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THE TIME-TEMPERATURE BEHAVIOR OF A UNIDIRECTIONAL GRAPHITE/EPOXY COMPOSITE

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ABSTRACT

A testing program to determine the time-temperature response of unidirectional T300/934 graphite/epoxy materials is described. Short-term (15 min. duration) creep test results of strip tensile specimen with the load at various angles to the fiber direction and at various temperature levels are reported. It is shown that the material is elastic at all temperature levels when the fiber is in the load direction. On the other hand, when the load is transverse to the fibers the viscoelastic response is shown to vary from small amounts at room temperature to large amounts for temperatures in excess of the T_g (180°C). For other fiber angles, i.e., 10°, 30°, 60°, the response is similar to the latter.

The time temperature superposition principle or the method of reduced variables is used to determine compliance master curves for each fiber angle tested. The 10° results are used to obtain the master curve for the in-plane viscoelastic shear modulus. The viscoelastic equivalent to the elastic orthotropic transformation equation is used incrementally to predict the master curves for the tensile compliance of the off-axis specimen. The incremental procedure by which the predictions are made are described in detail. The predicted master curves are compared to the ones determined experimentally. Further, 25 hour creep test results for T = 180°C are compared to predictions. Agreement between theory and experiment in all cases is reasonable.

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INTRODUCTION

Advanced composite laminates have enormous potential for wide use as primary and secondary structural components in aerospace applications due to their high strength to weight ratios. At present a number of factors inhibit the ready acceptance of these materials by air frame manufacturers. One such factor is the current lack of understanding of the mechanical behavior of polymer based laminates under certain long-term environmental exposure conditions. It is well known that the epoxy resins which are now used as the polymer matrix component of a composite exhibit viscoelastic or time effects which are significantly affected by exposure to both temperature and humidity. Epoxies soften as temperatures are increased with resulting loss of both moduli and strength [1-4]. In addition, they absorb moisture and swell giving rise to residual stresses [5-8].

Obviously, polymer based composite laminates will be similarly affected by moisture and temperature under certain circumstances. Fiber-dominated composites are not likely to suffer large reductions of either moduli or strength in the fiber direction. In other directions, properties will be affected by losses of both strength and modulus. Perhaps more importantly, drastic interply effects will occur. That is, delaminations due to internal residual stresses caused by both temperature and humidity are likely to occur together with a general loss of interply shear transfer capacity.

Because of these effects of environment, there is concern that time-dependent properties such as creep, relaxation, delayed failures, creep ruptures, etc., may be important long-term design considerations for the temperature and moisture levels anticipated in current structural applications. In fact, it has been shown that a delayed or viscoelastic fracture response was encountered during a relaxation test of a $[\pm 45^\circ]_{4s}$ tensile strip containing a circular hole at stress levels near ultimate [9]. That is, a creep to failure response occurred for individual plies even though the applied load to the laminate was relaxing in a fixed grip situation. Obviously, should the same phenomena occur in prototype structures, unsafe premature failures would occur.

To be able to predict the lifetime of a structure under known loading conditions requires either long-term testing or a model upon which long-term results can be predicted from short-term tests. Obviously, long-term testing equivalent to the duration of the lifetime of a structure is undesirable. The alternative is to develop analytical or experimental models which can be successfully used for extrapolation. For metals and polymers a variety of techniques are available such as linear elastic stress analysis, empirical extrapolative equations such as the Larson-Miller parameter method, Minor's rule and frequency independence, the time-temperature superposition principle, etc.

One objective of this investigation was to study and measure the thermomechanical viscoelastic response of a lamina that is required for the analytical characterization of a laminate. This was to be accomplished by the usual method of ascertaining the appropriate properties of a unidirectional *laminate* and assuming that the results would be applicable to a single *lamina*. Further, another objective was to determine if the time-temperature superposition principle could be applied to unidirectional properties and to use the master curves so generated together with the orthotropic transformation equation to predict the off-axis compliance of a unidirectional laminate.

MATERIAL PROPERTIES

When a unidirectional continuous fiber composite lamina is elastically deformed under a state of plane stress, only four elastic constants are required to completely characterize the material behavior. In terms of engineering constants, these quantities are the principal Young's moduli, Poisson's ratios, and the in-plane shear modulus. The Young's moduli are usually written as E_{11} and E_{22}, where E_{11} is the modulus when the applied load is parallel to the fibers, and E_{22} is the modulus when the fibers are perpendicular to the applied load. The Poisson's ratios are expressed as ν_{12} or ν_{21}, or simply as ν_{ij}, where ν_{ij} is the ratio of transverse strain in the j-direction to the longitudinal strain in the i-direction for a stress in the i-direction. And finally, the shear modulus is normally written as G_{12}, which is the intralamina shear modulus. Thus, it would appear that five elastic constants are necessary to characterize orthotropic behavior. However, the reciprocal relation $\nu_{12} E_{22} = \nu_{21} E_{11}$ must be satisfied, leaving only four elastic constants.

Experimentally, only three laminate orientations are required to obtain the four elastic constants. Three of the constants, E_{11}, E_{22}, and ν_{12} can be easily determined [10], and the intralamina shear modulus can be found using the 10° off-axis tensile test proposed by Daniel and Liber [11] and Chamis and Sinclair [12]. A recent comparative study by Yeow and Brinson [13] on the shear response of T300/934 graphite/epoxy laminates concluded that the 10° off-axis test gave better results for the shear modulus than either the $[\pm 45^\circ]_s$ tensile test method of the $[0^\circ/90^\circ]_s$ symmetric rail shear test method.

If the material is viscoelastic, the material parameters previously discussed will be a function of time, temperature, relative humidity, stress level, etc. [14]. However, the present study will address only the effects of time and temperature on the material response. The stress level will be such that the material response is linear. The viscoelastic response will be determined using experimental methods similar to those for elastic materials [15]; however, consideration must be given to the time, temperature and stress level.

EXPERIMENTAL CONSIDERATIONS

The particular composite materials studied in this investigation were manufactured from T300/934 graphite/epoxy pre-preg tapes by Lockheed-Sunnyvale, CA. All test specimens were cut from a single large panel of the material using a diamond wheel. In addition, all specimens were unidirectional laminates whose geometry and fiber orientations with respect to load direction were $[0^\circ]_{8s}$, $[90^\circ]_{8s}$, and $[10^\circ]_{8s}$. Load was transferred to the 1.27 x 25.40 x 0.23 cm (0.5 x 10 x 0.09 in.) specimens through fiberglass tabs bonded to the laminate, such that the distance between end tabs was 15.24 cm (6 in.).

Tensile specimens were instrumented with three-element rectangular strain gage rosettes, type CEA-06-125UR-350, from Micro-Measurements. The temperature range of this gage is -75°C to 205°C (-100°F to 400°F), which is adequate for the temperature range considered in the experimental program. Strain gage rosettes were bonded to specimens with M-Bond 610 adhesive. Two rosettes were bonded to each specimen, one on each side, and the gage outputs were averaged to eliminate any out-of-plane bending. Rosettes were temperature compensated using specimens whose orientation was identical to the stressed specimens, except in the thermal expansion measurements where a low expansion (0.5 $\mu\varepsilon$/°C) "dummy" specimen was used.

Load was applied to test specimens through special grips, similar to those used by Chamis and Sinclair [12]. As shown in reference [12], these grips together with the specimen length used, tended to minimize the in-plane bending discussed by Pagano and Halpin [16]. The grips used by Chamis and Sinclair were modified by the addition of a pin that extended through the midpoint of the clamped specimen section. This pin helped to reduce slippage between the clamps, bonded end tabs, and test specimen.

Lou and Shapery [17] found it necessary to mechanically condition fiberglass composites in order to produce the same strain output under the same load for repeated creep tests on the same specimen. In the work reported herein it was decided to subject specimens to both thermal and mechanical conditioning to check the strain response for the same load. It will be shown that neither type of conditioning substantially altered the mechanical properties. Thus, it was established that repeated use of the same specimen could be made.

Specimens were subjected to a constant load and thermally conditioned by slowly raising the temperature from room temperature to 210°C, then slowly cooling back to room temperature. When this process was repeated five times it was found that the mechanical properties were practically the same for the fifth cycle as for the first. Slightly higher coefficients of thermal expansion were measured during the initial stage of the first cycle, which is probably due to moisture in the material.

Mechanical conditioning was accomplished by first subjecting specimens to creep loads of approximately 50% of the room temperature ultimate strength. After 16 minutes the load was removed, the specimen was allowed to recover to approximately zero strain, and the process was repeated five times. The same specimens were then subjected to fifty low frequency cyclic loads. Mechanical properties measured after the cyclic loading were essentially the same as those measured after the first thermal cycle. Thus, conditioning of the specimens was unnecessary, and it was established that the same specimen could be used repeatedly.

DATA REDUCTION AND DISCUSSION OF RESULTS

The glass transition temperature was determined by placing specimens instrumented with three-element rectangular strain gage rosettes in a thermal chamber and measuring thermal expansions as a function of temperature. Each measurement was made after an equilibrium temperature was reached. The results, shown in Figure 1, indicate two transition temperatures. A secondary transition temperature, Tg^S occurred at 60°C, whereas the primary glass transition temperature, Tg, was at 180°C. Thermal cycling did not appreciably affect the results.

Figure 1 also reveals three distinct values of the coefficient of thermal expansion in the 90° (perpendicular to fibers) and 45° directions. The coefficient of thermal expansion in the 90° direction was twice that in the 45° direction, as it should be if the tensor transformation equations hold. Thermal expansion along the fibers was very small.

A typical creep and recovery curve is shown in Figure 2. In order to consider the material as linear viscoelastic, the instantaneous creep and creep recovery strains, ε_0, should be the same [18]. In all isothermal tests such a response was noted.

Another linearity check is shown in Figure 3, where creep stress is plotted against 15 min. creep strain, for two temperatures. This figure shows that there was an upper bound below which the axial stress - axial strain (15 min.) relationship was linear, and that the upper bound tended to decrease for increasing temperatures. In all subsequent calculations the applied load was such that the axial stress - 15 min. axial strain relationship was in the linear range.

Figure 3 also shows a linearity check for shear stress - shear strain. This behavior was obtained from a $[10^\circ]_{8s}$ tensile test, similar to that used by Chamis and Sinclair [12] and Yeow and Brinson [13 [illegible] [illegible]ra-lamina shear stress, τ_{12}, was determined from the transformation equation for stress as

$$\tau_{12} = \frac{1}{2} \sigma_x \sin 2\theta \tag{1}$$

where x denotes the load direction, and θ is the angle between the load direction and the fiber (or 1) direction. For the $[10^\circ]_{8s}$ specimen, $\theta = 10^\circ$.

Using the strain transformation equations, the shearing strain, γ_{12}, was calculated from the equation

$$\gamma_{12} = -(\varepsilon_x - \varepsilon_y) \sin 2\theta + \gamma_{xy} \cos 2\theta \tag{2}$$

where

$$\gamma_{xy} = 2\varepsilon_{45^\circ} - \varepsilon_x - \varepsilon_y$$

In these equations ε_x, ε_y and ε_{45° are the axial, transverse, and 45° strains, respectively, obtained from the three-element rectangular rosettes.

Five material parameters were measured in each isothermal creep test. These were the compliances and the Poisson's ratios represented by S_{11}, S_{22}, S_{66}, ν_{12} and ν_{21}. Two parameters, S_{11} and ν_{12}, were found to be time and temperature independent while the remaining three were found to be time and temperature dependent for the range of temperatures considered. The reduced viscoelastic compliances and Poisson's ratio were determined from the equations,

$$S_{22}(t) = \frac{\varepsilon_2(t)}{\sigma_0} \frac{T}{T_0} \tag{3}$$

$$\nu_{21}(t) = \frac{\varepsilon_1(t)}{\varepsilon_2(t)} \frac{T}{T_0} \tag{4}$$

$$S_{66}(t) = \frac{\gamma_{12}(t)}{\tau_{12}} \frac{T}{T_0} \tag{5}$$

respectively where t represents time, T was the absolute test temperature and T_0 was the absolute reference temperature (taken as the primary T_g of 453°K). The reduced properties $S_{22}(t)$ and $\nu_{21}(t)$ were obtained from creep test of $[90°]_{8s}$ tensile specimens where σ_0 was the applied stress. The reduced compliance $S_{66}(t)$ was obtained from creep test of $[10°]_{8s}$ tensile specimens and $\gamma_{12}(t)$ and τ_{12} were determined from equations (1) and (2).

It should be noted that the method of reduced variables normally used for polymers [1] was used to write equations (3), (4), and (5), that is

$$S_{ij}(t,T_0) = \frac{\rho T}{\rho_0 T_0} S_{ij}(t',T) \tag{6}$$

where

$$t' = \int_0^t \frac{dt}{a_T}$$

ρ and ρ_0 are densities at temperatures T and T_0, respectively, and a_T is the shift factor. For most solids the variation of densities with temperature is small and is often neglected. The same has been done herein.

Before discussing results, it might be appropriate to note that while creep tests were performed and compliances were calculated, we have elected to plot reciprocity of compliance in all cases. This was done such that properties could readily be interpreted qualitatively as the modulus variations one might expect in a relaxation test. Of course, no reciprocity between modulus and compliance is implied by our data or its representation.

Figure 4 illustrates the time-temperature behavior of the reduced reciprocal of $S_{22}(t)$, for short-time (16 min.) creep at various temperature levels. The reduced compliance was calculated using equation (3). Figure 4 shows the effect of time and temperature on $1/S_{22}$, and indicates that the effect of temperature is greatest for $T > T_g$ (= 180°C). A portion of the master curve is shown in Figure 4, while the complete master curve is shown in Figure 5, where the reference temperature was taken as the primary transition temperature.

Figures 6 and 7 show, respectively, the master curve for the reciprocal of the reduced shear compliance, and the master curve for the reduced Poisson's ratio. For all master curves, the change in creep properties is most pronounced after the primary transition temperature.

Figure 8 shows the unreduced one minute reciprocal of compliance vs. temperature results for temperatures from T = 20°C to T = 210°C for the $[90°]_{8s}$ laminate. The results shown in Figure 8 are as expected and are quite similar to the results one would obtain for a neat epoxy resin [1,2].

Each master curve (refer to Figures 5, 6, 7) was obtained by graphically shifting the reduced short-time curves, such as those shown in Figure 4. In some cases the curves at various temperatures did not overlap, and thus the shifting procedure required interpolation between constant temperature curves. The interpolative procedure is illustrated by referring to Figure 8. The constant time curve was fitted by a polynomial equation relating reciprocal of compliance to temperature. This procedure was repeated for other constant time curves. It should be noted that the polynomial equations were written in terms of reduced reciprocal of compliance, instead of reciprocal of compliance as shown in Figure 8. The resulting polynomial equations were then used to interpolate between short-time test results, such as those shown in Figure 4, to produce a complete master curve.

The values of the shift factor, a_T, as a function of temperature were obtained by graphically shifting reduced short-time curves (such as shown in Figure 4) along the log time axis until one continuous curve was obtained. This corresponds to multiplying every value of the absicca of a particular constant temperature curve by a constant factor. In all cases the curves were brought into alignment with the curve corresponding to the primary transition temperature. The results are shown in Figure 9. The shift factors for $1/S_{22}$ and $1/S_{66}$ are essentially the same. This result together with results for other fiber directions not included herein has established that [illegible] shift factor is independent of fiber orientation. A similar result was obtained by Moehlenpah, Ishai, and DiBenedetto [19].

The discrepancy between the reduced $\nu_{21}(t)$ shift factor and the other two components, as shown in Figure 9, might be the result of scatter in the transverse strain data of the $[90°]_{8s}$ specimens. For these specimens the axial (load direction) strain changed appreciably, whereas the transverse (fiber direction) strain showed small (approximately 5μ strain) random changes. Thus, it was difficult to draw smooth curves through the reduced ν_{21} - log time data (i.e., curves similar to Figure 4), which resulted in error in graphical shifting to find the shift factor.

Several analytical equations have been proposed for the shift factor behavior of polymers. An Arrhenius type equation is often observed when $T < T_g$, i.e.,

$$\log a_T = \frac{\Delta H}{2.303R}\left(\frac{1}{T} - \frac{1}{T_0}\right) \tag{7}$$

where ΔH is the activation energy, R is the universal gas constant, and T_0 is an arbitrary reference temperature. If the activation energy is constant, a plot of log a_T versus $\frac{1}{T}$ should be linear. Thus, it is easy to calculate the shift factor at any temperature less than T_g. However, Figure 10 reveals a nonlinear relationship between log a_T and $\frac{1}{T}$. Therefore, no further attempt was made at using an Arrhenius type equation.

Above the T_g a WLF type equation [1] is often used to predict the shift factor behavior of polymers. The results, depicted in Figure 9, show reasonable agreement between experimental data and the WLF type equation. The constants in the WLF equation were obtained by a best fit to the data and are different than those normally used for neat resins [1,2]. Below the T_g, the shift function differs markedly from the WLF equation, as would be expected. In addition, an inflection point in the shift function curve is observed at the Tg.

In subsequent calculations, neither the Arrhenius or WLF type equations were used. . Instead, polynomial equations were fitted to the shift function vs. temperature results. The use of such equations will be apparent later.

PREDICTION OF OFF-AXIS LAMINA COMPLIANCE

When using lamination theory it is necessary to calculate the transformed reduced compliances for a lamina arbitrarily oriented with respect to a laminate or global axis system [10]. For an elastic lamina this can be accomplished using the four orthotropic properties (S_{11}, S_{22}, ν_{12}, S_{66}) and the orthotropic transformation equation. In the present case, two properties, S_{22} and S_{66}, are time dependent and two properties, S_{11} and ν_{12}, are time independent. Thus, the orthotropic transformation equation can be written as,

$$S_{xx}(t) = S_{11} \cos^4 \theta + [S_{66}(t) - 2\nu_{12} S_{11}] \sin^2 \theta \cos^2 \theta + S_{22}(t) \sin^4 \theta \qquad (8)$$

where (t) denotes a function of time and S_{xx} represents the compliance of the lamina at the angle θ with respect to the fiber direction. Equation (8) is the viscoelastic analogue to the orthotropic elastic constitutive equation only for the case that S_{11} or ν_{12} is in fact time independent.

Often, under certain circumstances, the elastic version of equation (8) is used to predict S_{xx} or E_{xx} of a tensile specimen with the fibers at an arbitrary angle to the load (x) direction [10]. In order to assess the applicability of using the method of reduced variables and master curves in a viscoelastic version of lamination theory, equation (8) was used to predict the off-axis master curves for $S_{xx}(t)$ of $[30°]_{8s}$ and $[60°]_{8s}$ laminates.

As may be understood from an examination of Figure 4, a certain amount of experimental scatter was unavoidable. This was minimized by the smoothing operation used in obtaining the master curve for a particular function. Similar comments can be made relative to the shift function a_T. In addition, temperatures for each short-term test (16 min.) were difficult to repeat exactly. Therefore, it was desirable to work with the smoothed curves rather than discrete data points and to produce the entire S_{xx} master curve from a computer program written for that purpose which could be eventually used in conjunction with a lamination stress analysis program. Toward this end, data was taken from the master curves of Figures 5 and 6 for a generic point in time. This data was unreduced and by using the shift factor from the polynomial relation for a_T as a function of temperature for the appropriate point on the master curves of S_{22} and S_{66}, the master curve for S_{xx} was produced. As stated earlier the results are shown in Figures 11 and 12 for two different fiber orientations.

Also shown in Figures 11 and 12 are the master curves which were determined experimentally using the procedures described previously. As may be observed good agreement between experiment and our numerical solution of equation (8) was obtained.

In an effort to quantify the experimental error of short-term testing for the master curves and the calculation procedures used to determine off-axis master curves, twenty five hour creep tests were run for the $[10°]_{8s}$, $[30°]_{8s}$, $[60°]_{8s}$ and $[90°]_{8s}$ geometries at the reference temperature of $T = T_g = 180°C$. These results are shown superimposed on the master curves of Figures 5, 6, 11, and 12.

It is interesting to note that deviations between the twenty five hour (1.5×10^3 min) data and the master curves (i.e., S_{22} and S_{66}) produced from sixteen minute (1.6×10^1 min) tests were encountered. Obviously, such deviations could be due to either the experiments or the method of producing the master curves. We believe that the reason is likely the former and represents usual statistical variations encountered in composite materials.

Correlation between the three results, i.e., short-term tests, predictions, and twenty five hour tests, for the $[30°]_{8s}$ laminate is quite good. On the other hand, for the $[60°]_{8s}$ laminate, correlations among the three results are only reasonable. Again, it is felt that the explanation for these comparisons can be attributed to statistical variations normally found in composite materials and for that reason we feel the comparisons in all cases are quite good.

SUMMARY AND CONCLUSIONS

The results presented herein indicate that a particular graphite/epoxy material, T300/934, has properties which are both temperature and time dependent. It is likely that the results found are representative of other graphite/epoxy materials and, in fact, other polymer-based fiber reinforced materials.

Specifically, it has been shown that a glass-transition temperature of graphite/epoxy can be established by standard methods, i.e., measurement of thermal expansions. It might be noted that efforts were made to measure expansions using an extensometer but that significant error was produced due to the thermal expansion of the base metal of the extensometer. For example, negative expansions were sometimes indicated. Obviously, care must be exercised in interpreting results so produced and it is felt that electrical strain gages represent the better procedure.

The T300/934 graphite/epoxy material studied has two apparent transition temperatures, a small secondary transition at low temperature and a primary transition near the cure temperature ($\sim$ 180°C or 350°F) for the material. Further, these results likely are solely attributable to the resin.

A program of mechanical and thermal conditioning was performed and the results indicated that numerous cycling of either variable produced insignificant changes in mechanical properties. This fact was extremely important as time and temperature trends for the material could be established with a single specimen for each fiber orientation. Obviously, if reliable master curves for a material are to be produced then many tests will be needed to establish the statistical variation of properties with time and temperature. That was not the purpose of the present investigation.

It was shown that the compliance in the fiber direction, S_{11}, and Poisson's ratio, ν_{12}, were time independent. The former results were expected and simply verify that graphite fibers are insensitive to time and temperature. The latter results of time independence of ν_{12} can only be considered tentative at this time as the strain levels in the transverse direction were quite low with significant amounts of scatter.

The two compliances represented by S_{22} and S_{66} were shown to be time and temperature dependent, especially at temperatures near and above the primary T_g. Further, it was shown that the data could be used to produce master curves for these two functions. Shift functions were obtained and it was shown that an Arrhenius type equation did not model the data well. However, reasonable agreement between a WLF type equation [1] and the shift factor for temperatures above the Tg was obtained.

The master curves and shift functions were used in conjunction with a viscoelastic orthotropic transformation equation and additional master curves for two off-axis tensile geometries were predicted. In addition, twenty five hour creep tests were peformed and the results compared to predictions. In all cases the correlation between predictions and experiment were reasonable. Deviations were encountered and could be attributable to either the experiments (short- or long-term) or to the analytical methods used to make the predictions.

The fundamental conclusions to draw from this investigation are:

- Fiber dominated properties are time and temperature insensitive.
- Matrix dominated properties are time and temperature sensitive.
- Master curves of matrix dominated properties can be produced.
- Predictions of transformed reduced stiffness master curves can be made by usual procedures.

The latter two items essentially mean that a lamination theory computational procedure including viscoelastic properties is feasible.

Probably the most important significance of the data included herein is that composite structural components should not be allowed to encounter temperatures in excess of the Tg. If such were to happen significant matrix and interply viscoelastic effects should be expected.

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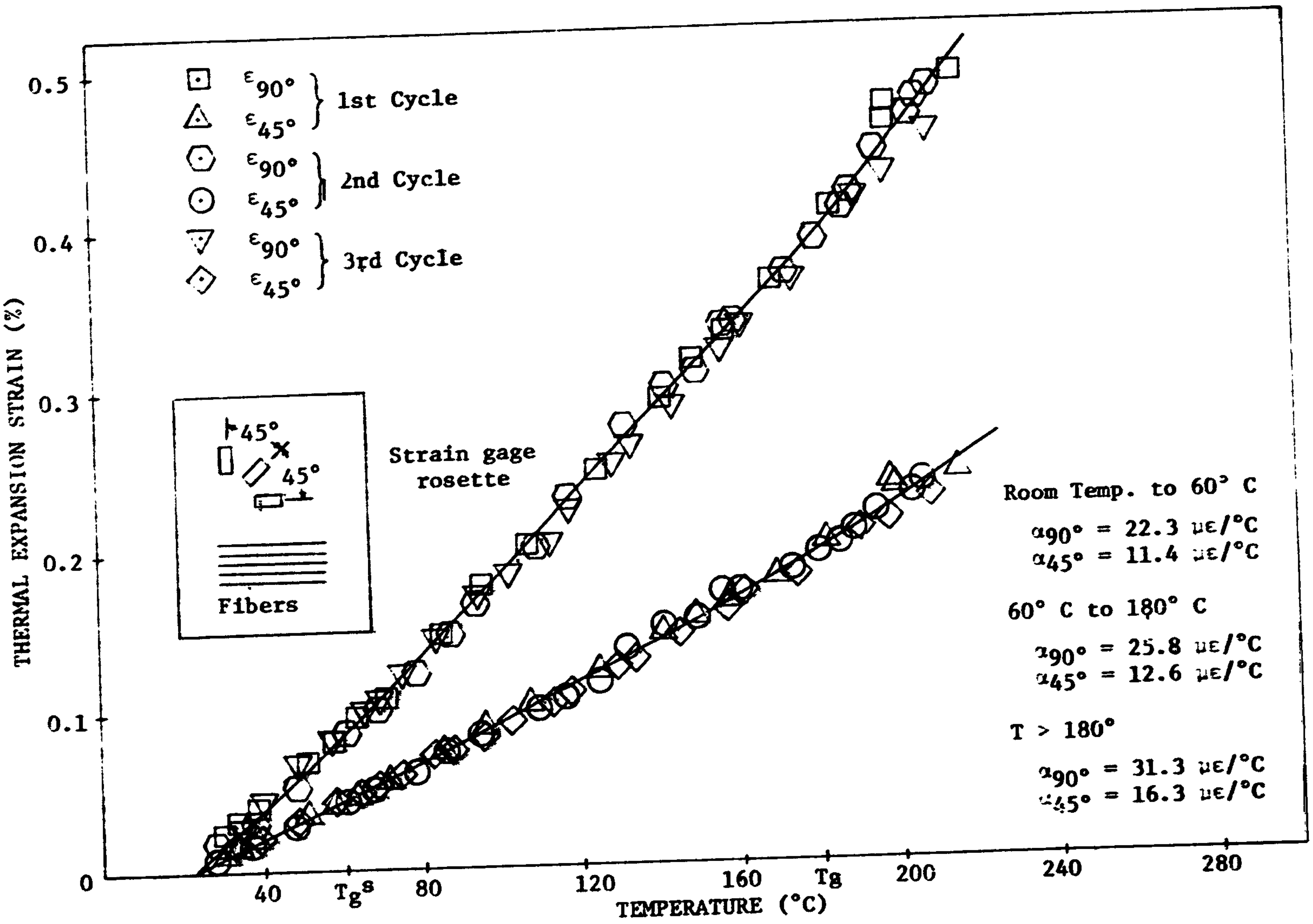

Fig. 1. Thermal Expansion of $[90°]_{8s}$ T300/934 G/E Laminate with Glass-Transition Temperatures Indicated.

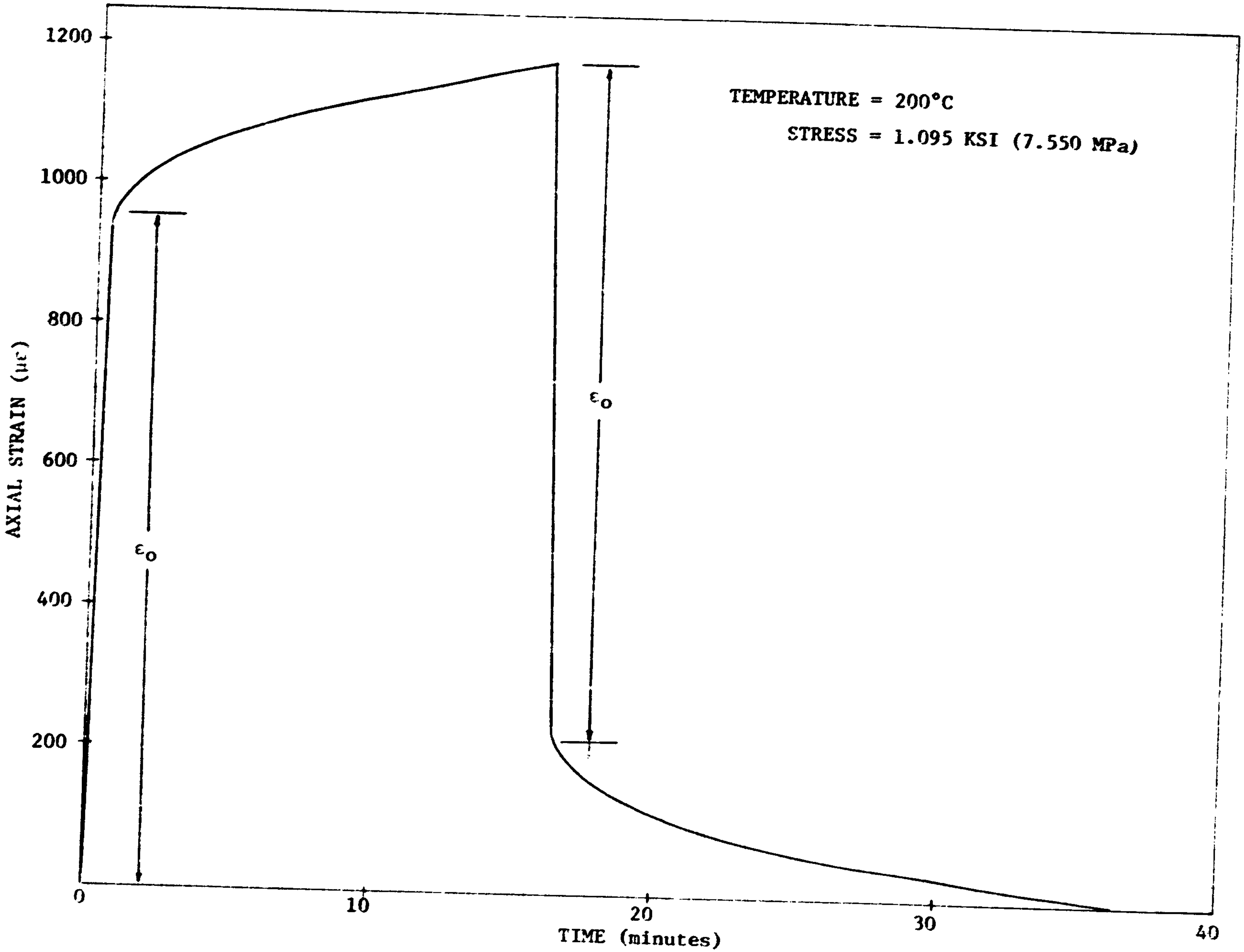

Fig. 2. Creep and Creep Recovery of $[90°]_{8S}$ Laminate.

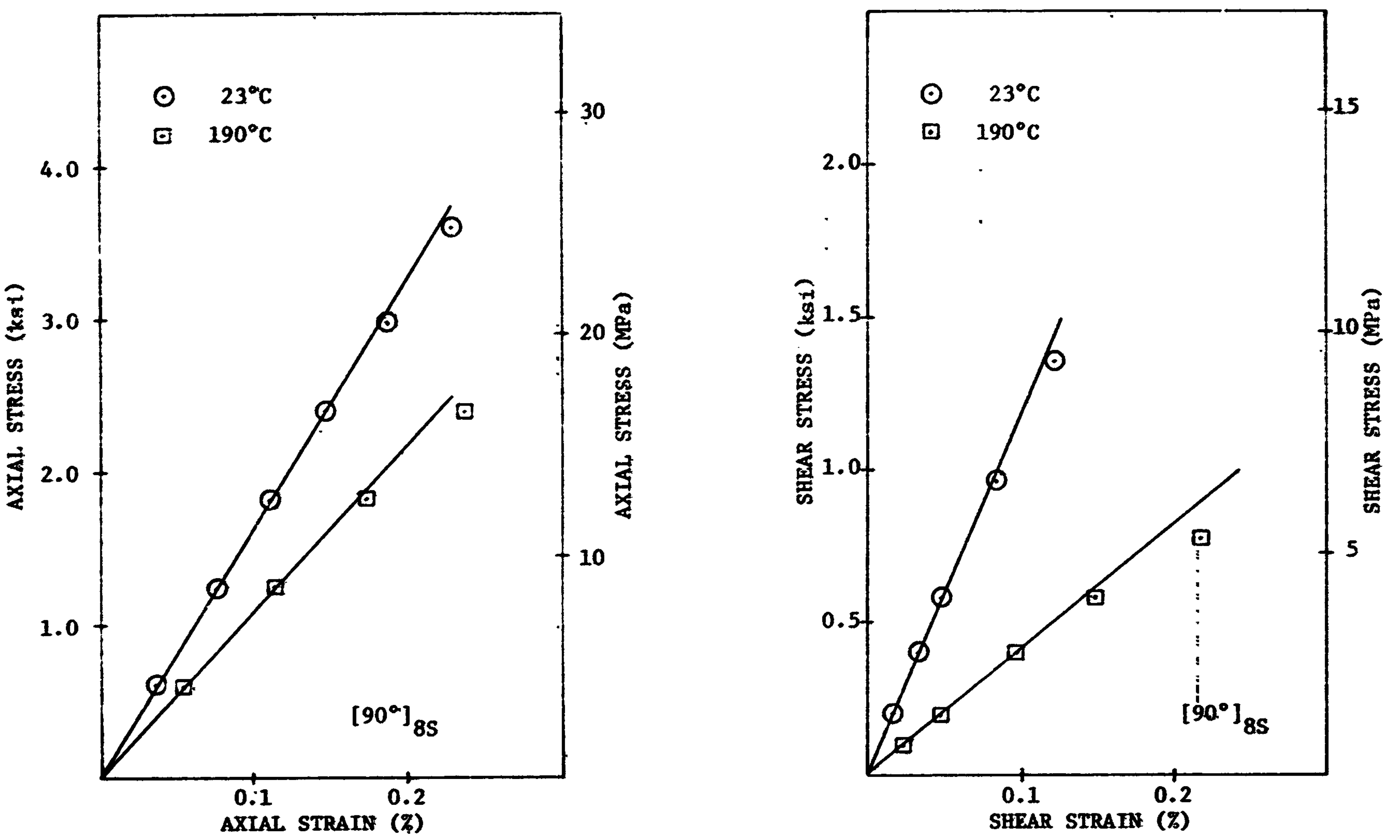

Fig 3. Linearity Check (Stress-Strain Curves After 15 Min. Creep).

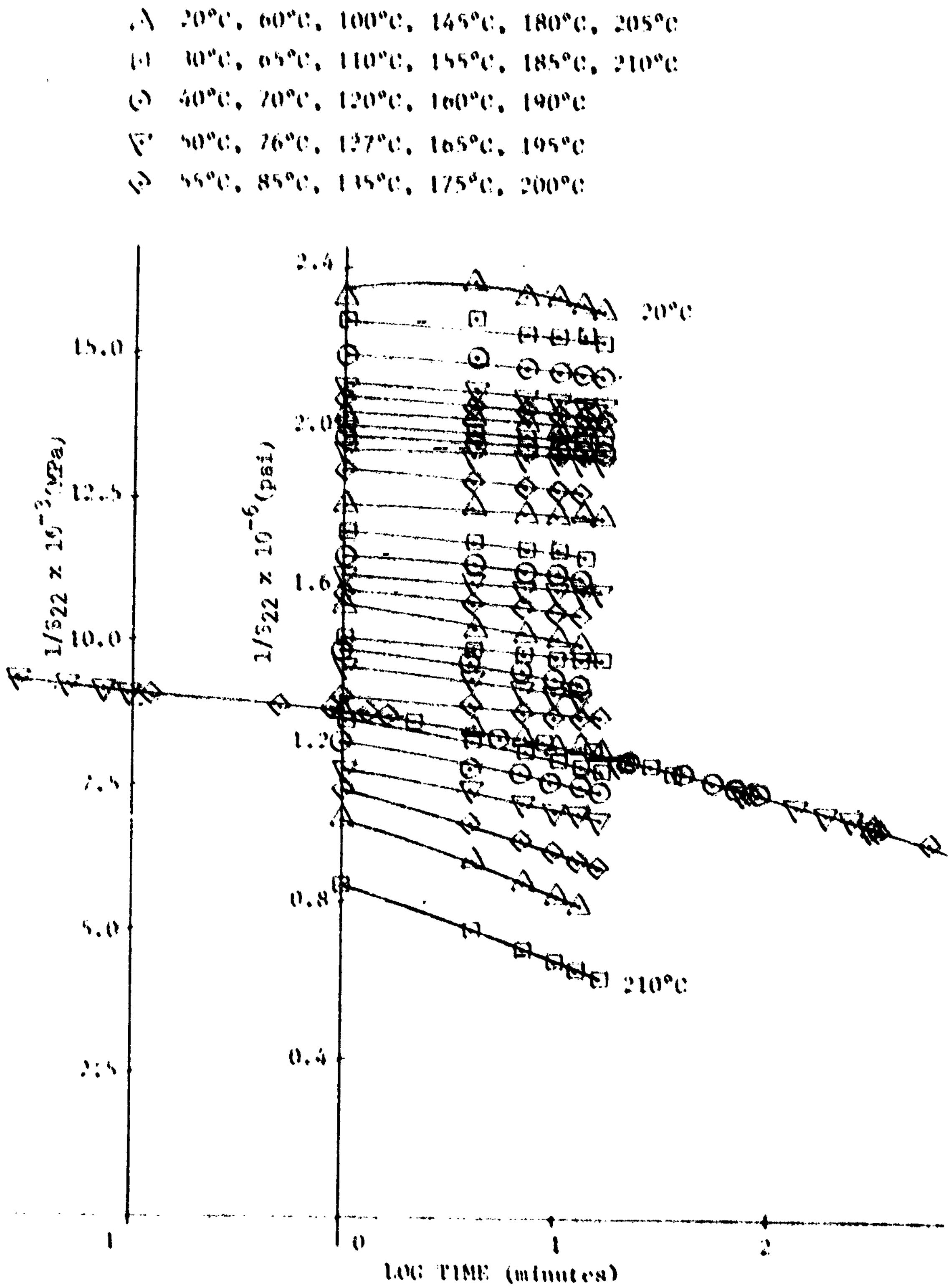

Fig. 4. Reduced Reciprocal of Compliance, $1/S_{22}$, and Portion of 180°C Master Curve for $[90°]_{8s}$ T300/934 Graphite/Epoxy Laminate.

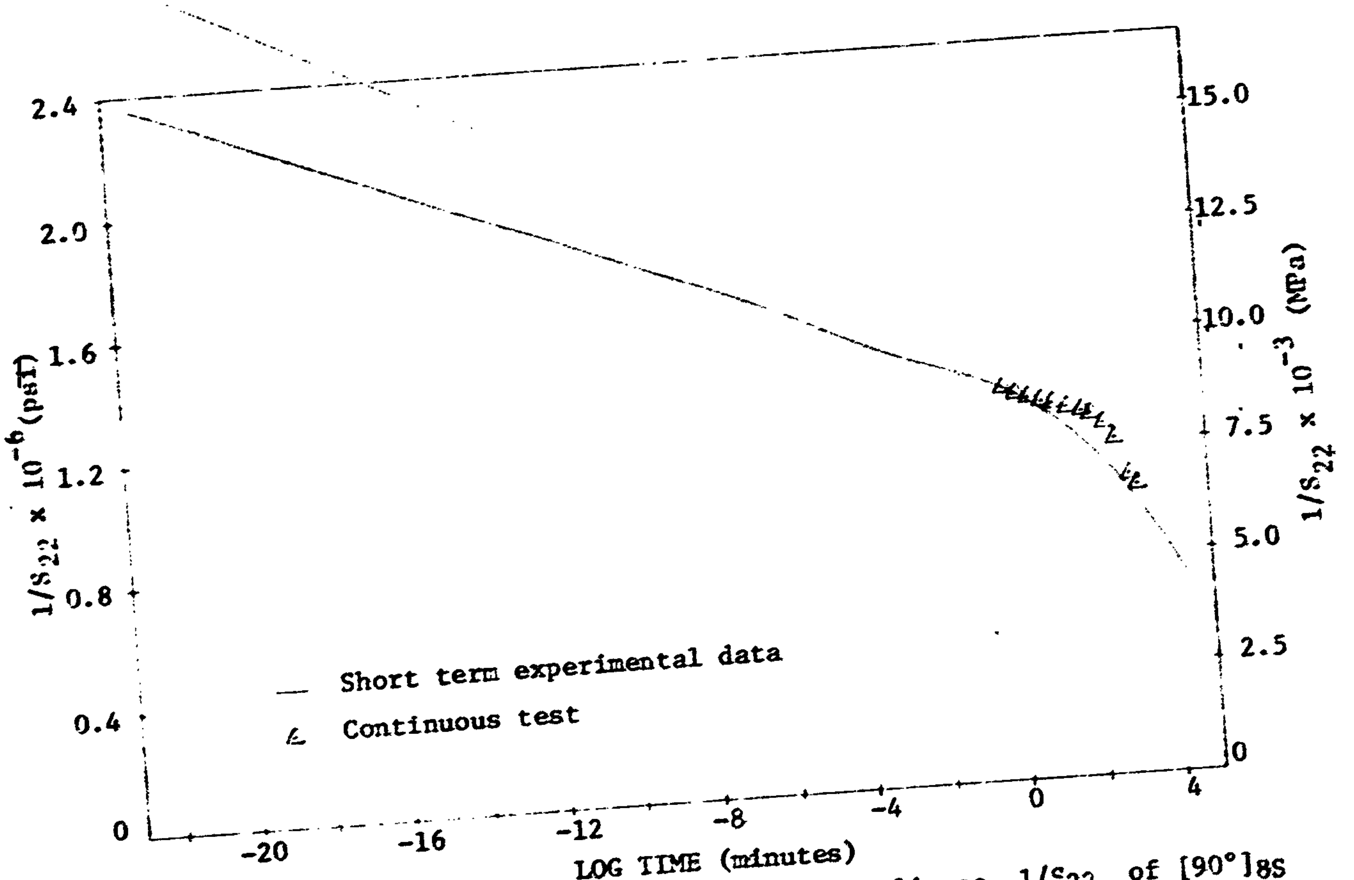

Fig. 5. Master Curve of the Reciprocal of Reduced Compliance, $1/S_{22}$ of $[90°]_{8S}$ Laminate at 180°C.

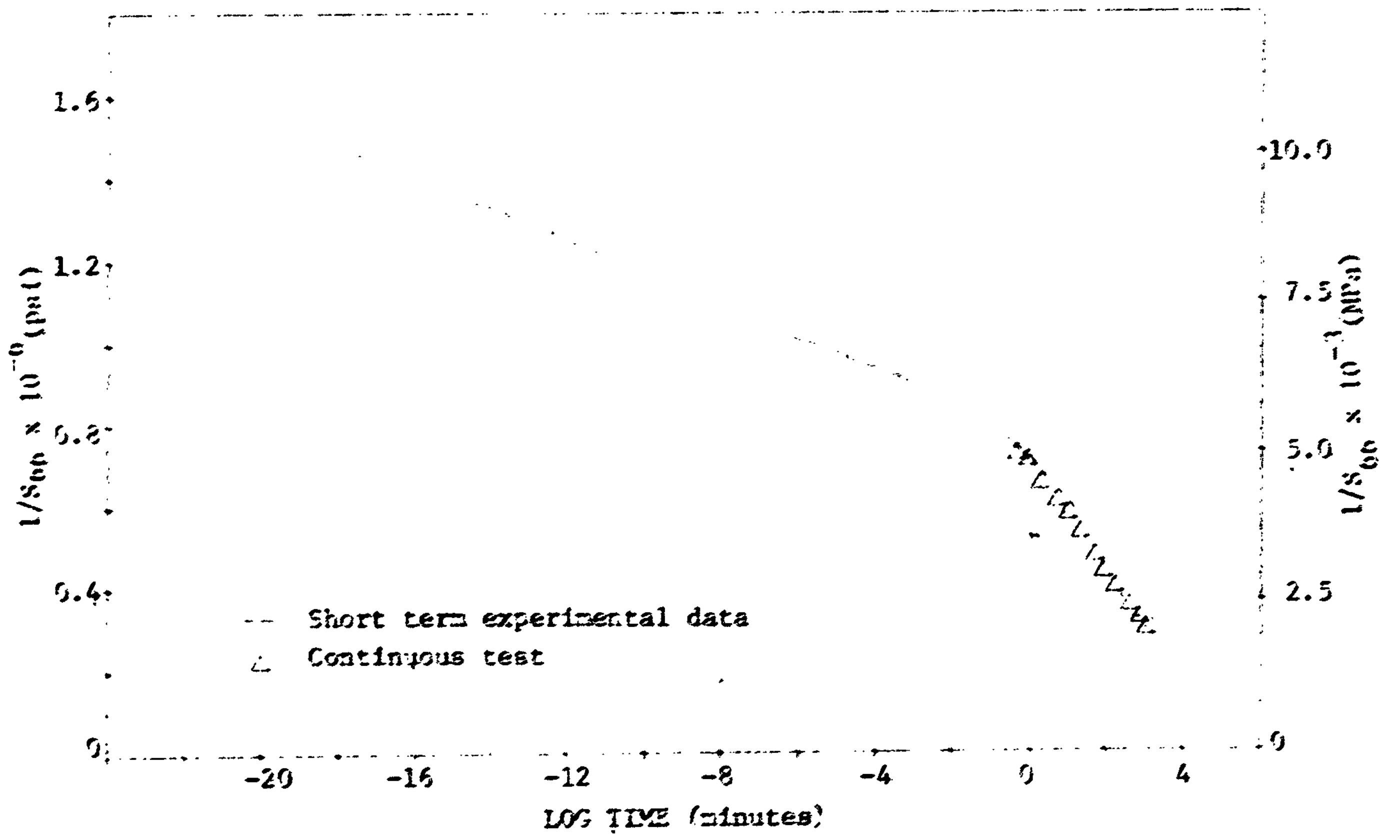

Fig. 6. Master Curve of the Reciprocal of the Reduced Shear Modulus, $1/S_{66}$, at 120°C.

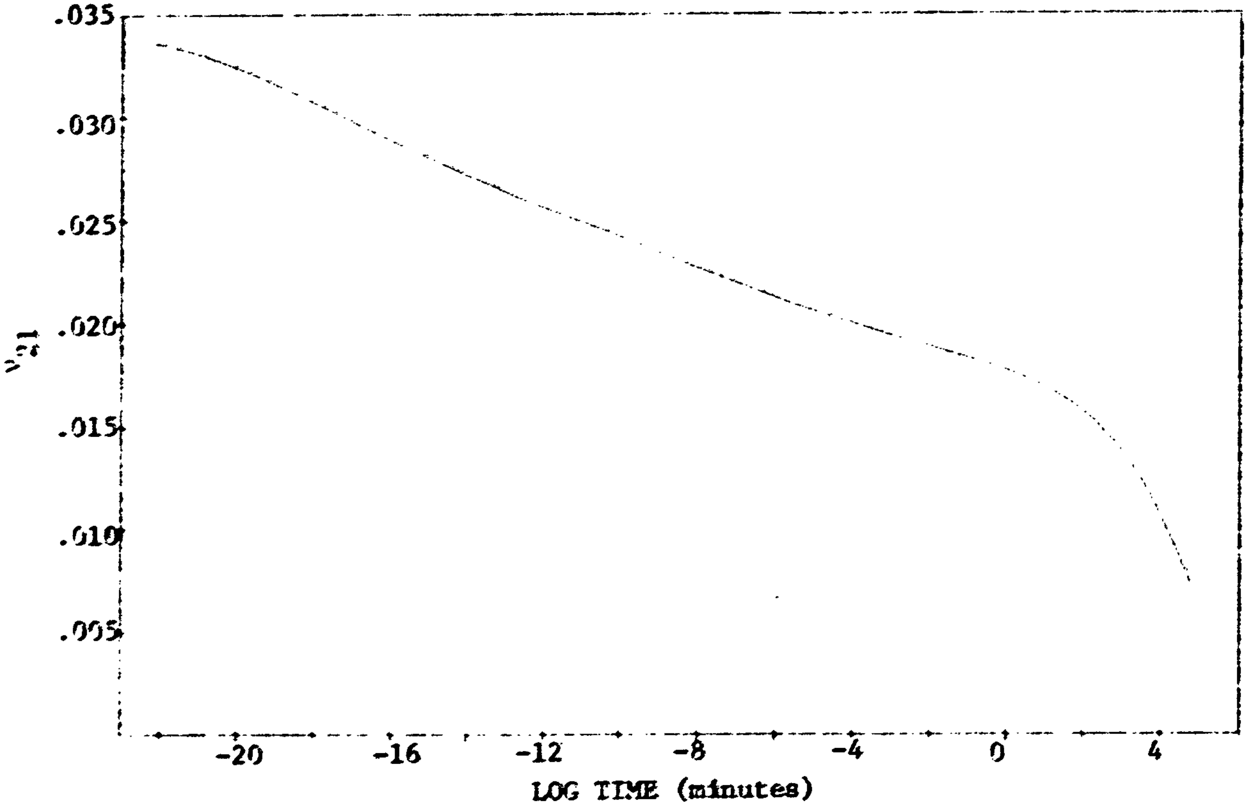

Fig. 7. Master Curve of the Reduced Minor Poisson's Ratio at 180°C.

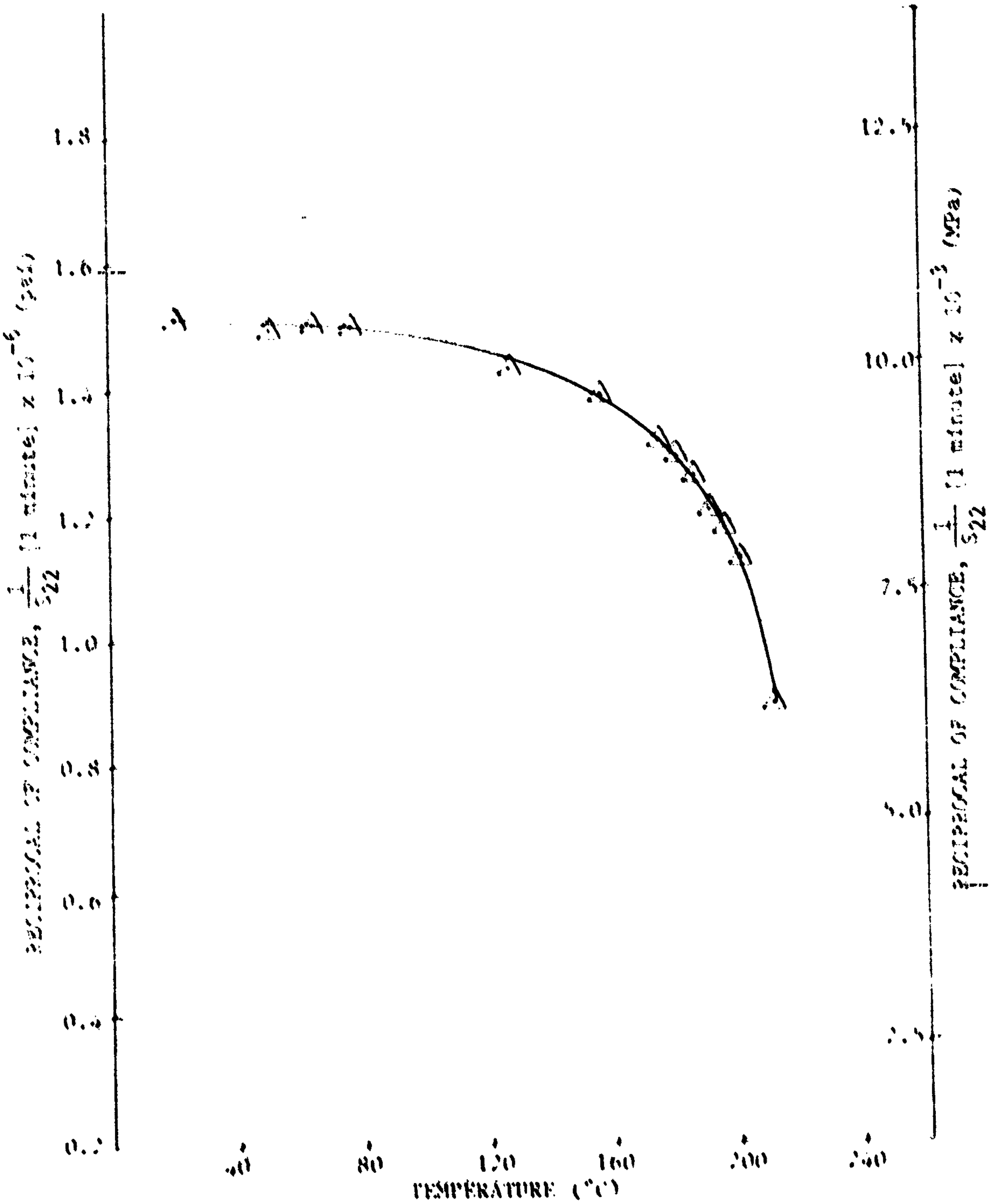

Fig. 8. One Minute Axial Creep Reciprocal of Compliance Showing Temperature Dependence for $[90°]_{8s}$ T300/934 G/E Laminate.

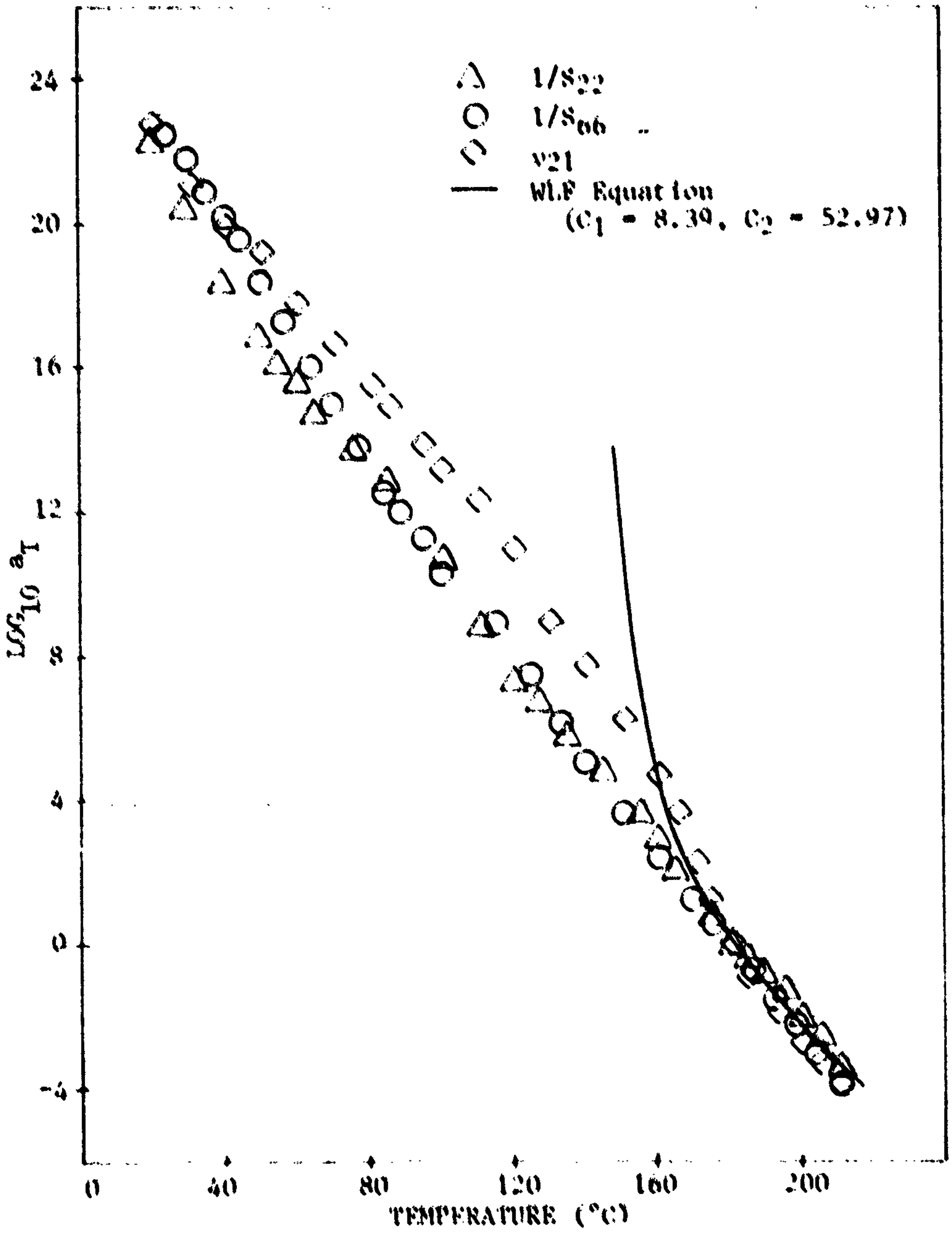

Fig. 9. Log a_T Versus Temperature.

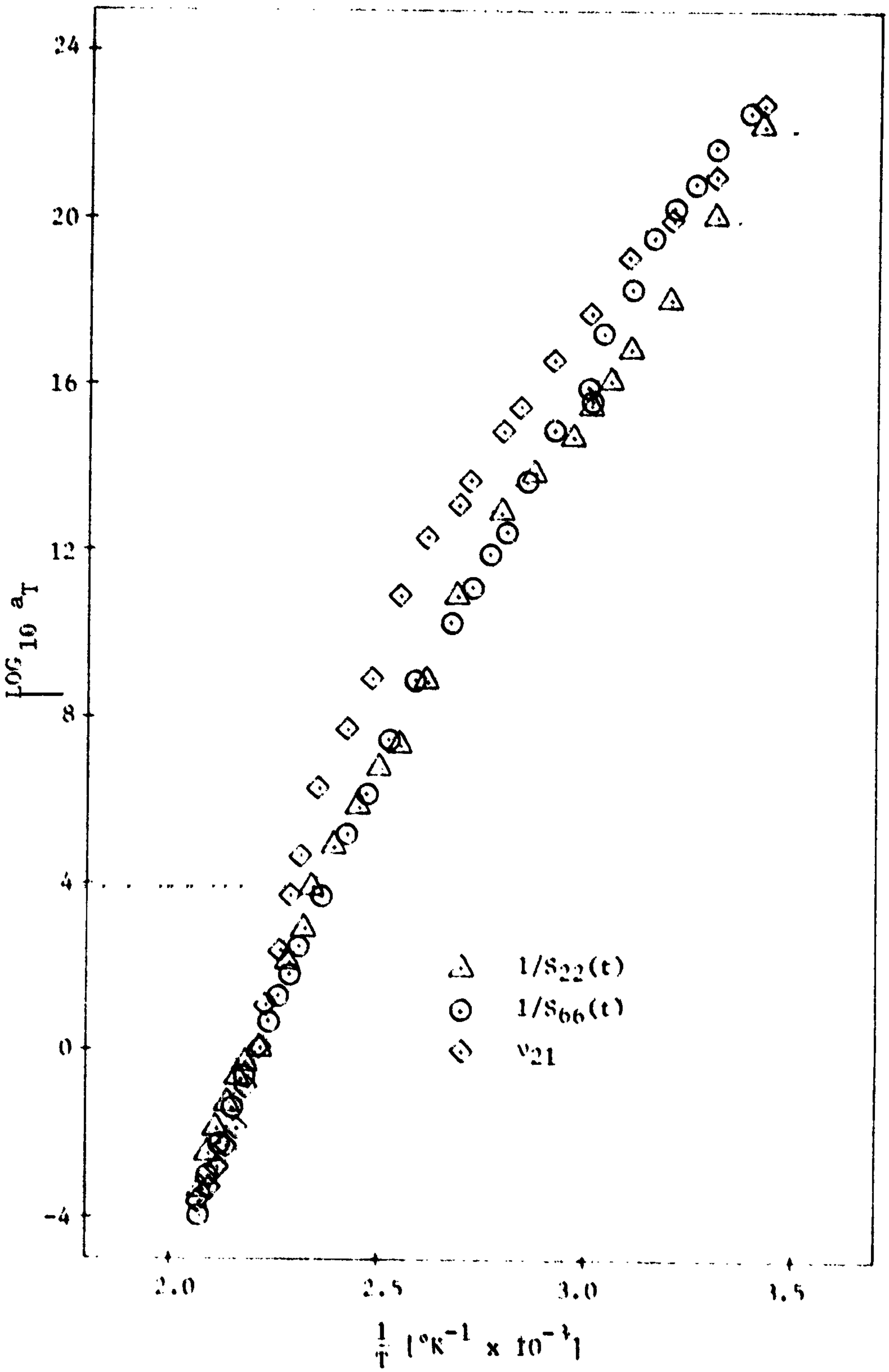

Fig. 10. Log a_T Versus Reciprocal of Absolute Temperature.

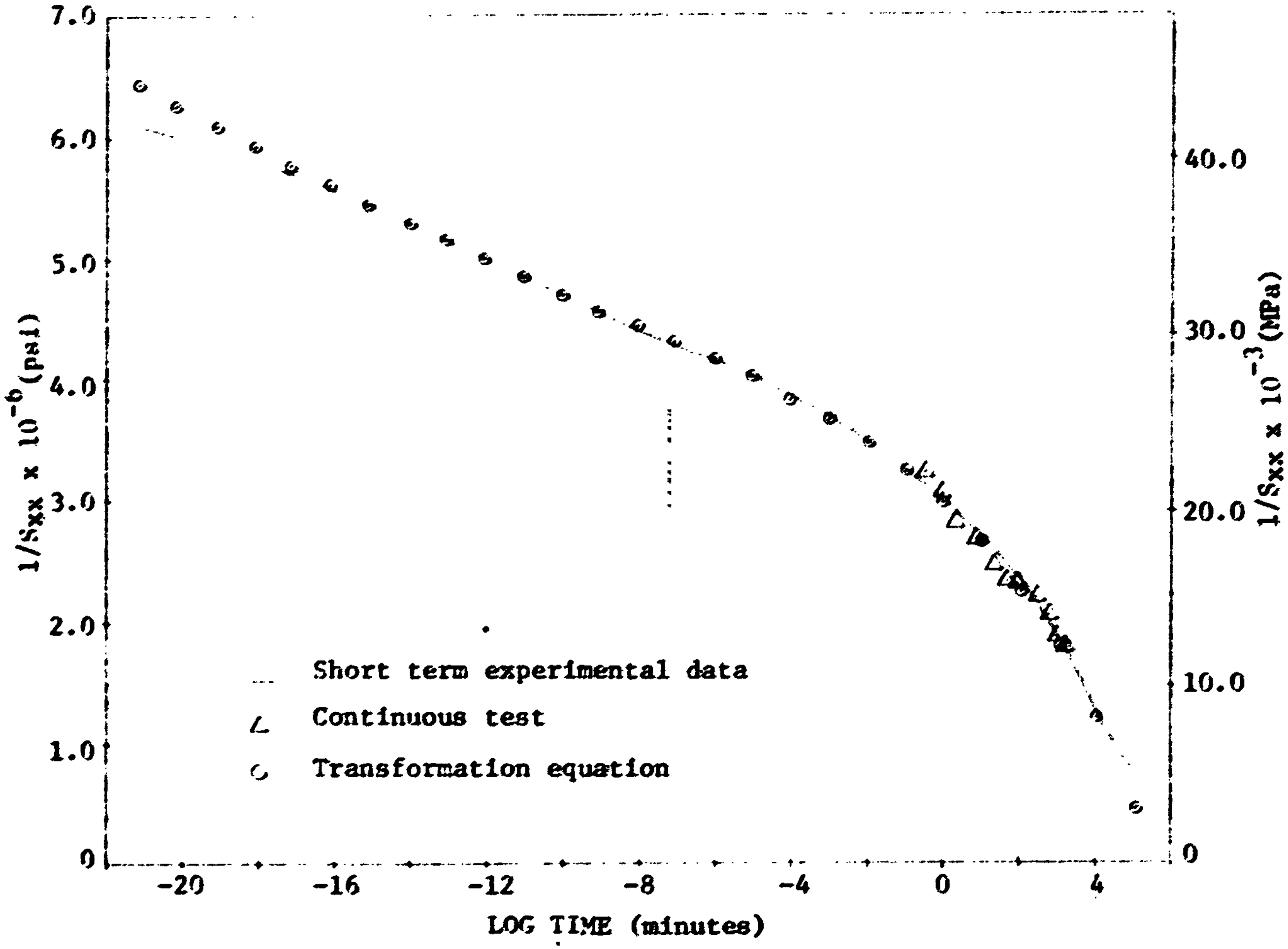

Fig. 11. Master Curve of the Reciprocal of Reduced Compliance, $1/S_{xx}$, of $[30°]_{8S}$ Laminate at 180°C.

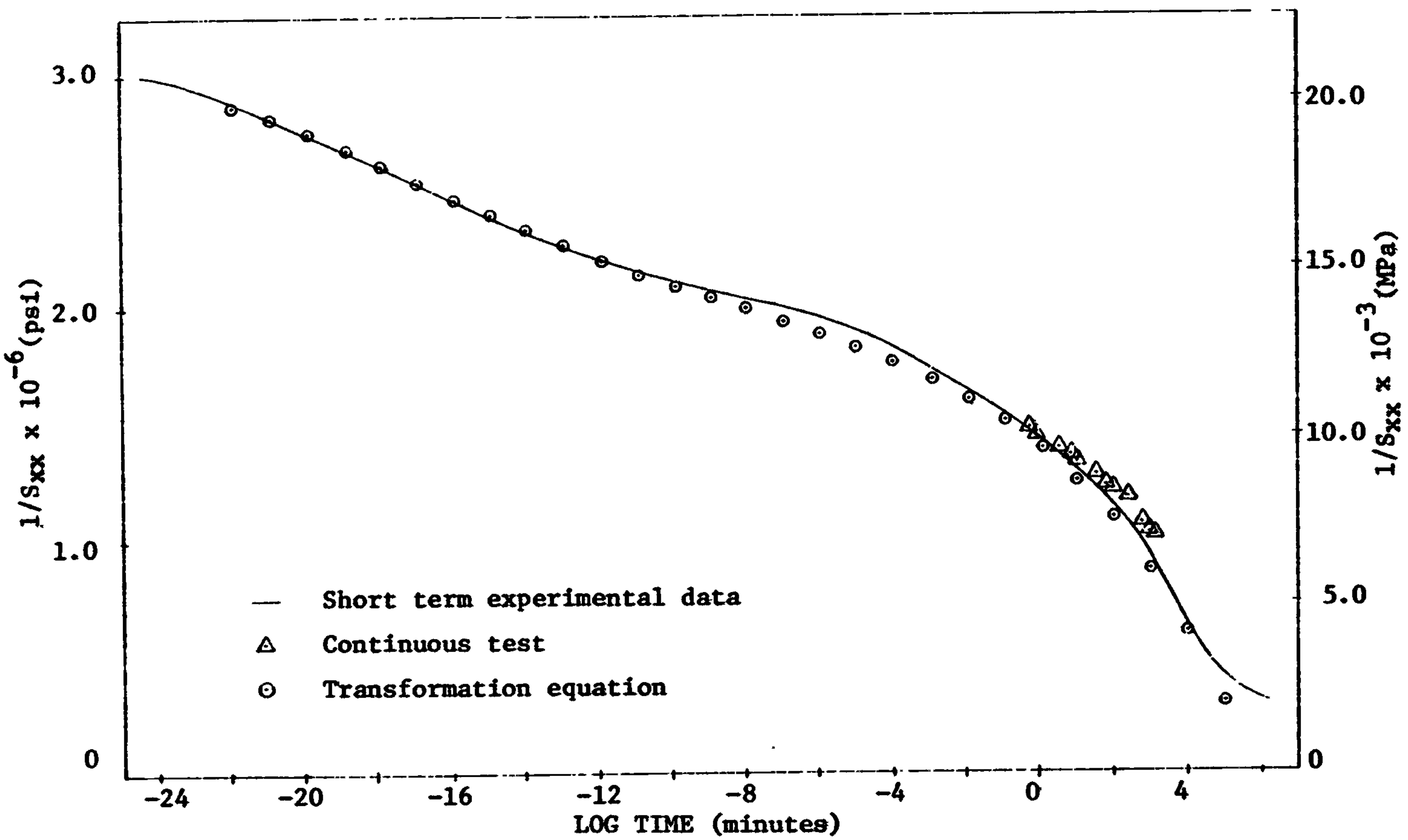

Fig. 12. Master Curve of the Reciprocal of Reduced Compliance, $1/S_{xx}$, of $[60°]_{8S}$ Laminate at 180°C.

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