$ and R , represent the cyclic-induced damage. In Eq. (14) i denotes the cycle number, as before. Representations of η_i , Δw_o , Δw_1 , and m , are shown in Figs. 8–13. Employing a "best fit" subroutine, one obtains

$$\begin{aligned} \Delta w_o &= -0.1178 + 1.0067\sigma_{\max}, \\ \Delta w_1 &= 0.0082 + 0.0083\sigma_{\max} - 0.0371R + 0.0138R\sigma_{\max}, \\ m &= 0.1624 + 0.0517\sigma_{\max} - 0.0905R, \end{aligned} \quad (15)$$

where $\sigma_{\max}$ is in MPa.

Recorded values of $\epsilon_{\max}$ and $\epsilon_{\min}$ vs the number of cycle i ("raw data") are shown in Figs. 14–16. Note the wide scatter in the above data. It turns out that the scatter in the foregoing data can be drastically reduced upon rescaling all deformations by the factors $E_{ok}/\bar{E}_o$, where E_{ok} is the initial stiffness of the individual coupon, and $\bar{E}_o$ is the average of the initial stiffnesses of all coupons employed in the cycling testing program. The scaled results are shown in Figs. 17–19. Note the most significant reduction in data scatter that resulted from the elimination of the sample-to-sample stiffness variability. Remarkably, the scaled data in Figs. 17–19 exhibit consistent trends with both $\sigma_{\max}$ and R , unlike the raw data shown in Figs. 14–16.

*Note that $\eta = 1/(1 - \omega)$ of the previous section.

In the viscoelastic formulation employed here it is advantageous to consider the cycle to cycle increments of damage. Thus,

$$\Delta\eta_i = \eta_i - \eta_{i-1} = \Delta w_1 [i^m - (i-1)^m] \quad i \geq 1 \quad (16)$$

In view of the fact that most damage occurs near peak load levels, it is assumed herein that all cyclic damage is focused at $\sigma = \sigma_{\max}$. This assumption greatly simplifies the subsequent mathematical computations. It also utilizes the only form of available damage data, as given in Eq. (13), because no information on damage growth *during* each load cycle is obtainable. Consequently, one has

$$\eta(t_N) = 1 + \Delta w_o H(t) + \sum_{i=1}^N \Delta\eta_i H[t - (k + 2i - 1)a] \quad (17)$$

Consider now the sawtooth stress history shown in Fig. 7(a). For computational purposes it is advantageous

to introduce $\hat{t} = t - t_o$, and $\hat{\sigma} = \sigma - \sigma_L$, whereby the stress rate is $r = \frac{\sigma_L}{t_o} = \frac{\sigma_u - \sigma_L}{a} = \frac{\hat{\sigma}_u}{a}$, where

$t_o = ka$.

The stress-time variation can be expressed by

$$\hat{\sigma}(\hat{t}) = \begin{cases} r(\hat{t} - 2ia) & ia < \hat{t} < (2i+1)a \\ -r[\hat{t} - 2(i+1)a] & (2i+1)a < \hat{t} < 2(i+1)a \end{cases} \quad i = 0, 1, \dots \quad (18)$$

resulting in

$$\sigma(t) = r[H(t) - H(t - t_o)] + r_o H(\hat{t}) + \hat{\sigma}(\hat{t}) H(\hat{t}) \quad (19)$$

Upon a straightforward manipulation, one obtains

$$\frac{d\sigma}{dt} = r[H(t) - H(t - t_o)] + \frac{d\hat{\sigma}(\hat{t})}{d\hat{t}} H(\hat{t}) \quad (20)$$

thereby,

$$\frac{d\sigma}{dt} = r[H(t) - H(t - t_o)] + \begin{cases} r & ia < \hat{t} < (2i+1)a \\ -r & (2i+1)a < \hat{t} < 2(i+1)a \end{cases} \quad i = 0, 1, \dots \quad (21)$$

Employment of Eq. (8) together with the power-law form (2), results in the present context in the following expression:

$$\epsilon(t) = D_o \eta(t) \sigma(t) + D_1 \int_0^t (t - \tau)^n \frac{d}{d\tau} [\sigma(\tau) \eta(\tau)] d\tau \quad (22)$$

Upon use of Eqs. (14)–(22), with Eq. (17) rewritten in terms of $\hat{t}$, and performing several tedious integrations one obtains

$$\epsilon_j^{\min} = \epsilon[t = ka + 2(j+1)a] = \epsilon[\hat{t} = 2(j+1)a] = I_1 + I_2 + I_3 + I_4 + I_5 \quad j = 0, 1, \dots \quad (23)$$

where

$$\begin{aligned}
I_1 &= D_o \sigma_L \eta_{j+1} , \\
I_2 &= \frac{D_1 r}{n+1} \left\{ [2(j+1)a + ka]^{n+1} - [2(j+1)a]^{n+1} \right\} , \\
I_3 &= D_1 \sum_{i=0}^j \frac{r}{n+1} \left\{ -[2(j+1)a - (2i+1)a]^{n+1} + [2(j+1)a - 2ia]^{n+1} \right\} \eta_{i+1} , \\
I_4 &= D_1 \sum_{i=0}^j \frac{r}{n+1} \left\{ [2(j+1)a - 2(i+1)a]^{n+1} - [2(j+1)a - (2i+1)a]^{n+1} \right\} \eta_{i+1} , \\
I_5 &= D_1 [2(j+1)a]^n \sigma_L \eta_o + D_1 \sigma_L \sum_{i=0}^j \left\{ [2(j+1)a - (2i+1)a]^n \right\} (\eta_{i+1} - \eta_i) .
\end{aligned}$$

For

$$\varepsilon_j^{\max} = \varepsilon[t = ka + (2j+1)a] = \varepsilon[\hat{t} = (2j+1)a] = J_1 + J_2 + J_3 + J_4 + J_5 \quad j = 1, 2, \dots \quad (24)$$

where

$$\begin{aligned}
J_1 &= D_o \sigma_u \eta_j , \\
J_2 &= \frac{D_1 r}{n+1} \left\{ [ka + (2j+1)a]^{n+1} - [(2j+1)a]^{n+1} \right\} , \\
J_3 &= D_1 \sum_{i=0}^j \frac{r}{n+1} \left\{ -[(2j+1)a - (2i+1)a]^{n+1} + [(2j+1)a - 2ia]^{n+1} \right\} \eta_i , \\
J_4 &= D_1 \sum_{i=1}^j \frac{r}{n+1} \left\{ -[(2j+1)a - 2ia]^{n+1} - [(2j+1)a - (2i-1)a]^{n+1} \right\} \eta_i , \\
J_5 &= D_1 [(2j+1)a]^n \sigma_L \eta_o + D_1 \sigma_L \sum_{i=1}^j \left\{ [(2j+1)a - (2i+1)a]^n \right\} (\eta_i - \eta_{i-1}) .
\end{aligned}$$

Note, however, that in view of Eq. (4), it is necessary to add the permanent component ε_p to $\varepsilon_{\min}$.

The predicted results for $\varepsilon_{\max}$ and $\varepsilon_{\min}$ vs number of cycles i are shown by the solid lines in Figs. 17-19. Note the excellent agreement with the scaled deformation curves. Predictions do depart from data in those cases where failures occurred prior to 2000 load cycles. The obvious cause for the above discrepancy is due to the fact that expression (14) does not contain a failure criterion similar to that implied in Eq. (9).

6. TEMPERATURE EFFECTS

The effects of temperature on creep were investigated at $T = 57^\circ\text{C}$, 88°C , and 121°C at stresses $\tilde{\sigma} = 28, 55, 69$, and 83 MPa, extending the previous room temperature data at $T = 23^\circ\text{C}$ (Refs. 1, 2). Prior to elevated temperature testing at stresses $\tilde{\sigma}$ that lie within the nonlinear range of response, each specimen was subjected to three preliminary tests at room temperature: (1) The application of a step-stress of 28 MPa, followed by its removal after 30 min, and recording creep and recovery strains for a total period of 120 min. The purpose of this test was to establish the relative stiffness of the coupon at hand to scale out the variability in this parameter. (2) The application of an elevated stress $\sigma = \tilde{\sigma} + 7$ MPa for a period of 5 min followed by 50-min recovery. The purpose of this test was to "stress condition" the coupon to delimit the instantaneous damage caused by loading into the nonlinear range. (3) The application of step-stress $\sigma = \tilde{\sigma}$ for a period of 30 min followed by 300-min recovery. The purpose of this test was to establish the nonlinear creep response of the coupon at hand at room temperature. In this manner it was conceivably possible to separate out the imprints of the elevated stress $\sigma = \tilde{\sigma}$ from the subsequent effects of elevated temperature.

At least three replicate specimens were tested at each level of stress and temperature. Results, scaled by the same factors $E_{ok}/\bar{E}_o$ as in the previous section, are shown in Figs. 20–22.

It turned out that the above results could be expressed as follows:

$$\varepsilon(t; \sigma; T) = D_o(\sigma; T)\sigma + D_1(\sigma) \left[\frac{t}{a_T(T)} \right]^n + \alpha \Delta T, \quad (25)$$

where $D_o(\sigma, T) = D_o K_\sigma(\sigma) F(T)$ and $D_1(\sigma) = D_1 K_\sigma(\sigma)$ with $K_\sigma(\sigma)$ given in Eq. (3).

The temperature-dependent functions $F(T)$ and $a_T(T)$ were derived by fitting the experimental values plotted in Figs. 23 and 24, which were obtained upon reducing the data shown in Figs. 20–22 in accordance with Eq. (25). The averaged values shown in Figs. 23 and 24 (which exhibit data scatter as well) were represented by the following expressions:

$$F(T) = 1 + 0.00443 (T - 23^\circ\text{C}), \quad (26)$$

$$\log[a_T(T)] = 3.2855 - 0.1676T + 0.0007T^2, \quad (27)$$

with T in degrees Celsius.

For fluctuating stresses and temperatures, the time-temperature analogy, when employed together with superposition, yields in the present case

$$\varepsilon(t; \sigma; T) = K_\sigma[\sigma(t)] \left\{ D_o F[T(t)]\sigma(t) + D_1 \int_0^t [\xi(t) - \xi(\tau)]^n \frac{d\sigma(\tau)}{d\tau} d\tau \right\} + \alpha \Delta T, \quad (28)$$

where

$$\xi(u) = \int_0^u \frac{dv}{a_T[T(v)]}.$$

A verification test was performed in the linear range of stress with $\sigma = \sigma_o[H(t) - H(t - t_o)]$, where $\sigma_o = 25$ MPa and $t_o = 30$ min, and under a monitored temperature fluctuation as shown in Fig. 25. For the material at hand $\alpha = \mu\text{e}/^\circ\text{C}$ and $\Delta T = T - 23^\circ\text{C}$.

Preliminary creep and recovery data were collected at room temperature with five strain gages located at equally spaced positions along a coupon 20 cm long. These are shown in Fig. 26. Note the wide disparity in deformation over the length of the coupon that, similarly to Fig. 4, exhibits direct correlation with variability in local stiffnesses. Predictions were made with a specific value of D_o that corresponded to the value recorded at a specific gage location in Fig. 26. Creep and recovery data are collected by that gage at the fluctuating temperature shown in Fig. 27. These results are in good agreement with a prediction calculated according to Eq. (28). In the present circumstance that equation reduces to

$$\epsilon(t; T) = \left\{ D_o F(t) + D_1 [\xi(t)]^n \right\} \sigma + \alpha \Delta T \quad (29)$$

In addition, Fig. 27 provides a comparison between data and linear elastic analysis, demonstrating the limitations of the latter approach.

7. THE ABSORPTION OF DISTILLED WATER AND ITS EFFECTS ON DEFORMATION

This section provides information on the absorption of distilled water in swirl-mat composite coupons and the effects of such absorption on the stress-strain behavior of the material.

7.1 ABSORPTION OF DISTILLED WATER

The absorption of distilled water was recorded by periodic measurement of weight gain data in standard test coupons that were immersed in baths, which were maintained at three controlled levels of temperature: 22°C, 52°C, and 82°C. Weight gain measurements were performed at approximately equal intervals on the square root of timescale, which allowed for direct comparisons against predictions of classical linear (Fickian) diffusion theory. In most cases the test coupons were predried prior to immersion in distilled water until attaining equilibrium dry weights (typically, this predrying period lasted for about 7 weeks). Typical results are shown in Fig. 28, where the ORNL data were obtained for "as-received," rather than predried, samples. Note that, in contrast to predictions of classical diffusion theory,⁷ none of the weight gain data approached equilibrium levels. Furthermore, the sharp upswing of the 82°C data is an unmistakable indicator of permanent, irreversible fluid-induced damage in the polymeric composite.⁸

It was possible to assess the magnitude of the fluid's diffusion coefficient D at all three levels of temperature by utilizing the short-term weight gain data and comparing them with classical diffusion predictions. Upon employment of a well-established methodology,⁹ the values of $D = D(T)$ can be evaluated and plotted in the standard format vs $1/T$ (T in degrees Kelvin). Results are exhibited in Fig. 29. Upon comparing these results against analogous values for epoxy-resin composites, it was concluded that the diffusion coefficient for the swirl-mat composite at hand exceeds the commonly encountered values by about one order of magnitude.

7.2 EFFECTS OF DISTILLED WATER ON DEFORMATION

The effects of distilled water on the creep and recovery response of the swirl-mat composites were evaluated by subjecting predried coupons to a step stress of 28 MPa followed by its removal at time $t_o = 30$ min and recording strains for a total period of $4t_o$. Subsequently subgroups of the above mentioned coupons were immersed in distilled water at temperatures of 22°C, 52°C, and 82°C, respectively. The immersion times were ~3600 h, 1200 h, and 900 h until attaining weight gain values of 2.2%, 2.4%, and 4.9% for the foregoing three levels of temperature, respectively. All coupons were then cooled to 22°C and subjected to the same stress history as in their predried state. Creep and recovery strains were recorded and compared against their dry state levels. The wet coupons were then redried and subjected again to the original stress history to observe any lingering effects of past exposures to distilled water. Such effects were indeed noted

in many cases through the observations of enhanced levels of creep and recovery strains, as compared with the originally dry circumstance.

Results for immersion at 22°C are shown in Fig. 30, while those for immersion at 52°C are exhibited in Fig. 31 and those at 82°C are shown in Fig. 32. All results demonstrate that exposure to distilled water enhances the deformation of swirl-mat composites. However, although this enhancement is undoubtedly more pronounced with increasing exposure temperature, the current data base does not suffice to quantify this temperature dependence due to excessive scatter in the experimental results, as shown in Fig. 33.

8. CONCLUDING REMARKS

This work employed previously obtained short-time data^{1,2} on the static response of swirl-mat polymeric composites as well as a constitutive formulation to represent their coupled damage/viscoelastic behavior.⁴ The above data were augmented by additional experimental results to exhibit the correlation between observed scatter in stiffness and failure localization. Subsequently, the constitutive formalism was utilized to relate experimental data for deformation under cyclic loading at several fixed levels of $\sigma_{\max}$ and R ratios as well as to predict the response of the foregoing polymeric composite at several elevated temperatures, including fluctuating temperature histories. In all circumstances stress levels included both linear and nonlinear range of material response.

The main practical implications of the current work are that they demonstrate the capability of the model to predict long-term response on the basis of short-term data and its ability to account for time-varying stress and temperature histories.

9. REFERENCES

1. L. V. Smith and Y. J. Weitsman, "Inelastic Behavior of Randomly Reinforced Polymeric Composites and Cyclic Loading," *Mechanics of Time-Dependent Materials*, 1, 293–305 (1998).
2. L. V. Smith and Y. J. Weitsman, "The Visco-Damage Mechanical Response of Swirl-Mat Composites," *Int. J. Fracture*, 92(3), 201–220 (1999).
3. J. M. Corum et al., *Durability-Based Criteria for an Automotive Structural Composite, Part 2. Background Data and Models*, Lockheed Martin Energy Research Corp., Oak Ridge National Laboratory, ORNL-6391, February 1998.
4. Kh. Abdel-Tawab and Y. J. Weitsman, "A Coupled Viscoelasticity/Damage Model with Application to Swirl-Mat Composites," *Int. J. Damage Mechanics*, 7, 351–380 (1998).
5. Kh. Abdel-Tawab and Y. J. Weitsman, "A Strain-Based Formulation for the Coupled Viscoelasticity/Damage Behavior," University of Tennessee Report MAES97-3.0-CM, July 1997 (to appear in *J. Appl. Mechanics*).
6. J. Lemaitre and J.-L. Chaboche, *Mechanics of Solid Materials*, English translation, Cambridge University Press, 1985.
7. J. Crank, *The Mathematical Theory of Diffusion*, Oxford University Press, Oxford, 2nd edition, 1975.
8. Y. J. Weitsman, "Effects of Fluids on Polymeric Composites—A Review," University of Tennessee Report MAES-5.0-CM, August 1998 (to appear as Chap. 12 in Vol. 2 of *Comprehensive Composite Materials*, Elsevier Science Pub.).
9. C.-H. Shen and G. S. Springer, "Moisture Absorption and Desorption of Composite Materials," pp. 15–33 in *Environmental Effects on Composite Materials*, G. S. Springer, ed., Technomic Pub. Co., 1980.

Fig. 1. Photograph of the "dry" random glass fiber swirl-mat preform.

(a) virgin specimen

(b) after 7 days at 60% of σ_{ult}

Fig. 2. Confocal laser scanning microscope images of swirl-mat coupons.

Fig. 3. Relation between stiffness and strength over a temperature range of $-40^{\circ}\text{C} \leq T \leq 121^{\circ}\text{C}$ (Ref. 3).

Fig. 4. The stress-strain response of a multiply gaged specimen under monotonically increasing load.

Fig. 5. Creep rupture data for a multiply gaged specimen under $\sigma = 138$ MPa at room temperature. Numbers marked by # correspond to gage location (starting from the top of the coupon). These are accompanied by the varying values of stiffnesses, evaluated at the above locations.

Fig. 6. Creep-to-failure response under gage 2 in Fig. 5. Data (ooo), prediction of "best fit" power-law (---), and prediction accounting for damage growth (---).

a. Saw tooth stress input

b. Strain response

Fig. 7. Schematic diagram for cyclic testing.

Fig. 8. Compliance enhancement at $R = 0$ and various stresses. Data scatter and model predictions (---).

Fig. 9. Compliance enhancement at $R = 0.1$ and various stresses. Data scatter and model predictions (---).

Fig. 10. Compliance enhancement at $R = 0.3$ and various stresses. Data scatter and model predictions (---).

Fig. 11. Factor Δw_0 vs stress.

Fig. 12. Factor Δw_1 vs stress.

Fig. 13. Factor m vs stress.

Fig. 14. Unscaled fatigue deformation data at $R = 0$ and various stress levels.

Fig. 15. Unscaled fatigue deformation data at $R = 0.1$ and various stress levels.

Fig. 16. Unscaled deformation data for $R = 0.3$ and various stress levels.

Fig.17. Scaled fatigue deformation at $R = 0$ and various stress levels. Data scatter and model predictions (—).

Fig. 18. Scaled fatigue deformation data at $R = 0.1$ and various stress levels. Data scatter and model predictions (—).

Fig. 19. Scaled fatigue deformation data at $R = 0.3$ and various stress levels. Data scatter and model predictions (—).

Fig. 20. Creep strain at various stress levels at temperature $T = 57^\circ\text{C}$.

Fig. 21. Creep strain at various stress levels at temperature $T = 88^\circ\text{C}$.

Fig. 22. Creep strain at various stress levels at temperature $T = 121^\circ\text{C}$.

Fig. 23. Factor $F(T)$ vs temperature T .

Fig. 24. The shift factor function a_T vs temperature T .

Fig. 25. Temperature history vs time.

Fig. 26. Creep response of a multiply gaged coupon at 23°C under $\sigma_0 = 28$ MPa.

Fig. 27. Data and model predictions under both stress and fluctuating temperature at a selected location. Elastic prediction is shown for comparison.

Fig. 28. Weight gain of distilled water in swirl-mat composites at various temperatures. (Note: ORNL data are for "as-received" samples that contain about 0.8% of moisture.)

Fig. 29. Approximate values of the diffusion coefficient D vs the inverse of temperature (T in degrees Kelvin). Swirl-mat composites immersed in distilled water.

Fig. 30. Creep data for dry and wet specimens (submerged in 22°C distilled water, weight gain = 2.2%) at 28 MPa.

Fig. 31. Creep data for dry and wet specimens (submerged in 52°C distilled water, weight gain = 2.4%) at 28 MPa.

Fig. 32. Creep data for dry and wet specimens (submerged in 82°C distilled water, weight gain = 4.9%) at 28 MPa.

Fig. 33. Wet/dry compliance ratios for specimens preimmersed in distilled water at various temperatures (average values with scatter bands). Values of weight gain w , in percentages, are indicated.

INTERNAL DISTRIBUTION

1. R. L. Battiste	15. J. M. Starbuck
2. R. G. Bowman	16. P. A. Sklad
3. C. R. Brinkman	17. C. D. Warren
4-7. J. M. Corum	18-30. Y. J. Weitsman
8. W. G. Craddick	31. G. T. Yahr
9. S. Deng	32. S. Simunovic
10. D. L. Erdman	33. R. E. Ziegler
11. J. G. Hansen	34. Central Research Library
12. L. D. Klett	35. ORNL Laboratory Records-RC
13. R. E. Norris	36-37. ORNL Laboratory Records-OSTI
14. M. B. Ruggles	

EXTERNAL DISTRIBUTION

- 38-66. E. M. Hagerman, Automotive Composite Consortium, General Motors, 30500 Mound Road, I-6, Box 9055, Warren, Michigan 48090-9055.
67. J. M. Henshaw, Department of Mechanical Engineering, The University of Tulsa, 600 S. College Avenue, Tulsa, Oklahoma 74104-3189.
68. G. A. Holmes, National Institute of Standards and Technology, Bldg. 224, Room B116, MS: Room B108, Gaithersburg, Maryland 20899.
69. K. Liechti, Engineering Mechanics Research Laboratory, Department of Aerospace Engineering and Engineering Mechanics, The University of Texas at Austin, Austin, Texas 78712.
70. D. Oplinger, AAR-431, FAA William J. Hughes Technical Center, Atlantic City International Airport, New Jersey 08405.
71. T. A. Reinhart, The University of Dayton Research Institute, 300 College Park Drive, Dayton, Ohio 45469-0130.
72. W. Ren, AFRL/MLLN, 2230 Tenth St., Bldg. 655, Rm. 23, WPAFB, OH 45433-7817.
73. G. Sandgren, Owens Corning Science & Technology Centre, 2790 Columbus Road, Route 16, Granville, Ohio 43023-1200.
74. R. A. Schapery, Engineering Mechanics Research Laboratory, Department of Aerospace Engineering and Engineering Mechanics, The University of Texas at Austin, Austin, Texas 78712.
75. C. R. Schultheis, National Institute of Standards and Technology, Building 224, Room A209, Gaithersburg, Maryland 20899.
76. T. D. Seagrave, Bayer Corporation, 100 Bayer Road, Pittsburgh, Pennsylvania 15205.
77. L. V. Smith, Washington State University, School of Mechanics and Materials Engineering, Pullman, Washington 99164-2920.
- 78-81. J. A. Carpenter, U.S. Department of Energy, 1000 Independence Avenue, SW, Washington, DC 20588.
82. P. G. Patil, U.S. Department of Energy, 1000 Independence Avenue, SW, Washington, DC 20588.
83. M. Rowllins, U.S. Department of Energy, Oak Ridge Site Office, Oak Ridge, Tennessee 37831.