JOURNAL OF MATERIALS SCIENCE 9 (1974) 665–672

Deformation characteristics of reinforced epoxy resin

Part 2 *An analysis of thermally activated deformation*

E. PINK*, J. D. CAMPBELL
Department of Engineering Science, University of Oxford, UK

Tensile specimens of reinforced epoxide resin of various orientations with respect to the fibre direction, were tested at low strain-rates (3.3×10^{-4} sec^{-1}) and over a wide range of temperatures (77 to 380 K). The rate sensitivities were measured by strain-rate change tests. Activation parameters were determined, by means of which it was possible to correlate the behaviour of composites of various orientations with that of the unreinforced epoxy. Three temperature regions have been established in which it appears that the deformation is controlled by three separate rate-controlling mechanisms.

1. Introduction

Part 1 [1] described effects of strain-rate and temperature on some mechanical properties of pure epoxy and fibre-glass composites. The object of the present report is to offer more detailed information from combined temperature and strain-rate changes which, in turn, may allow the correlation of the properties of unreinforced and reinforced materials, and which may help to establish the rate-controlling mechanisms.

The rate sensitivity of partially crystalline polymers is believed to be governed by deformation processes which occur in the amorphous fraction [2]. For the amorphous epoxy (Araldite MY 753 resin and HY 951 hardener) used in the present tests, an attempt was made to maintain a uniform density of cross-linking in both the reinforced and unreinforced resin by keeping the processing conditions constant. Further details on the woven glass reinforcement and on the definition of orientation have been given in Part 1.

2. Testing procedures and results

The combined effect of temperature and strain-rate was determined by straining a tensile specimen at a certain temperature, and by increasing and decreasing the strain-rate a number of times during the test. The advantages of this method have been discussed in a previous report [3]. Stress changes, $\Delta\sigma$, corresponding to a certain strain-rate change from $\dot{\epsilon}_1$ to $\dot{\epsilon}_2$ or vice versa are recorded during the test (see insert in Fig. 1), and the results can be represented in the normalized form of the strain-rate sensitivity

$$\lambda = \frac{\Delta\sigma}{2\ln(\dot{\epsilon}_2/\dot{\epsilon}_1)} \tag{1}$$

whose physical meaning will be discussed later. In our tests the strain-rate was usually increased from $\dot{\epsilon}_1 = 3.3 \times 10^{-4}$ sec^{-1} to $\dot{\epsilon}_2$ by a factor of 5. The strain dependence of the strain-rate sensitivity has previously been demonstrated for room temperature [1]. In the present investigation, this result has been basically confirmed for all temperatures, but variations of the numerical values have been observed. From plots of this type one can obtain the λ-values for the various strains at various temperatures, and Fig. 1 shows the combined results of λ versus T at a certain strain for all orientations. The figure also shows the values obtained from tests on the pure epoxy, and they are distinctly lower. Below 170 K, λ for the pure epoxy is effectively zero. However, this is not a true value, characteristic of the inelastic deformation, but is owing to the fact that the yield point has risen beyond the fracture stress (in fact, the temperature dependence of the yield point of epoxy seems to exhibit a discontinuity at 170 K; the values at lower temperatures suddenly become larger [11]).

---

*Now at: Erich-Schmid-Institut für Festkörperphysik, Österreichische Akademie der Wissenschaften, Leoben, Austria.

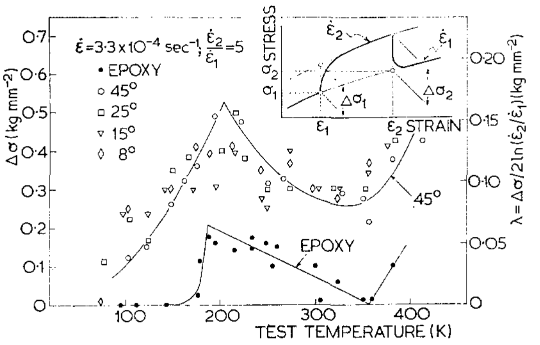

Figure 1 The temperature dependence of the strain-rate sensitivity measured for various orientations at a strain of 2%.

Some other properties and their temperature dependence have been reported in Part 1. It is noteworthy that strength-temperature curves exhibit similar discontinuities at the same temperatures as those which appear in Fig. 1.

3. Discussion

3.1. General remarks

In any study which attempts to relate the deformation behaviour of a rate-sensitive material to the test temperature, it is sensible to consider the possible existence of a deformation mechanism which is aided by thermal vibration of the participating elements.

Since, through an Arrhenius relation, such a process is time dependent and thus related to the strain-rate, any results from strain-rate changes should be considered in terms of thermal activation. According to this theory, no change in stress is expected to arise from a strain-rate change at zero absolute temperature. At high enough temperatures, when thermal activation alone is sufficient to overcome all barrier potentials, again no stress change is anticipated. Between those extremes, and assuming that only one type of obstacle is rate-controlling, $\Delta\sigma$ and, consequently, λ should vary smoothly with temperature. Any discontinuity in the temperature dependence of λ is, therefore, an indication of a change in the rate-controlling process. The dependences in Fig. 1 are, therefore, interpreted in this way: three different regions I, II and III are present, though it is perhaps difficult to associate the λ-values everywhere immediately with a single process: in the ranges 100 to 200K and 350 to 400K they may represent only transition properties.

In the case of metals a change or modification of the deformation mechanism generally affects the stress-temperature dependence. In plots of the fracture stress, e.g. those in Fig. 9 of Part 1, this was already visible. This should be still more obvious if one compares stresses at well defined strains, at which scattering should be reduced compared with that of the fracture stresses. Further, although there are differences in the absolute stress levels for samples with various orientations, the degree of the temperature dependence makes it possible to superimpose all these curves, as shown in Fig. 2. All results are consistent with the existence of three separate regions corresponding to those determined by the strain-rate change method. These features are reproducible: they have also been observed in other series of similar tests carried out on a different batch of material. Although the evidence from the temperature dependence of the stress is less conclusive than that presented in Fig. 1, it does provide confirmation that there are, in fact, three deformation mechanisms operating in different temperature ranges. Of further significance is the fact that the three-sectioned appearance is observed for stresses measured at either total strains ϵ or at strains ϵ' taken from the onset of yielding (i.e. the deviation from linearity). Slopes and transitions between the three regions

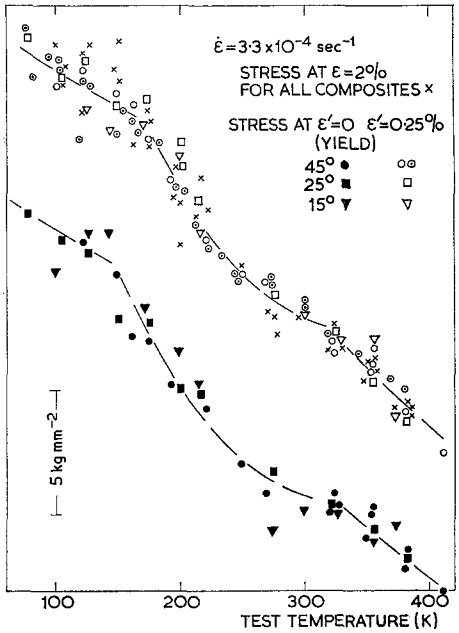

Figure 2 Superposition of stress-temperature relations for various strain levels and composites.

agree with each other, implying first, that the mechanism in samples with orientations between 45 and 15° is the same, and secondly, that the mechanism during elastic deformation is not different from that during inelastic deformation. Also shown in Fig. 2, and pointing to the same conclusions, is the temperature dependence of the yield stress, though owing to the difficulties connected with its exact determination the scatter band is larger. Results for the 8° composite are not included, as their scatter band is especially large.

Finally, it should be emphasized that in each plot, i.e. the λ-T or the σ_ϵ-T curves, the transitions between the three regions appear to occur at essentially the same temperatures, namely 180 and 300 to 340 K.

The conclusion from Part 1, that the deformation of composites is largely controlled by the matrix, has now been further substantiated. It can be deduced from a comparison with results from the unreinforced epoxy. In Fig. 1, the transitions occur at the same temperatures for both reinforced and unreinforced epoxy. However, there still remains the considerable difference in the values of the strain-rate sensitivities and of the stresses. In Part 1 it was argued that there is a stress-raising effect of the reinforcement; since the strain-rate sensitivity generally increases with stress, λ is greater for the composites than for the unreinforced polymer. For a 45° composite this explanation seems quantitatively satisfying (see Section 3.2). Inconsistencies arise, however, when the behaviour of specimens with near-0° orientations are taken into account (Section 3.3).

In the following sections we discuss the observed behaviour in terms of the thermal-activation rate theory originally due to Eyring [4], which has been modified and applied to a variety of deformation processes in both metals and polymers [5-9].

3.2. Epoxy and 45° composites

From rate theory [2, 8] it follows that the strain-rate is given by

$$\dot{\epsilon} = \dot{\epsilon}_0 \exp\left[-\left(\Delta H_0 - \int_0^\tau v_\tau^*\,\mathrm{d}\tau + \int_0^p v_p^*\,\mathrm{d}p\right)\Big/kT\right]\cdot \qquad (2)$$

The pre-exponential factor $\dot{\epsilon}_0$ may include quantities such as a molecular vibrational factor, an entropy term, the density and spacing of participating elements and an orientation factor which relates the local deformation rate to the overall tensile rate. The total activation enthalpy necessary to surmount a thermally activatable barrier is ΔH_0, and it can partly be supplied from external work done by a shear stress $\tau = \sigma/2$, given by the first integral, which contains the shear-stress activation volume v_τ^*. In addition, ΔH_0 can be increased by externally applied pressure p, for which a pressure-activation volume v_p^* is defined. In the present case, for tests at atmospheric pressure, we shall neglect the second integral in Equation 2. The stress-activation volume can then be calculated from

$$v_\tau^* = kT\left[\frac{\partial \ln(\dot{\epsilon}/\dot{\epsilon}_0)}{\partial \tau}\right]_{T,p}\cdot \qquad (3)$$

Normally it is evaluated from stress changes $\Delta\sigma = 2\Delta\tau$ obtained by changing the strain-rate between $\dot{\epsilon}_1$ and $\dot{\epsilon}_2$. Neglecting any possible change in $\dot{\epsilon}_0$ during these strain-rate changes, we can write

$$v_\tau^* = 2kT\left[\frac{\ln(\dot{\epsilon}_2/\dot{\epsilon}_1)}{\Delta\sigma}\right] = \frac{kT}{\lambda} \qquad (4)$$

An indication that $\dot{\epsilon}_{01}$ is approximately equal to $\dot{\epsilon}_{02}$ has been given by tests where no appreciable change of v_τ^* was recorded for increasing ratios $\dot{\epsilon}_2/\dot{\epsilon}_1$ [10].

The stress-activation volumes have been evaluated by means of Equation 4 for the epoxy and for the 45° composite. They are plotted

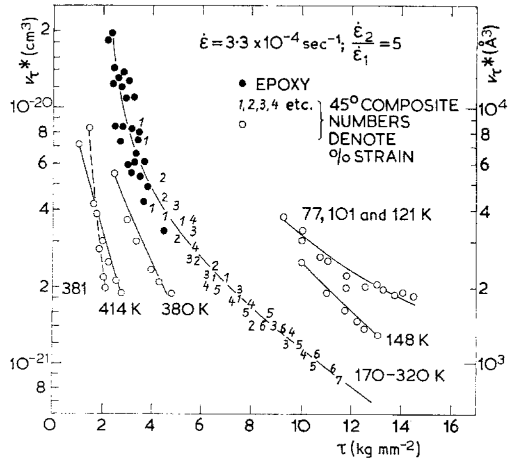

Figure 3 The stress dependence of the stress-activation volume for epoxy and a 45° composite.

against stress in Fig. 3. The following observations can be made:

(a) one essentially continuous curve represents the v_τ*-τ dependence of epoxy and 45° composite in the temperature range 170 to 320K;

(b) in this temperature region the decrease of v_τ* with strain can be described by the same dependence as is given by the general v_τ*-τ relation;

(c) for temperatures above 320K, the curve is shifted to lower stresses;

(d) for temperatures below 170K, the curve is shifted to higher stresses. Finally, another band is formed which contains all results from tests at temperatures below about 120K.

The conclusions drawn from these observations are as follows:

(a) the activation volume is a function of stress only, and not a function of strain;

(b) the distinct v_τ*-τ bands at high, medium, and low stress levels are associated with deformation mechanisms I, II and III, respectively;

(c) the deformation at certain temperatures, for which the v_τ*-τ relations lie between the general bands, is not controlled by the pure processes I, II and III; transition properties are exhibited here. Thus, the deformation process I is undisturbed only below 120K. At 380 to 410K process III is possibly not yet exclusively rate-controlling;

(d) the v_τ*-τ relation for the 45° composite is apparently the continuation of that of the pure epoxy.

Using the equation

$$\Delta H = -kT^2 \left[\frac{\partial \ln(\dot{\epsilon}/\dot{\epsilon}_0)}{\partial \tau}\right]_{T,p} \left[\frac{\partial \tau}{\partial T}\right]_{\dot{\epsilon},p} = -v_\tau{}^* T \left[\frac{\partial \tau}{\partial T}\right]_{\dot{\epsilon},p} \qquad (5)$$

with the slope $\partial\tau/\partial T$ from Fig. 2 and the values of v_τ* from Fig. 3, we can calculate the activation enthalpy. The results are plotted in Fig. 4 to show the dependence of ΔH on the stress for each of the three temperature (or deformation) regions. Extrapolation to zero stress is necessary to obtain the activation enthalpy ΔH_0 which is a characteristic quantity for a thermally activated process. The extrapolation for process I (77 to 120K) is difficult due to the lack of data at low stresses, but it appears that $\Delta H_0{}^{\mathrm{I}}$ may be below 1 eV. The result for region III is also indefinite, since at temperatures below 380K the data may relate to a combination of processes II and III.

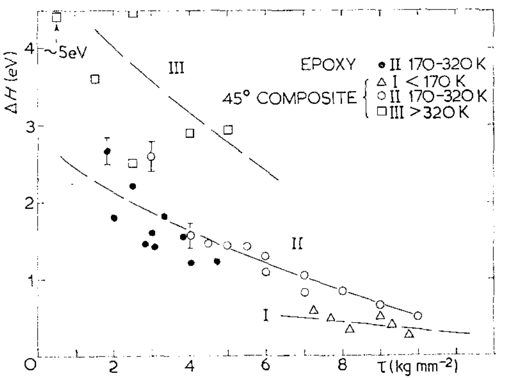

Figure 4 The stress dependence of the activation enthalpy for epoxy and a 45° composite.

However, the result that $\Delta H_0{}^{\mathrm{I}}$ is larger than $\Delta H_0{}^{\mathrm{II}}$ seems self-evident for another reason: process III operates at higher temperatures. For process II we have data from both the epoxy and the 45° composites, and they are complementary. Without too much uncertainty, an activation enthalpy $\Delta H_0{}^{\mathrm{II}}$ around 3 eV is obtained by extrapolation to zero stress.

So far we have calculated the parameters in Equation 2 neglecting any change in the frequency factor $\dot{\epsilon}_0$ during strain-rate changes as well as the term which gives the pressure dependence of the activation enthalpy. Using Equation 2 in the form

$$\dot{\epsilon}_0 = \dot{\epsilon} \exp[\Delta H/kT] \qquad (6)$$

where

$$\Delta H = \Delta H_0 - \int_0^\tau v_\tau{}^* \, d\tau \qquad (7)$$

which is plotted in Fig. 4, we have calculated values of the pre-exponential factor $\dot{\epsilon}_0$ for various conditions of strain, stress and temperature in the deformation range II. As Fig. 5 indicates, $\dot{\epsilon}_0$ is found to vary strongly with stress during a single test; considerable variation with temperature is also evident. However, as the results for composites show clearly and those for the unreinforced epoxy less conclusively, the pre-exponential factor for a certain material appears to be approximately constant at all test temperatures when it is measured at comparable strains.

The theoretical temperature dependence of the stress can be obtained from the combined Equations 6 and 7:

$$\int_0^\tau v_\tau{}^* \, d\tau = \Delta H_0 + kT \ln\left(\frac{\dot{\epsilon}}{\dot{\epsilon}_0}\right) \cdot \qquad (8)$$

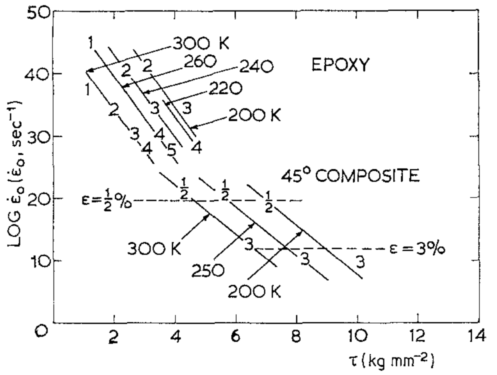

Figure 5 The stress, temperature and strain dependence of the frequency factor. Numbers along the curves denote the strain (in %) for which $\dot{\epsilon}_0$ was calculated.

The right hand side of the equation, representing the work done by the stress τ (the upper limit of the integral) can easily be determined from known constants. On the other hand one can plot, instead of $v_\tau{}^*$ versus τ, the work

$$\int_0^\tau v_\tau{}^* \, \mathrm{d}\tau \text{ versus } \tau\,,$$

and correlate the result of Equation 8, obtained for a certain temperature, with a stress τ. This analysis has been carried out in detail on epoxy [11]. For a mere qualitative treatment it is sufficient to use the simplified form for Equation 7

$$\Delta H = \Delta H_0 - v_\tau{}^* \, \tau \qquad (9)$$

and, instead of Equation 8,

$$\tau = \frac{1}{v_\tau{}^*}\left[\Delta H_0 + kT \ln\left(\frac{\dot{\epsilon}}{\dot{\epsilon}_0}\right)\right] \cdot \qquad (10)$$

When $\dot{\epsilon}$, ΔH_0 and $\dot{\epsilon}_0$ are constant for a certain strain ϵ, and $v_\tau{}^*$ changes non-linearly with stress or temperature, the relation between stress and temperature is not linear, and the temperature sensitivity increases with decreasing temperature. This has indeed been found for the deformation range II in composites (Fig. 2) as well as in epoxy [11]. It disagrees with findings for polycarbonate which have been discussed in the literature [12].

When the rate-controlling process changes between regions I and II, a kink in the temperature dependence of the stress is observed (Fig. 2). At high stresses the activation enthalpies of processes I and II are nearly equal (see Fig. 4) and the increase of the temperature dependence when process II becomes operative is due to the sudden decrease of the activation volume according to Fig. 3. From Equation 10 it follows that the change in slope is inversely proportional to the change of $v_\tau{}^*$ (when the change of ΔH between the two mechanisms can be neglected as appears to be justified at high stresses). In the case of the transition between I and II the ratios amount to 1:2 or 3 for both slopes and the inverse of activation volumes (see Figs. 2 and 3). A similar argument can be applied to justify the kink at the II/III transition.

In a tensile test, in contrast to a torsion test, the material is subjected to a hydrostatic stress component $p = -\sigma/3$. Since σ is the tensile stress, we should discuss the pressure effect in all materials where a temperature variation can change the tensile stress. It appears then that the measured activation volume is not $v_\tau{}^*$ alone but that it also contains a contribution from the change in the hydrostatic stress component. The experimentally determined contribution from external work which reduces ΔH_0 is thus correctly

$$w = \int_0^{\bar{\tau}} \bar{v}^* \, \mathrm{d}\bar{\tau} = \int_0^{\tau} v_\tau{}^* \, \mathrm{d}\tau - \int_0^{p} v_p{}^* \, \mathrm{d}p$$
$$= \int_0^{\sigma/2} v_\tau{}^* \, \mathrm{d}\left(\frac{\sigma}{2}\right) + \int_0^{\sigma/3} v_p{}^* \, \mathrm{d}\left(\frac{\sigma}{3}\right) \cdot \qquad (11)$$

A separation of the measured $\bar{v}^*$ (which we have so far denoted as $v_\tau{}^*$ since we have neglected the pressure term) into the two parts $v_\tau{}^*$ and $v_p{}^*$ could be achieved by comparison with the results of torsion tests, in which the hydrostatic component is zero. The results of such tests would give the correct value $v_\tau{}^*$ alone. If the pressure-activation volume is indeed as small as 100 to 300 Å^3 at room temperature, as quoted in an investigation on polyethylene at high pressures [2], then the stress dependence of the activation volume obtained in the present work is still valid.

3.3. 8, 15 and 25° composites

The stress-activation volumes in range II of composites with orientations of 8, 15 and 25° are given in Fig. 6, together with the curve for 45° orientation. The previous statement, derived from the properties of a 45° composite, that the behaviour of the composite is the extension of that of pure epoxy to higher stress levels, is no longer applicable.

Tests with a different material, in which the reinforcement consists of independent layers of fibres in mutually perpendicular directions

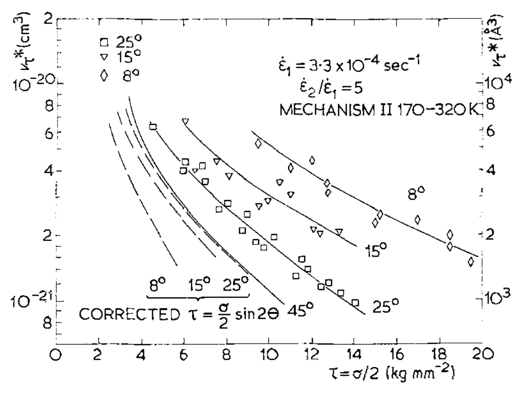

Figure 6 The stress dependence of the activation volume for composites of various orientations.

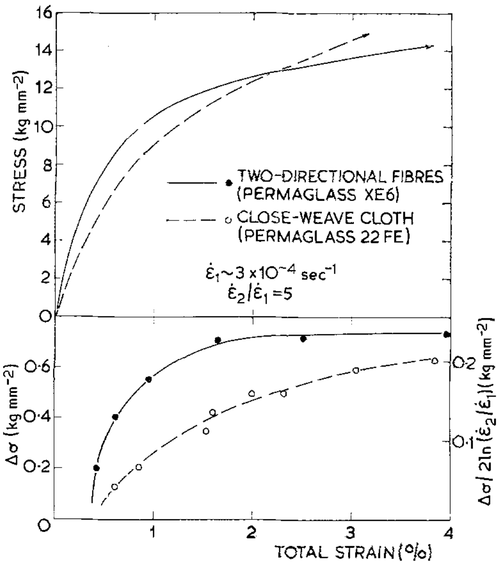

Figure 7 Stress-strain curves and the strain dependence of the strain-rate sensitivity for two different types of reinforced material, tested at 300 K and in 25° orientation.

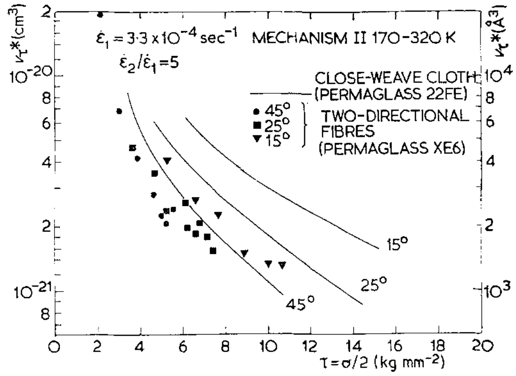

Figure 8 The stress dependence of the activation volume for two different types of reinforced material, tested in various orientations.

("Permaglass XE 6"), show that the high activation volumes for orientations near 0° are not necessarily an inherent property. This particular material has different stress-strain curves: higher initial slope and a more pronounced levelling-off at larger strains, as shown in Fig. 7. Activation volumes for this material stressed at various orientations have been determined in the usual way. The strain dependence of the strain-rate sensitivity is not the same (Fig. 7) but agreement is obtained when (e.g. as in the case of 45° composites) the stress dependences of the activation volumes are compared (Fig. 8). This proves again that the activation volume is not strain-dependent but stress-dependent. What is of special interest in using this new material is the fact that even for a 15° orientation large inelastic strains can be obtained. Activation volumes, determined in the flat portion of such a stress-strain curve are not much different from those of a 45° composite (see Fig. 8).

In composites with fibre orientations other than 45° (as in Fig. 6) the activation volume for a given maximum shear stress is found to be greater than that of the 45° composite. However, it is clear that for these orientations the applied shear stress on planes at 45° to the tensile axis will be partly taken by tensile forces in the fibres. Thus, unless $\theta = 45°$, the shear stress in the epoxy on these planes is less than the value plotted ($\sigma/2$). A more significant value may be the shear stress on planes parallel to the fibres, $(\sigma/2) \sin 2\theta$, since this component of stress cannot be taken by tensile forces in the fibres.* Plotting the results against this stress component, as indicated by the dashed lines in Fig. 6, gives good agreement for all orientations except $\theta = 8°$; at this orientation, because the rate sensitivity is low, the accuracy of the results may be reduced. It appears, therefore, that at least to a first approximation, the rate sensitivity is governed by the shear stress in the epoxy on planes parallel to the fibres. In cases where the bond between fibre and matrix is impaired at an early stage of deformation (see Figs. 7 and 8), the activation volume correlates better with the maximum shear stress $\sigma/2$. This implies that measurements

---

*It is assumed that the shear stress in the fibres accounts for a constant fraction of the total shear stress.

of the rate sensitivity of specimens with fibre orientations other than 45° may be used to give an indication of the extent of the breakdown of the composite material.

4. Conclusions

The analysis of temperature and strain-rate effects has shown that between 77 K and the glass-transition temperature three different deformation mechanisms are operating: mechanism I, between 77 and 120 K, with a zero-stress activation enthalpy ΔH_0^{I} smaller than 1 eV; mechanism II, between 170 and 320 K, with $\Delta H_0^{\mathrm{II}} \sim 3$ eV; and mechanism III, above 380 K, with $\Delta H_0^{\mathrm{III}} > 3$ eV. A detailed analysis of the mechanisms I and III was not possible since each operates only in a small part of the temperature range which has been investigated. More information is available relating to mechanism II. The main results are as follows:

(a) the apparent activation volumes are stress-dependent. They depend on the strain only inasmuch as stress increases with strain.* Whether the activation volume is temperature dependent cannot be determined, since the strain-rate range covered is small;

(b) the apparent activation enthalpy, calculated from Equation 5, is stress dependent;

(c) the pre-exponential factor is different for epoxy and for the composite, but for each material it is approximately independent of temperature or stress when measured at a certain strain during the deformation process. In contrast to the activation volume and enthalpy, the frequency factor $\dot{\epsilon}_0$ is strain dependent;

(d) the flow properties of pure epoxy and the 45° composite can be interpreted as being controlled by the same basic deformation process over a wide range of stresses. The fracture in epoxy, however, appears not to be controlled by the stress alone, but also by the amount of strain. Owing to fibre reinforcement, the deformation of the epoxy is delayed until the bond between fibres and matrix becomes weakened at high stresses; then the deformation of the epoxy sets in, showing activation characteristics as they would occur if those of pure epoxy were extra-polated to higher stresses;

(e) the hydrostatic component of stress may affect the deformation, but in this investigation its contribution could not be separated. The effect is expected to be minor;

(f) in composites with fibre orientations other than 45° the activation volumes are found to be larger if the shear stress is resolved on planes inclined at 45° to the tensile axis. If the fibres are assumed to determine the direction of the maximum shear stress in the matrix, then the actual shear stress has to be calculated by taking into account a correction factor, thus $\tau = (\sigma/2)\sin 2\theta$. If the results are normalized in this way, the $v_\tau^* - \tau$-relations for all orientations fall within a fairly narrow band.

Our description differs from previous ones by the idea of a stress-dependent activation volume. Instead of the stress integral

$$\int_0^\tau v_\tau^* \, \mathrm{d}\tau$$

in Equation 2, the Eyring equation assumes that v_τ^* is independent of τ, so that the activation enthalpy varies linearly with the applied stress. We believe that in the long run this adds to the limitation of the Eyring analysis to explain, for instance, why the necessary external work varies with temperature. In our analysis, the differences which come from various degrees of straining, for instance, can be traced back to variations of the pre-exponential factor $\dot{\epsilon}_0$. At present, and until detailed models for deformation mechanisms are developed, we can only suppose that the change of the pre-exponential factor is related to the activation entropy, the distance swept, the length of the participating element of the polymer molecules, etc. The aim of such an analysis is to give a molecular interpretation of the deformation process.

Our results can be compared with others given in various recent papers, which are usually derived from an Eyring analysis. The activation volumes for PVC, PC and similar materials are generally 8000 to 10 000 Å^3, comparable to the values we have obtained at room temperature. The activation volume has been related to the size of the statistical random link in the polymer chain [14]. Activation energies for PC and PVC in the low-temperature region are 0.4 to 0.6 eV and in the medium region 3 to 3.3 eV [12, 15]. The deformation at lowest temperatures was said to be governed by the same molecular movement which gives rise to the β-damping peak in PC and PVC [12, 16]. For PC or epoxy

---

*A decrease in activation volume with increasing stress has also been reported for nylon-epoxide adhesive [13]. This decrease was, however, attributed to changes in the deformation mechanism with increasing strain.

resins, the β-transition was interpreted as caused by the mobility of the di-ether linkage of bisphenol A, when the backbone chain is in a stretched configuration [17]. A region, perhaps corresponding to III in our investigation, was only mentioned twice before [13, 18]. Of these two papers, one states that the data do not fit Eyring's theory [18]. The other [13] implies that at high temperatures the looseness in the chain structure is increased and that local melting starts at weak links, resulting in viscous flow of the liquid (activation energies, calculated for these temperatures, do not agree with our results for III).

A recent publication [19] also introduced the idea of a stress-dependent activation volume and enthalpy. There, these quantities have been determined by both strain-rate and temperature cycling on highly-strained PE which was partly crystalline. The same sectioning into three regions, but different transition temperatures, were found. Although the activation enthalpies of these regions (also denoted with I, II and III, starting with I at lowest temperature) were comparable, there are a few points which seem to testify against any identity with our investigation. It was claimed from evidence of other work that the I-II transition is due to the glass transition of the amorphous phase. While region I was insufficiently explained by "deformation of the amorphous phase" whose exact mechanism remains uncertain, the rate-controlling process above the glass transition, i.e. in region II, was assumed to be governed by the thermally activated scission of tie molecules. The results for region III were related to changes in microstructure during the evaluation of activation parameters by means of temperature cycles, e.g. with tie molecules slipping out of the crystal blocks.

Despite similar activation enthalpies we can apparently not consider the three deformation regions for PE and epoxy as equivalent. Our epoxy has a glass-transition temperature of 150° C, so that the deformation in regions I, II and III occurs completely in the brittle state. One indication of the difference is the stress activation volumes of PE and epoxy which differ by a factor of 3. The explanation for region III in the case of PE is based on the crystallinity of the material and on its changes during temperature cycles. Neither of these prerequisites applies in our case: first, epoxy is amorphous, and second, region III was found in our investigation by rapid strain-rate changes during which crystal growth, if it were present in epoxy at all, would not have been possible.

The present study has, it is hoped, provided data which can be used in further evaluating the applicability of models of the deformation mechanisms controlling the strength properties of glass-reinforced epoxy, as well as in considering the use of such materials to replace conventional metals and alloys in engineering structures.

Acknowledgement

We would like to thank Permali, Ltd, Gloucester, and particularly Mr Gaunt, for supplying us with the normal and specially produced materials.

References

1. E. PINK and J. D. CAMPBELL, *J. Mater. Sci.* **9** (1974) 658.
2. L. A. DAVIS and C. A. PAMPILLO, *J. Appl. Phys.* **42** (1971) 4659.
3. E. PINK and J. D. CAMPBELL, Report No. 1028/71 Dept. Engineering Science, University of Oxford, 1971.
4. H. EYRING, *J. Chem. Phys.* **4** (1936) 283.
5. H. CONRAD, *J. Metals* **16** (1964) 582.
6. O. ISHAI, *J. Appl. Polymer Sci.* **11** (1967) 963.
7. *Idem, ibid* **11** (1967) 1863.
8. J. C. M. LI, *Canad. J. Phys.* **45** (1967) 493.
9. A. G. EVANS and R. D. RAWLINGS, *Phys. Stat. Sol.* **34** (1969) 9.
10. E. PINK and J. D. CAMPBELL, Report No. 1040/72, Dept. Engineering Science, University of Oxford, 1972.
11. *Idem*, to be published *Mat. Sci. Eng.* (1974).
12. C. BAUWENS-CROWET, J.-C. BAUWENS and G. HOMÈS, *J. Mater. Sci.* **7** (1972) 176.
13. H. K. SHEN and J. L. RUTHERFORD, *Mat. Sci. Eng.* **9** (1972) 323.
14. I. M. WARD, *J. Mater. Sci.* **6** (1971) 1397.
15. C. BAUWENS-CROWET, J.-C. BAUWENS and G. HOMÈS, *J. Polymer Sci. A* **2-7** (1969) 735.
16. J.-C. BAUWENS, *ibid C* **33** (1971) 123.
17. E. CUDDIHY and J. MOACANIN, "Epoxy Resins" (Amer. Chem. Soc., 1970) p. 96.
18. J.-C. BAUWENS, C. BAUWENS-CROWET and G. HOMÈS, *J. Polymer Sci. A* **2-7** (1969) 1745.
19. C. A. PAMPILLO and L. A. DAVIS, *J. Appl. Phys.* **43** (1972) 4277.

Received 25 January and accepted 3 October 1973.