

## Stress-Strain Behavior of Rubber-Reinforced Glassy Polymers\*

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### Synopsis

The main objective of this work has been to develop some rational hypothesis for the toughness of rubber-modified glassy polymers as defined by the energy to rupture in a tensile measurement. It is shown that the rigid phase yields and that the toughness is a result of the large energy absorption involved in the cold drawing of the matrix. The rubber phase acts principally to induce a yielding in the matrix. A triaxial stress field in the environment of the dispersed particles results in local increases in free volume which aid in the initiation of cold drawing and provides the rubber with sufficient breaking stress to prevent premature crack propagation.

### INTRODUCTION

At environmental temperatures well below their glassy transition,  $T_g$ , some amorphous polymers, e.g., polystyrene, exhibit brittle or glassy fracture. However, other amorphous or slightly crystalline polymers, such as quenched poly(ethylene terephthalate), poly(vinyl chloride), and polycarbonates are highly extensible, that is, they cold-draw. (The distinction between a ductile and brittle response is not absolute, since brittle fracture in organic resins as unoriented polystyrene does involve viscous flow on a micro-scale.<sup>1</sup> Also, a yield point may often be introduced by various conditioning treatments, such as cooling rapidly through the glass temperature.<sup>2</sup>) These ductile polymers are tough and resistant to impact. Polyblending of the former with finely dispersed, crosslinked but partly compatible rubber particles results in a dramatic and remarkable increase in toughness as assayed by a tensile curve of stress versus strain. The main objective of this work has been to develop some rational hypothesis for the energy to rupture in such a tensile experiment.

Intimately associated with this is the problem of how extremes of macroscopic behavior, i.e., brittle fracture after a few per cent extension contrasted with cold drawing of several hundred per cent, are both observed for pure polymers below their  $T_g$ . Therefore, it is principally to the problems of the initiation of cold drawing and the role of dispersed rubber in this connection that this work is directed.

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## BACKGROUND

Important advances toward elucidating the role of the rubber in preventing crack growth were made by Merz, Claver, and Baer<sup>3</sup> for the system butadiene-styrene rubber in a polystyrene matrix. These authors argued, however, that the rubber particle (butadiene-styrene) acted so as to hold the fracture surfaces together. As long as the tensile stress did not exceed the adhesive strength the particle would not be abstracted from the matrix.

Since the rubber is capable of extensive elongation, it was argued that the rubber phase absorbed more energy during tensile failure than an equivalent volume of continuous phase, and that in this manner the impact strength of the mixture was increased. Merz et al. do not explain, however, how high elongation is achieved in the rubber phase when the composite fractures at about 15% strain. One way in which this may be realized is discussed in the Appendix; also, it is shown there that this model leads to an energy absorption of the rubber phase which is only about  $1/10$  of the total energy absorbed by the whole polyblend. Clearly, the rubber in itself does not absorb sufficient energy, according to this model, to account for the toughness of the composite.

In their studies of polystyrene modified with crosslinked butadiene-styrene rubber, Schmitt and Keskkula,<sup>4</sup> report the formation of myriad cracks around the rubber particles when the sample is stretched. Formation of a crack surface requires energy; Schmitt and Keskkula argued, therefore, that the formation of many cracks would account for the energy absorption of polyblends. In addition, these authors allude to the role of the rubber particle as preventing crack propagation. No details at all are provided, however, as to how these mechanisms account for the overall load-deformation curve, particularly the observed large elongations.

In discussing the toughness of polyblends, the problem of how this property shall be assayed cannot be ignored. High-speed tensile, drop dart, notched Izod, etc., are commonly used tests but are not equivalent. A high-rate tension test provides information as to the entire load-deformation curve up to fracture over a range of test speeds. Toughness may be defined as the total area under the stress-strain curve, and this is the definition which will be employed throughout this report.

It should be pointed out, in this connection, that the use of a notched sample alters markedly the total energy to fracture. As has been shown by Strella,<sup>5</sup> the energy to fracture of a notched sample in the Izod impact or tensile test is principally the energy to yield. Thus, the energy from yield to fracture, which is the major portion of the energy to failure for an unnotched sample, accounts for only a small portion of the energy to fail of a notched sample. A notch may also behave as a Griffith crack and lower the yield strength and fracture strength. Use of a notched sample in an Izod test thereby discloses the crack sensitivity of a given material.

## EXPERIMENTAL

## Poisson Ratio Measurements

With the use of SR4 strain elements affixed to a U-gauge as depicted in Figure 1, measurements of longitudinal and lateral strain were obtained simultaneously during stretching of various samples with an Instron. The outputs of the gauges were so arranged that the pen and chart were each driven by one of the gauges. Sensitivities used were  $6.0 \times 10^{-4}$  in./in. of chart for the lateral dimension and  $5.8 \times 10^{-3}$  in./in. of chart for the longitudinal dimension. Poisson's ratio  $\mu$  was then calculated from the slopes of the plots simply as

$$\mu = -\epsilon_y/\epsilon_x$$

More accurately,

$$\epsilon = \int_{l_0}^l dl/l = \ln(1 + \epsilon)$$

so that

$$\mu = -\ln(1 + \epsilon_y)/\ln(1 + \epsilon_x)$$

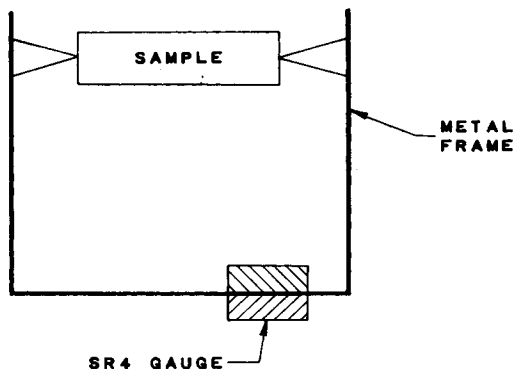


Fig. 1. Schematic diagram of U-gauge used for Poisson ratio measurements.

TABLE I

Results of Measurements on Poisson's Ratio for S/AN (Styrene-Acrylonitrile) copolymer and S/AN Polyblends

| Sample                                     | Treatment    | No. samples | $\mu = -(\epsilon_y/\epsilon_x)$ |
|--|--------------|-------------|----------------------------------|
| 1 S/AN (72/28)                             | Annealed     | 3           | $0.32 \pm 0.01$                  |
|  | Not annealed | 4           | $0.33 \pm 0.02$                  |
| 2 S/AN (70/30)                             | Annealed     | 3           | $0.32 \pm 0.01$                  |
|  | Not annealed | 4           | $0.36_5 \pm 0.02_5$              |
| 3 S/AN-rubber polyblend (10% rubber graft) | Annealed     | 3           | $0.33 \pm 0.01$                  |
|  | Not annealed | 4           | $0.36_5 \pm 0.02_5$              |
| 4 S/AN-rubber polyblend (25% rubber graft) | Annealed     | 2           | $0.35 \pm 0.00$                  |
|  | Not annealed | 4           | $0.375 \pm 0.01$                 |

but this makes a small correction at high strains. Some values of these measurements obtained are shown in Tables I and II.

TABLE II  
Results of Poisson Ratio Measurements on Rigid PVC at Different Strain Levels

| Strain ( $\Delta l/l_0$ )100, % | No. samples | $\mu = (-\epsilon_y/\epsilon_x)$ |
|---------------------------------|-------------|----------------------------------|
| ~5                              | 4           | 0.41 ± 0.01                      |
| 8.6                             | 1           | 0.40 <sub>8</sub>                |
| 11.5                            | 1           | 0.37 <sub>0</sub>                |

### High-Rate Tensile Measurements

Tensile measurements were carried out with a hydraulically operated tensile tester capable of speeds up to 5000 in./min. This device has already been described<sup>1</sup> and will, therefore, not be discussed here.

### Dielectric Measurements

A limited number of dielectric measurements were performed on poly(vinyl chloride) (PVC) with a General Radio Corp. Bridge, Model 716C. The sample was coated on opposite faces with a conductive silver lacquer, and, by using a parallel arrangement of the sample with a calibrated capacitor, changes in sample capacitance due to either temperature or strain could be compensated with the calibrated capacitor. In this simple fashion, capacitance changes of 0.01  $\mu\mu\text{fd.}$  could be read directly.

For measurements of capacitance change with strain, a precise stretching jig previously described<sup>7</sup> was employed.

To compensate for dimensional changes occurring with strain, the following analysis was employed. For a parallel plate condenser (the sample), the capacitance  $C$  is given by

$$C = B\epsilon'A/t \quad (1)$$

where  $t$  is the thickness,  $\epsilon'$  the dielectric constant,  $A$  the area of one face, and  $B$  a constant. From the total differential for  $\Delta C$ , it can be shown that

$$(\Delta C t / BA) + \epsilon'[(\Delta t / t) - (\Delta A / A)] = \Delta \epsilon'$$

However, from the definition of Poisson's ratio we have

$$(\Delta t / t) - (\Delta A / A) \simeq -\epsilon_x$$

where  $\epsilon_x$  is the tensile strain in the  $x$  direction; combining equations and simplifying leads to

$$(\Delta C / C) - \epsilon_x = \Delta \epsilon' / \epsilon' \quad (2)$$

Equation (2) provides a simple relation for calculating  $\Delta \epsilon' / \epsilon'$  corrected for all dimensional changes accompanying and including the strain,  $\epsilon_x$ .

For measurements at different temperatures, an oven regulated to  $\pm 0.1^\circ\text{C.}$  was employed. Sample temperature was measured by placing a

thermocouple in contact with the specimen. Corrections for thickness and area changes due to the thermal expansion were made by use of the known coefficient of thermal expansion of PVC.

## RESULTS AND DISCUSSION

### Additivity of Properties

The tensile behavior at 50 in./min. test rate of pure S/AN copolymer, a grafted rubber (RG) and a 50/50 blend of these two are shown in Figure 2. Test samples were fabricated by compression molding, the blend and RG samples having first been hot mill-rolled. Figure 3 shows the ultimate strength and elongation of these samples. It is observed that the ultimate elongation and strength approximate a linear function of the RG content. On the other hand, the initial modulus is logarithmically dependent of the

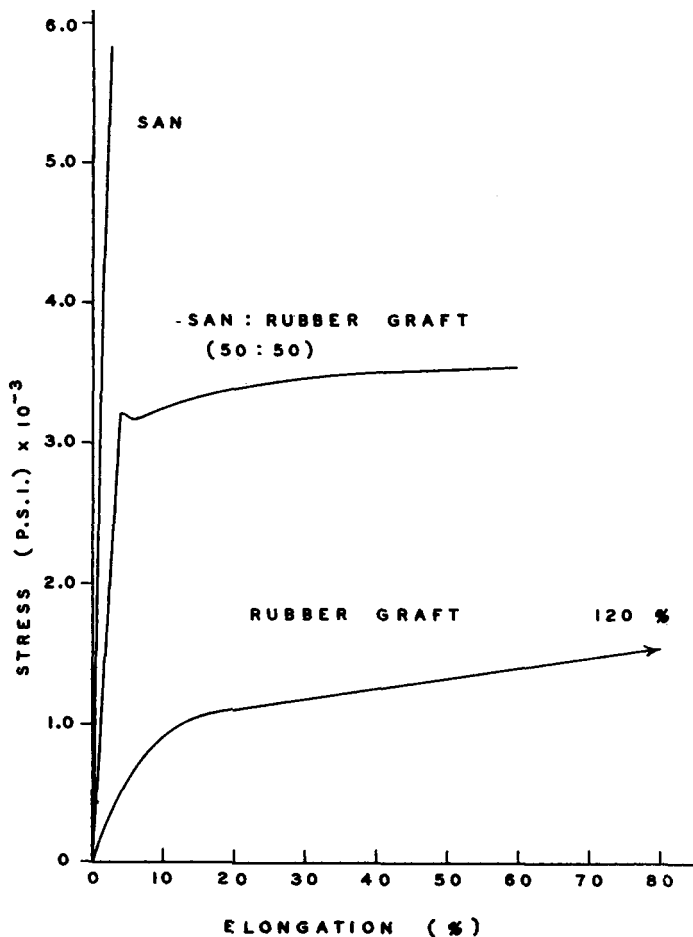


Fig. 2. Tensile curves of S/AN, grafted rubber, and a 50/50 polyblend of the two.

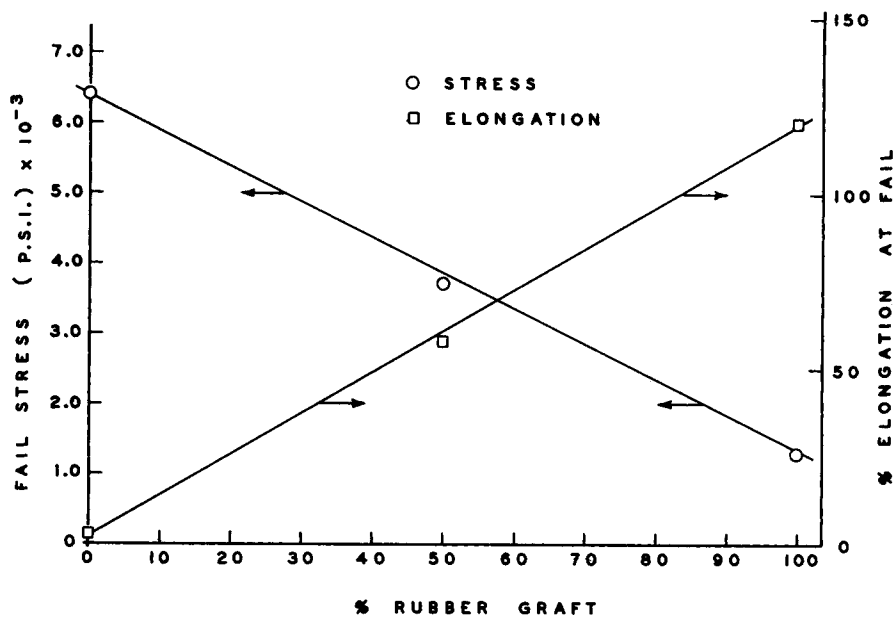


Fig. 3. Ultimate elongation and fail stress for S/AN-rubber graft polyblends for pure S/AN, pure rubber graft, and a 50/50 polyblend.

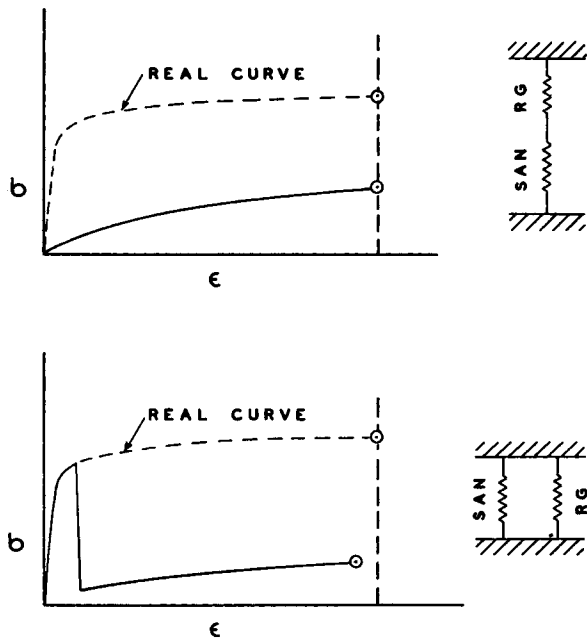


Fig. 4. Comparison of real tensile curves with theoretical curves (based on models) of a 50/50 S/AN-rubber graft polyblend.

composition. (The last-mentioned relation is believed to be analogous to the modulus of a rigid foam whose rigidity is exponentially dependent on foam density. In this connection, the rubber particle in the polyblend approximates the void of a foam.)

If one considers the blend to be representable by an S/AN spring in series with a RG spring, then the hypothetical tensile curve shown in Figure 4 is obtained. It is observed that this model fits the final elongation but does not match the initial yield and final strength. A model in which the two springs are in parallel approximates the yield and final elongation but not the final strength. Taking combinations of these models still did not permit the matching of the curves.

However, if we assign to the rigid phase in the presence of the rubber the ability to yield, then a model can be constructed which fits the observed behavior. This model requires survival of the matrix as a continuous phase up to the point of final rupture. To substantiate this view it remains to be shown (a) that drawing does in fact occur in the matrix and (b) how the rubber might serve to make this possible.

### Matrix Orientation on Straining

To demonstrate that molecular orientation of the matrix rather than rupture takes place when the polyblend is strained beyond the yield point, measurements were carried out on the extent of strain recovery resulting from heat treatment. After prestraining beyond the yield point, strain recovery was allowed to take place for 2 hr. at 150°C. As a control, virgin polymer was subjected to the same thermal history. With the use of dumbbell specimens with about a 3.5-in. gauge section, it was found that the central portion of the neck of the virgin (unstrained) samples decreased in length by a factor of about 0.85–0.95, presumably as a consequence of the orientation in the piece resulting from fabrication by injection molding.

The central necked portion of similar dumbbells prestrained approximately 50% shrank by a factor of 0.65–0.68 after an identical heat treatment. If we assume that all of the straining imposed on the sample occurred in the necked portion, then the 3.5-in. gauge section would have extended roughly to 5.2 in. Return to the original length corresponds to a reduction in length by a factor of 0.67, in agreement with the observed value. This rough calculation neglects the small additional shrinkage to be expected from the orientation due to molding.

It is concluded on the basis of the observations on several sets of samples, therefore, that the major portion, if not nearly all, of the straining of the polyblend is recoverable by thermal treatment above the glass transition temperature  $T_g$  of S/AN. This recovery is to be expected when the sample has become molecularly oriented during prestraining and, as a consequence, develops a rubberlike retractive stress above its  $T_g$ .

### Microscopic Examination of Strained Materials

In order to determine whether the reinforcement process could be viewed microscopically, examinations were made of commercial polyblends containing relatively large disperse rubber particles.

Films of two different commercial ABS polymers were compression-molded to a thickness of about 0.5 and 1.5 mils, respectively. A small specimen from each sheet was then drawn by hand to produce a fissure at one edge; photographs were taken of various portions of the specimen viewed with oil immersion lenses at 1000 $\times$  and 1500 $\times$  magnification for one sample and at 540 $\times$  and 820 $\times$  magnification for the other.

Observations made from these photographs are as follows. (1) Some disperse phase particles present were 1–2  $\mu$  in diameter. (2) In regions toward the end of the fissure, elongated rubber particles were seen with axial ratios of about 3. The fact that these rubber particles were permanently deformed confirmed the view that the matrix had yielded and had become permanently oriented. The photographs also indicate involvement of the entire rubber particle and not a small portion of it as indicated by the Merz model. No cracks are evident around these elongated particles. Further evidence of ductility in the matrix is seen in some photographs where highly drawn regions of matrix are observable near a void with no rubber evident in the drawn matter.

In summary, it appears from these photographs that the entire matrix yielded, and that extensive matrix and disperse phase elongation occurred without evidence of visible cracks.

### Initiation of Cold Drawing

The preceding sections present evidence for the occurrence of localized cold drawing in the matrix of rubber containing polyblends. Further, it is this cold drawing which has been shown to account for the extensive elongation exhibited in Figure 1. A question is raised, therefore, as to how a rubber particle might influence the ability of the surrounding matrix to cold-draw. Further, any attempt to answer this must be subject to our understanding of the general mechanism of cold drawing in rigid thermoplastic materials. Two factors appear of importance with regard to the possible role of the rubber in this connection: (a) prevention of premature crack propagation and (b) introduction of local density fluctuations. Both of these are related to the development of complex stresses in the neighborhood of dispersed rubber particles.

#### *Cold Drawing in Glassy Polymers*

In a ductile response, a neck is initiated and extensive orientation takes place at this site. Since the polymer molecules are essentially immobile at temperatures below  $T_g$ , it has been of long-standing interest to account for the high extensibility and large-scale molecular motion that is observed. Two theories advanced in the past are: (a) that sufficient heat generation



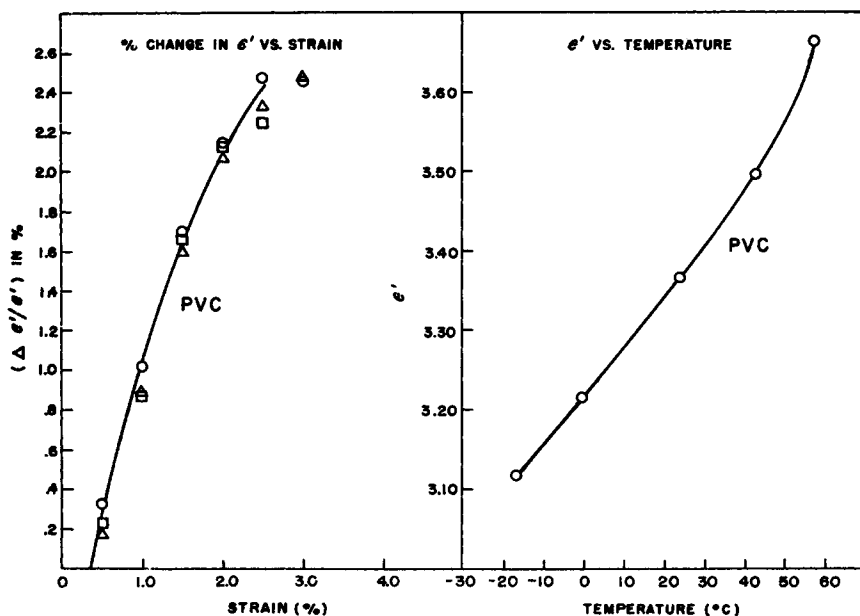


Fig. 5. Change of dielectric constant vs. strain and dependence of dielectric constant on temperature for PVC.

due to working occurs<sup>8,9</sup> to raise the local temperature above  $T_g$  and thereby induce rubberlike behavior and (b) that the  $T_g$  is lowered by the dilational strain induced by the applied stress.<sup>10,11</sup> Though neither theory is completely satisfactory, the latter appears to be more appropriate to the initiation of cold drawing and is also more amenable to treatment. As pointed out by Ferry and Stratton,<sup>12</sup> an increase of the fractional free volume of the polymer effectively shifts the time scale of relaxation and creep processes. Thus, when a polymer whose Poisson's ratio  $\mu$  is less than  $1/2$  is subjected to simple strain  $\epsilon$ , a dilation occurs according to the relation

$$\partial\bar{v}/\partial\epsilon = (1 - 2\mu) \quad (3)$$

where  $\bar{v}$  refers to specific volume. For PVC, we find Poisson's ratio to be about 0.4 and to remain almost unchanged up to about 5% strain, the yield strain (Table II). This corresponds, according to the above relation, to a change of specific volume of 0.01. If we equate this volume change to the volume change that occurs on raising the sample temperature, we obtain

$$\epsilon(1 - 2\mu) = \alpha\Delta T \quad (4)$$

where  $\alpha$  is the coefficient of volume expansion. This corresponds to as much as a 50°C. shift in  $T_g$ . Thus, the dilation on tensile strain does correspond to the volume changes occurring between room temperature and the glass temperature. The quantitative predictions of this equation are subject to question, however, because of a lack of consensus as to the partitioning of the volume increase between free volume and occupied volume.<sup>13</sup>

In a further attack on the effect of strain in the glassy state, dielectric measurements were carried out on poly(vinyl chloride) as described in the experimental section.

The resulting data, shown in Figure 5, demonstrate that an increase of the dielectric constant  $\epsilon'$  of PVC occurs with tensile strain. This observation provides additional tentative experimental evidence that the effective temperature of a sample is shifted closer to the glass transition temperature by tensile strain. From a knowledge of the increase of  $\epsilon'$  with temperature, and strain, an equivalence is noted between 1% strain and about a  $+8^\circ\text{C}$ . shift in effective temperature. The influence of several per cent strain up to the yield point brings the effective sample temperature close to the glass transition, in agreement with the previous arguments based on volume considerations.

When the effective sample temperature is near the glass temperature, a condition favorable to the onset of cold drawing is established, thus, a local density fluctuation which results in a local modulus fluctuation may trigger the onset of a neck. From separate computations, it can be shown that the straining of a specimen containing modulus (or volume) fluctuations can lead to a yield point if some softening mechanism is at hand whereby the modulus diminishes with increased strain, as, for instance, the kind outlined above in our discussion of tensile strain and free volume. Such a situation leads to an accelerating deformation at the locale of the density fluctuation.

For S/AN copolymer which has a  $T_g$  of about  $115^\circ\text{C}$ ., some  $50^\circ\text{C}$ . higher than PVC, and which possesses considerably less total strain to the yield point, it appears that the increase of free volume and hence the shift of effective temperature is inadequate to establish a condition favorable to the formation of a neck.

However, in a polyblend, the presence of a dispersed rubber phase could introduce additional local increases in free volume and, thereby, aid in the triggering or initiation of cold drawing. These density fluctuations may arise from several sources: (a) differences in shrinkage of the two phases on cooling after processing; (b) hydrostatic tension arising during strain from differences in Poisson's ratio of the two phases (i.e., the matrix tends to increase its volume during strain while the rubber phase with  $\mu \simeq 0.5$  tends not to); (c) stress concentrations in the matrix arising from the rubber functioning as an inclusion.

Let us discuss in greater detail the hydrostatic tension during strain. Consider now the rubber-modified S/AN system. From Table I it is seen that  $\mu$  is about 0.32 for S/AN. Thus S/AN will expand when subjected to a tensile strain. However, rubber will exhibit a negligible expansion because of a Poisson's ratio of  $\sim 0.5$ . If the rubber were poorly bonded to the matrix, then on straining a blend, the expanding matrix would separate from the rubber; voids would result. However, there is no evidence for the formation of voids at low strain levels.

In the absence of void formation, straining a blend must cause the matrix and rubber particle to exert multiaxial stresses on each other. These

stresses result in an additional expansion of the matrix and/or an expansion of the rubber phase. This argument is believed to be consistent with the result that the  $\mu$  value for both S/AN and polyblend are about the same (Table I).

A rough estimate of the possible increase in matrix expansion may be arrived at in the following manner. Assume the blend consists of a unit cube of S/AN in which is centrally imbedded a cube of rubber. The volume of the rubber is  $1/3$  that of the S/AN. Now for a given tensile strain  $\epsilon$  on the total sample,  $\Delta V$  for the blend (using the approximate value of  $\mu$  of 0.35) is  $0.3\epsilon$ . Now if the rubber expands proportionately then the net expansion of the S/AN matrix is  $(3/4)0.3\epsilon$ . However, if the rubber does not expand, and assuming no void formation to occur, the expansion of the S/AN matrix becomes  $(3/4)0.3\epsilon + (1/4)0.3\epsilon$ . This gain in matrix volume expansion would be localized to sections lateral to the rubber. As a result of this effect, the local lateral volume expansion can be at least twice that of a pure S/AN material subjected to the same tensile strain. Hence, the lowering of  $T_g$  would, in the localized volume, be twice that for pure S/AN under the same strain. It is thus shown that the rubber can function to further increase the S/AN free volume and thus lower the  $T_g$  of the adjacent matrix by an amount sufficient to allow cold-drawing in this localized region.

#### *Strength of Rubber and Crack Propagation*

The above calculation presumes that the rubber itself, as well as the bond to the matrix, is sufficiently strong to withstand the forces placed on it. Now, the tensile strength of the rubber is considerably poorer than the matrix. However, as is seen from the above discussion the rubber is not subjected solely to a tensile stress but actually experiences a triaxial tension. Under a triaxial stress field, if one accepts the constant strain energy criterion\* that the energy to yield is a constant, then the strength or failure point under a combined stress field is:

$$\sigma_{yp}^2/2E = (1/2E) (\sigma_x^2 + \sigma_y^2 + \sigma_z^2) - (\mu/E) (\sigma_x\sigma_y + \sigma_x\sigma_z + \sigma_y\sigma_z) \quad (5)$$

where  $\sigma_{yp}$  is the yield point or strength in simple tension;  $\sigma_x$ ,  $\sigma_y$ ,  $\sigma_z$  are principal stresses; and  $E$  is Young's modulus.

For a pure hydrostatic tension,  $\sigma_x = \sigma_y = \sigma_z = \sigma$ ; thus from eq. (5)

$$\sigma_{yp}^2 = 3(1 - 2\mu)\sigma^2$$

Now if  $\mu$  for the rubber is estimated to be 0.499, then

$$\sigma = 40\sigma_{yp}$$

Under a pure hydrostatic tension the rubber will thus withstand a stress up to 40 times its yield strength. Actually the stress field on a rubber

\* The Von Mises criterion which appears to fit biaxial data on plastics is the limit of the constant strain energy expression as  $\mu \rightarrow 1/2$ .<sup>14</sup>

particle in an S/AN matrix will not necessarily be symmetric, that is  $\sigma_x \neq \sigma_y \neq \sigma_z$ . However, if  $\sigma_x, \sigma_y,$  and  $\sigma_z > 0$ , then the strength under such a triaxial stress will be greater than the uniaxial tensile strength. It would be desirable to know the exact stress field around the rubber particle.

On the basis of the above it is believed that because of triaxial stressing, the rubber not only is strong enough to withstand the dilational forces involved in the tensile straining of polyblends, but that it is also strong enough to prevent a neighboring crack from propagating. By not allowing premature crack propagation, the allowable tensile strain on the matrix is increased. It seems entirely reasonable that if the rubber particle served to limit the flaw size or growth in polystyrene, the fracture stress would remain high, as shown by Berry.<sup>1</sup> With an enhanced fracture stress, the process of yielding is further favored over and above the considerations previously advanced with regard to the lowering of the  $T_g$ .

### SUMMARY AND CONCLUSIONS

The main objective of this work has been to develop some rational hypothesis for the toughness of rubber polyblends as defined by the energy to rupture in a tensile measurement. It is shown, both by induction and by direct experimental evidence, that the S/AN rigid phase yields and cold draws. It is thereby established that the toughness of a polyblend is a result of the large energy absorption involved in the cold drawing of the matrix.

The presence of inherent flaws or cracks limits the strength and ductility of some glassy polymers, whereas others are known to cold-draw. It is believed, therefore, that the rubber phase in a polyblend acts principally to induce a yielding in the S/AN matrix. There are at least two factors which contribute to this yielding. (a) By limiting the crack growth, the rubber phase effectively strengthens the rigid phase, prevents premature fracture, and permits yielding. In this connection, it is believed that under a tensile strain, the difference in the Poisson's ratio of the rubber and matrix phases results in a triaxial stress field in the environment of the rubber particle. Under such a triaxial stress field, the rubber particle possesses a strength which is greater than in tension and which is sufficient to prevent crack propagation. (b) Secondly, the triaxial field will result in an increase in the free volume of the SAN matrix adjacent to the rubber particle. This, combined with other factors, causes an increase in free volume sufficient to lower the glass transition temperature of S/AN to the point where large-scale molecular motion or cold drawing is further favored.

### APPENDIX

In the model advanced by Merz, Claver, and Baer,<sup>3</sup> the rubber particles serve to hold together fracture surfaces separated in tension. The tensile energy absorbed by the disperse phase was believed to be larger than ab-

sorbed by an equivalent volume of continuous phase on the basis of the tensile curves for both matrix and rubber.

According to the tensile curves given in the publication cited the rubber phase can absorb more energy only if it becomes highly stretched. Apart from the fact that the stress on the rubber is more complex than simple uniaxial tension, it is of interest to compare the total energy absorbed by the composite with the possible energy absorption by the rubber. Firstly, for the rubber phase to absorb maximum energy, it will be assumed that it is taken to failure. Thus, in accordance with the data given, let us take a sample containing 10% rubber and having the dimensions of a 1-in. cube. We may consider the cube to consist of 0.9 in.<sup>3</sup> of matrix and 0.1 in.<sup>3</sup> of rubber phase.

Let us assume that the sample ruptures when the rubber is taken to its reported ultimate elongation of 1900%. Since the ultimate elongation of the composite is only 15%, then the effective gauge length,  $l_0$ , or undeformed length of the rubber is given by the relation

$$100(\Delta l/l_0) = 1900$$

but  $\Delta l = 0.15$ ; hence,  $l_0 = 0.008$  in. On this basis it is seen that only 8% of the rubber is involved in fracture.

The total amount of energy absorbed by the rubber phase may be calculated from the tensile curve and results in 54 in.-lb. This may be compared with the energy of 400 in.-lb. absorbed by the composite.

Thus, the energy absorbed by the rubber according to the model is only about one-tenth the total energy absorbed by the composite.

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### Résumé

Le principal objectif de ce travail a été de développer certaines hypothèses rationnelles concernant la résistance au choc de polymères vitreux additionnés de caoutchouc, définie par l'énergie de rupture lors d'une mesure de tension. On montre que la déformation anélastique de la phase rigide et que la résistance au choc résultent de la grande énergie d'absorption mise en oeuvre dans l'étréage à froid de la matrice. La phase caoutchouteuse intervient principalement pour induire une déformation anélastique dans la matrice. Une tension triaxiale appliquée aux environs de particules dispersées provoque des augmentations locales du volume libre qui favorisent l'initiation de l'étréage à froid et procurent au caoutchouc une tension de rupture suffisante pour prévenir la propagation prématurée de fissures.

### Zusammenfassung

Das Hauptziel der vorliegenden Arbeit war die Entwicklung einer rationellen Hypothese für die durch die Reissenergie bei einem Zugversuch definierten Zähigkeit kautschukmodifizierter, glasartiger Polymerer. Es wird gezeigt, dass die starre Phase zu fließen beginnt und dass die Zähigkeit durch die Grösse, bei der kalten Dehnung der Matrix auftretende Energieabsorption bedingt ist. Die Hauptwirkung der Kautschukphase besteht in der Induktion einer Fließfähigkeit in der Matrix. Ein triaxiales Spannungsfeld in der Umgebung der dispergierten Teilchen führt zu einer lokalen Zunahme des freien Volumens, was die Initiierung der Kaltreckung erleichtert und dem Kautschuk eine genügend hohe Bruchspannung zur Vermeidung eines vorzeitigen Risswachstums verleiht.

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