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MECHANICAL PROPERTIES OF POLYCARBONATE-POLYSULFONE AND POLYCARBONATE-POLYETHERIMIDE BLENDS

By
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13. ABSTRACT (*Maximum 200 words*)

This project is an attempt to produce a ballistic impact resistant transparent eye armor with improved scratch and chemical resistance over polycarbonate. The first stage of the project examines the thermal and mechanical properties of the polymers. Five, ten, and twenty percent by weight blends of polysulfone in polycarbonate and polyetherimide in polycarbonate were produced by melt blending using a counter-rotating twin screw extruder at 320°C. The materials were injected molded into plaques 11.4 cm square by 3.2 mm and 1.6 mm thick. Mechanical analysis consisting of tensile, pendulum impact, and ballistic impact testing was conducted using the plaques or samples machined from the plaques. Dynamic mechanical testing was performed using bars machined from the 3.2 mm plaques to observe the low temperature loss peak and determine the activation energy of the loss mechanism. Differential scanning calorimetry was conducted to measure the glass transition temperatures of the material. The presence of two glass transition temperatures and lack of transparency have indicated that the blends are immiscible. Both sets of blends exhibit fair to good ballistic impact properties up to at least 20% polysulfone or polyetherimide. The tensile data of the blends show an increase in tensile strength over the same range.

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PREFACE

The work described in this report was authorized under Work Unit number 1L161102AH5202M00 entitled "New Transparent Polymer for Ballistic Eye Protection," and covers the period October 1987 to September 1990.

The author would like to thank Dr. Stephen Petrie, Prof. Stephen Orroth, Jr., and Mr. William Blood of the Department of Plastics Engineering at the University of Lowell; Mr. Peter Dehmer and Mr. David Dunn of the United States Army Materials Technology Laboratory; and Mr. John Song and Mr. Phillip Cunniff of the Individual Protection Directorate, United States Army Natick Research, Development and Engineering Center for their assistance in this research. Thanks are again extended to Mr. John Song and to Mr. John Troung of the Individual Protection Directorate, United States Army Natick Research, Development and Engineering Center for their constructive comments while reviewing this report.

MECHANICAL PROPERTIES OF POLYCARBONATE-POLYSULFONE AND POLYCARBONATE-POLYETHERIMIDE BLENDS

INTRODUCTION

Currently, the material of choice for eye protection is polycarbonate. The impact properties of polycarbonate are outstanding; however, the scratch and chemical resistance of the material are poor. There is a need to create a material with good scratch and chemical resistance without sacrificing polycarbonate's superior ballistic impact behavior and optical clarity.

Recent work[1-3] has shown that polycarbonate blended with a polyarylate becomes miscible when processed at high temperature. It is suggested that the miscibility is a result of transesterfication reactions between the two polymers, forming block copolymers that function as compatibilizers. Other transparent engineering thermoplastics formed by condensation polymerization may undergo similar reactions with polycarbonate.

This work investigates the effect of blending polysulfone and polyetherimide with polycarbonate at high temperature in order to determine if polycarbonate behaves similarly to polyarylate with other amorphous engineering thermoplastics. This work, while attempting to produce a transparent material with improved scratch and chemical resistance, only examines the thermal and mechanical properties of the polymers.

Early work by Myers and Brittain[4] on polycarbonate-polysulfone blends has found that they were immiscible but the cohesion across phase boundaries and the variations in properties with composition indicated a degree of interaction. However, Myers and Brittain's use of solution blending and film casting encouraged phase separation. Samples they produced for testing may not have had sufficient mixing. The work of this study in contrast with Myers and Brittain's employs blending and processing methods to insure a high level of mixing.

MATERIALS

The polycarbonates (PC) and the polyetherimide (PEI) were obtained from the General Electric Company. The polycarbonates were natural transparent general purpose Lexan® 121, 141, and 161 with reported melt flow rates of 16.5, 9.5, and 8.0 g/10 min., respectively. The polyetherimide was natural transparent Ultem® 1000.

The polysulfone (PSF) was obtained from Amoco Performance Products, Inc. and was the natural transparent general purpose grade Udel® P-1700.

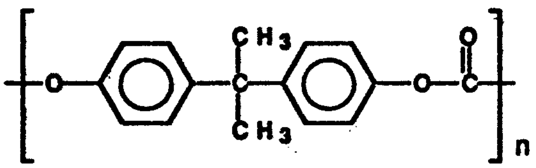


Figure 1: The Structure of Polycarbonate.

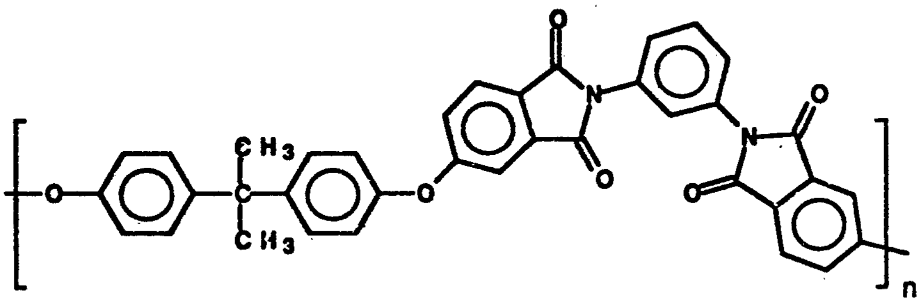


Figure 2: The Structure of Polyetherimide.

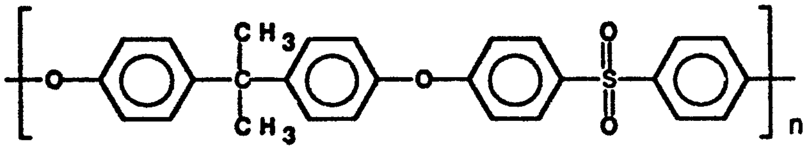


Figure 3: The Structure of Polysulfone.

EXPERIMENTAL

Material Processing

Thin Films. A small amount of each polymer was dried for a minimum of 16 hours at 125°C. Thin (<0.25 mm) polycarbonate films were produced with a heating press at 230°C. Polyetherimide and polysulfone films were pressed at 270°C. These films were used for dynamic mechanical analysis and thermal analysis.

Injection Molding of Plaques. Ten kilograms of polymer were dried for a minimum of 16 hours at 125°C. Plaques 3.2 and 1.6 mm thick were injection molded with a Van Dorn 200, a 200-ton clamping force injection molding machine. The 3.2 mm thick plaques were 11.43 cm square, fan--gated, with a 20 cm flow length. The 1.6 mm thick plaques were 11.5 cm long and 10.8 cm wide, fan--gated, with a 28 cm flow length. Molding was switched between the cavities by use of a gate valve. The maximum processing temperature for the polycarbonates and polycarbonate blends was 330°C. For neat polysulfone, 360°C was the maximum processing temperature for the 3.2 mm thick plaques

plaques and 400°C for the 1.6 mm thick plaque. The maximum processing temperature for neat polyetherimide was 390°C. The mold temperature was set at 5° C below the glass transition temperature (Tg) of the polymer being molded. The polyetherimide was not able to be molded in the 1.6 mm thick cavity. Even under maximum injection pressure and barrel temperature conditions, the result was a short shot due to the long flow length and the thinness of the plaque.

Polymer Blending. After the mechanical properties analysis and observations of the appearance of the molded samples, Lexan 161 was chosen to be blended with the polysulfone and polyetherimide. Portions of polysulfone and polyetherimide were weighed out to produce 5, 10, and 20% by weight blends when each was mixed with 22.7 kg of polycarbonate. The 22.7 kg of polycarbonate pellets were placed into aluminum trays. The polysulfone or polyetherimide pellets for one-blend composition were added to the polycarbonate and mixed by hand until a visually uniform distribution was obtained. The mixture was dried overnight at 125°C and then transferred to a drying hopper at 121°C. A Leistritz Laboratory Extruder LSM 3034 counter-rotating twin screw extruder set with a low shear profile was used to blend the polymers. The material was extruded at 320°C with a 105 rpm screw rotation. A vacuum pump was connected at a vent zone of the extruder as a precaution to pull off any additional moisture and low molecular weight material. The polymer was extruded into a water bath and then fed into a granulator where it was chopped into pellets and collected. The material processing rate was between 7-10 lb/h. In addition to the 5%, 10%, and 20% polycarbonate-polysulfone and polycarbonate-polyetherimide blends, a blank of virgin polycarbonate was extruded to determine the effect of additional processing on the material.

Thin films and molded plaques of the blends were produced in a similar manner as stated for the neat resins.

Material Characterization

Thermal Characterization. The thermal characterization was performed on a DuPont 1090 thermal analyzer with a 910 differential scanning calorimeter (DSC) cellbase. Each material was cycled three times through its heating profile to eliminate the effects of its previous thermal history and to check for reproducibility. A scan rate of 5°C/min was used while purging with dry nitrogen at 40 mL/min. The polycarbonate was scanned from 30°C to 200°C, while the polysulfone, polyetherimide, and the blends were scanned to 300°C.

Dynamic Mechanical Analysis. Bars approximately 1.25 cm wide were machined from the 3.2 mm plaques. These bars were then tested on a DuPont 983 Dynamic Mechanical Analyzer controlled by a 2000 Thermal Analyzer from -150°C to the glass transition temperature of the sample. A multiplexing-thermal step data acquisition program was used to analyze the samples. The frequencies used were 0.33, 1.0, 3.3, and 10 Hz with a temperature step interval of 2.5°C. Cooling of the sample and temperature control was accomplished by a DuPont LNCA-II liquid nitrogen cooling accessory.

Ballistic Testing. The ballistic testing was performed according to MIL–STD–662E V_{50} Ballistic Test for Armor[5] using a high-pressure helium gas gun. A 17–grain fragment simulator was used as the projectile. The test panels were rigidly held in the sample holder. A 0.05 mm thick aluminum witness plate was used to record complete penetrations. Four light screens were used as triggers to record the time–of–flight of the projectile before and after impact.

Two different velocities, V_{50} and V_c, were calculated. V_{50}, the velocity at which 50% of the impacts result in complete penetration, was calculated from the arithmetic mean of the five highest partial and five lowest complete penetration impact velocities. V_c, the critical velocity for complete penetration, was calculated by fitting the following equations[6]

$$V_r^2 = A V_s^2 - B \tag{1}$$

$$V_c^2 = \frac{B}{A} \tag{2}$$

$$V_r = (A(V_s^2 - V_c^2))^{1/2} \tag{3}$$

where V_s = the striking velocity of the projectile
V_r = the residual velocity after penetration
V_c = the critical velocity for complete penetration
A = the slope of the line
B = the intercept

to a plot of all striking velocities greater than and equal to the lowest complete penetration velocity versus the residual velocities. A minimum of 32 shots was used for each set of samples, with at least eight shots spread over the range from V_{50} to approximately 120 m/s above the V_{50}.

Impact Testing. Izod and Charpy bars were cut from the 3.2 mm thick molded plaques and separated into two sets: bars cut either transverse or longitudinal to the flow direction. Each sample set contained between 13 and 18 samples. The bars were machined to size, notched, and tested with a TMI pendulum impact machine according to ASTM D256-84 Standard Test Methods for Impact Resistance of Plastics and Electrical Insulating Materials[7] with the exception that the Charpy bars were only 11.4 cm long instead of the standard minimum of 12.45 cm due to the dimensions of the mold cavity.

Tensile Testing. Microtensile bars type V were cut from the molded plaques and separated into two sets: bars cut either transverse or longitudinal to the flow direction. The bars were machined to size and tested at 0.05 and 5.0 in/min strain rate with an Instron mechanical testing machine in accordance with ASTM D638-87 Standard Test Methods for Tensile Properties of Plastics[8]. Each sample set contained 10 samples, 5 for each rate of testing. Yield and break load, stress, elongation, and energy were recorded.

Gel Permeation Chromatography. Gel permeation chromatography (GPC) was used to determine if the number-, weight-, and z-average molecular weight of the polycarbonate and polycarbonate-polysulfone blends changed with processing. The GPC instrument consisted of Waters Associates M6000A pump, U6K injector, M440 fixed wavelength detector set at 254 nm; all controlled by a 480 data station. The columns were of a mixed-bed type, 50 cm long, supplied by Jordi Associates. The carrier solvent was tetrahydrofuran with a flow rate of 1.1 mL/min. The sample concentration was 0.10 % (w/v) and an injection volume of 80 μL. Twenty-one polystyrene standards ranging from 1,250 to 4,000,000 molecular weight were used to calibrate the column. A third order polynomial was found to fit the calibration curve with a correlation coefficient of 0.99958. The polyetherimide and its blends could not be analyzed on this system as it was set up since polyetherimide is insoluble in tetrahydrofuran.

RESULTS AND DISCUSSION

Material Processing

Material Appearance. The 3.2 mm thick molded polycarbonate plaques appeared as normal polycarbonate possessing the characteristic bluish coloring when seen from the side. The 1.6 mm plaques were very different in coloring from one another. The Lexan® 121 lacked the bluish coloring but otherwise appeared normal. The Lexan® 141 had a pale orange tint that was very noticeable when viewed from the side. The Lexan® 161 also lacked the bluish coloring and appeared grayish when viewed from the side. These color changes are important because they reflect chemical changes in the polymer system; however, since absorption constants for colored material can be quite large, only an extremely small fraction of the material may have changed.

Causes for the differences in coloration between the plaques of different thickness as well as different materials may be traced to barrel temperature and injection or "boost" pressure. The 1.6 mm thick Lexan® 121 plaques were processed at the same temperature as the 3.2 mm plaques but at a much higher injection pressure. The 1.6 mm thick Lexan® 141 and 161 plaques were molded at both a higher temperature and injection pressure than the 3.2 mm plaques in order to fill the thinner cavity. The shot size for the thinner plaques was 30 percent smaller than for the thicker plaques. This difference results in the material experiencing a 30 percent longer residence time in the barrel of the injection molder at a higher temperature. The higher injection pressure and thinner section cause greater shear forces in the material. These conditions could lead to material degradation resulting in discoloration and/or reduction in material properties.

The polysulfone and polyetherimide plaques appeared normal with no discoloration or any other visible defects.

The polycarbonate-polysulfone blends were white in color ranging from translucent to opaque. The polycarbonate-polyetherimide blends were opaque and ranged from beige to light tan in color. The plaques molded from the extruded polycarbonate blank were light amber in color, greatly resembling natural polysulfone.

Material Characterization

Thermal Characterization. Table 1 summarizes the results of the DSC characterization of the neat polymers and blends. The presence of two Tg's in the blends, with little or no shifting from where they would normally occur in each of the neat resins, indicates that the polycarbonate-polysulfone and polycarbonate-polyetherimide blends are immiscible.

Table 1. Differential Scanning Calorimetry Results

Material	First Tg (°C)	Second Tg (°C)
Lexan® 121	146	n/a
Lexan® 141	147	n/a
Lexan® 161	147	n/a
Processed Lexan® 161	146	n/a
Polysulfone (PSF)	186	n/a
Polyetherimide (PEI)	216	n/a
5% PSF blend	146	187
10% PSF blend	147	184
20% PSF blend	145	186
5% PEI blend	146	216
10% PEI blend	148	216
20% PEI blend	145	212

Gel Permeation Chromatography. The molecular weights of the polycarbonates, polysulfone, and their blends were determined from pellets of virgin and processed polymer and shavings from the 3.2 and 1.6 mm molded plaques. The results in table 2 show that for the polycarbonate and polysulfone samples, there was no significant change in the molecular weight of the material with processing. The results from the blended polymers are somewhat curious. The 5% polysulfone blend has higher *Mn*, *Mw*, and *Mz* for each corresponding sample of the 20% polysulfone blend. This fact is difficult to explain since polysulfone has a higher molecular weight than polycarbonate and there is more polysulfone in the 20% blend. The 10% polysulfone samples have molecular weights above, between, and below the 5 and 20% samples and, with one exception, between the molecular weights of the polycarbonate and polysulfone.

Since the molecular weight of the polymers appears to be generally stable, the possibility of reactions between the different polymers seems unlikely as these should affect the molecular weight distribution of the polymers.

Table 2. Gel Permeation Chromatography Results

Sample	*Mn* (g/mol)	*Mw* (g/mol)	*Mz* (g/mol)
Lexan® 121, pellets	15,400	38,600	58,700
Lexan® 121, 1.6 mm plaque	15,500	38,800	59,000
Lexan® 141, pellets	16,400	42,800	66,300
Lexan® 141, 1.6 mm plaque	16,800	42,400	65,500
Lexan® 141, 3.2 mm plaque	15,600	41,400	64,500
Lexan® 161, pellets	16,800	43,200	66,100
Lexan® 161, 1.6 mm plaque	17,300	43,400	66,600
Lexan® 161, 3.2 mm plaque	16,700	42,800	65,500
Processed Lexan® 161, pellets	16,500	42,900	66,400
Processed Lexan® 161, 1.6 mm plaque	16,700	42,200	65,800
Processed Lexan® 161, 3.2 mm plaque	16,100	42,100	65,700
Polysulfone, pellets	19,300	47,900	74,000
Polysulfone, 1.6 mm plaque	18,000	45,300	73,200
Polysulfone, 3.2 mm plaque	18,900	46,700	72,700
5% Polysulfone-PC, pellets	19,200	46,000	71,400
5% Polysulfone-PC, 1.6 mm plaque	18,700	43,400	67,500
5% Polysulfone-PC, 3.2 mm plaque	18,500	43,900	68,100
10% Polysulfone-PC, pellets	18,800	44,300	68,400
10% Polysulfone-PC, 1.6 mm plaque	17,800	42,400	65,700
10% Polysulfone-PC, 3.2 mm plaque	18,800	44,400	68,700
20% Polysulfone-PC, pellets	18,700	44,600	69,400
20% Polysulfone-PC, 1.6 mm plaque	18,400	43,100	66,700
20% Polysulfone-PC, 3.2 mm plaque	17,900	42,700	66,200

Dynamic Mechanical Analysis. A polynomial curve-fitting routine with up to 10 parameters was used to characterize the low temperature loss peak of the polymers. A typical plot of the data with the fitted curve is shown in figure 4. The calculated curve was then used to determine the temperature of the peak maximum and peak area. Activation energies for the polymers were determined from a plot of ln(frequency) versus the reciprocal absolute temperature of the peak maximum. The slope of this line, multiplied by the gas constant R, yielded the activation energy. The peak area was determined by integrating the fitted curve and subtracting the area below a linear baseline connecting the upper and lower limits of the integral. It was then normalized by dividing by the range of the upper and lower limits to produce a dimensionless value. A listing of the peak maxima and activation energies from the 983 DMA data are given in table 3.

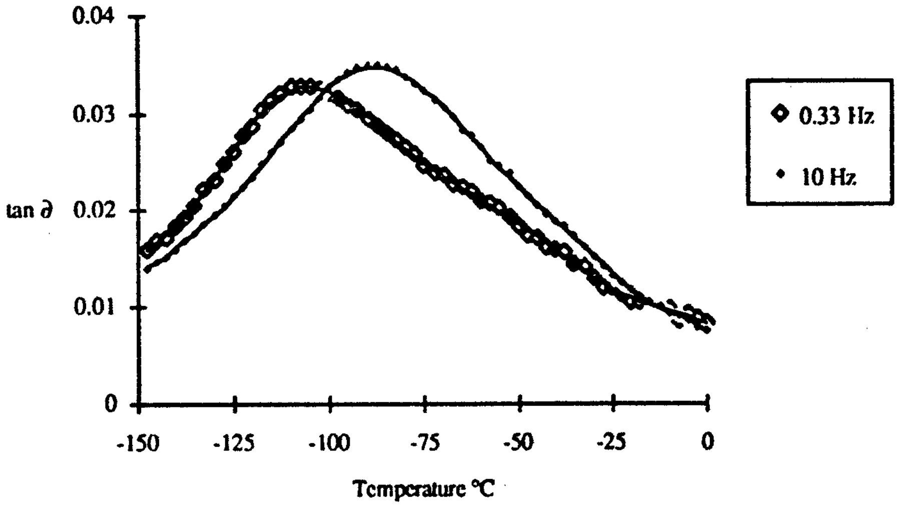


Figure 4: Tan δ versus Temperature for processed polycarbonate at 0.33 and 10 Hz.

Table 3. Dynamic Mechanical Analysis Results

Material	Peak Temperatures (°C) 0.33 Hz	1.0 Hz	3.3 Hz	10. Hz	Activation Energy (kJ/mol)
Lexan® 121	-102.0	-97.0	-90.9	-84.7	53
Lexan® 141	-105.1	-100.0	-93.5	-86.9	49
Lexan® 161	-104.0	-98.9	-92.5	-86.3	50
Processed Lexan® 161	-104.4	-99.1	-92.9	-86.9	51
Polysulfone	-112.5	-107.4	-101.7	-95.7	48
Polyetherimide	-114.1	-108.7	-102.0	-95.8	43
5% PSF blend	-105.9	-101.4	-94.6	-88.1	49
10% PSF blend	-105.7	-101.2	-94.5	-88.5	50
20% PSF blend	-106.7	-101.8	-95.3	-89.0	49
5% PEI blend	-105.2	-100.4	-94.1	-87.6	50
10% PEI blend	-106.6	-101.4	-94.8	-88.2	47
20% PEI blend	-106.2	-101.2	-94.7	-88.2	48

The activation energy for the neat polymers and the blends is basically the same within error except for polyetherimide, which is lower than the rest. The bisphenol A repeat unit is the major common feature between the polymers and since it is the major energy absorbing feature in the polymers, it is not surprising that the activation energy is similar.

Upon examining the data from the peak area calculation, it was found that a plot of the peak area versus ln(frequency), shown in figures 5 and 6, yielded a linear fit with a correlation coefficient of 0.969 or better for all materials. The slope of these lines ranged from 5.4×10^{-4} to 8.2×10^{-4} for the polycarbonates and the blends, 4.7×10^{-4} for polysulfone, and -1.5×10^{-4} for polyetherimide. The magnitude of the peak area is a reflection of the material. The polycarbonates with the higher melt viscosity had larger peak areas. For the polysulfone blends as seen in figure 5, the magnitude of the peak area is at the same level as for the neat polysulfone but the slope of the lines is equivalent to the polycarbonate line.

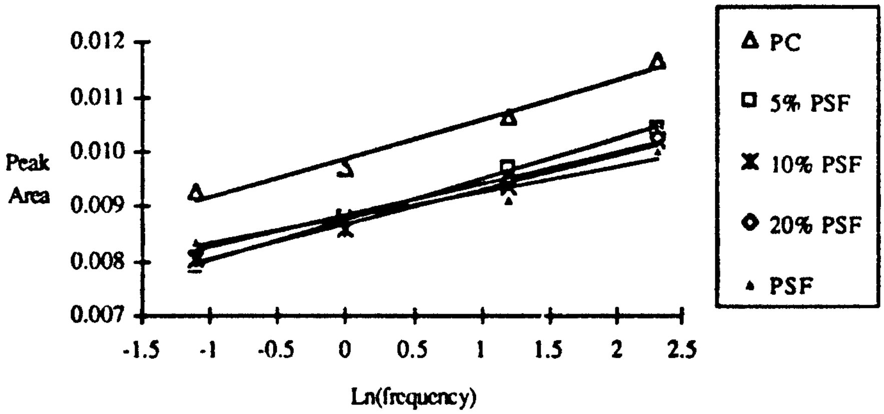


Figure 5: Low Temperature Loss Peak Area Versus Ln Frequency of Polysulfone Blends.

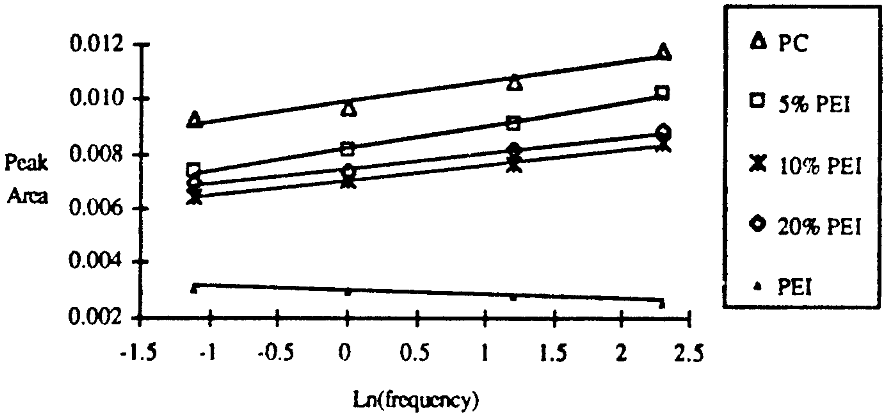


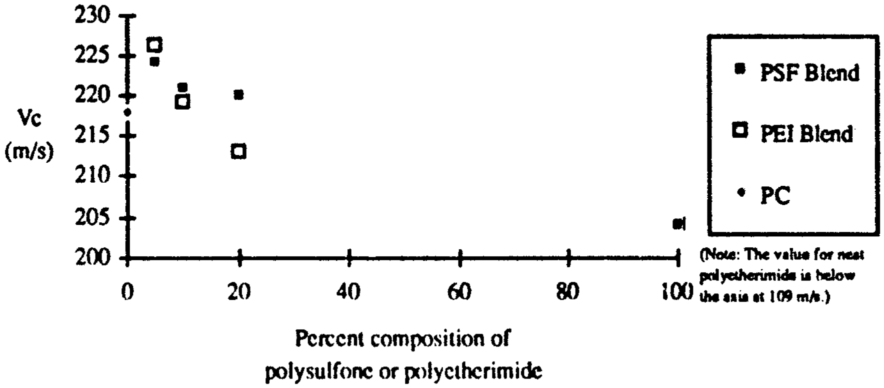
Figure 6: Low Temperature Loss Peak Area Versus Ln Frequency of Polyetherimide Blends.

Figure 6 shows a similar behavior for the polyetherimide blends with a drop in the magnitude of the peak area from the neat polycarbonate to a level slightly lower than the polysulfone blends. The slope of the lines for the polyetherimide blends are also equivalent to the polycarbonate line.

Ballistic Testing. Table 4 lists the calculated V_{50} and V_c for the 3.2 and 1.6 mm thick plaques of the neat and blended polymers. Graphs of the material composition versus V_c are shown in figures 7 and 8.

Table 4. Ballistic Testing Results

Material	V_{50} (3.2 mm) (m/s)	V_c (3.2 mm) (m/s)	V_{50} (1.6 mm) (m/s)	V_c (1.6 mm) (m/s)
Lexan® 121	230	231	141	137
Lexan® 141	221	221	133	136
Lexan® 161	218	221	130	136
Processed Lexan® 161	215	218	139	136
Polysulfone	202	204	132	137
Polyetherimide	106	109	na	na
5% PSF blend	225	224	137	136
10% PSF blend	220	221	136	134
20% PSF blend	219	220	133	135
5% PEI blend	225	226	138	139
10% PEI blend	217	219	134	134
20% PEI blend	212	213	123	121



(Note: The value for neat polyetherimide is below the axis at 109 m/s.)

Figure 7: Ballistic Test Results of 3.2 mm Plaques.

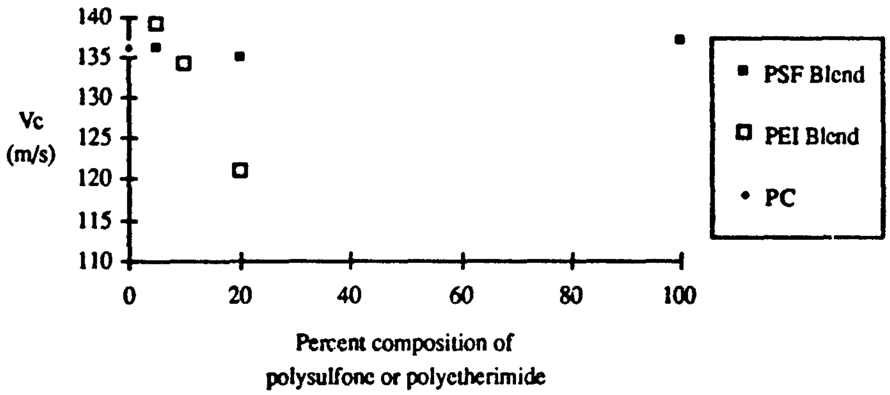


Figure 8: Ballistic Test Results of 1.6 mm Plaques.

In addition to the V_c and V_{50}, the failure mechanism is an important consideration. The desired failure mechanisms are punching or petalling, which are of a ductile nature. Spalling, whether it is delamination of a rear section or a blowout of a section larger than the impact area, is undesirable. Sharp edges and high velocities of spall can cause damage to body tissues. The MIL-STD-662E V_{50} Ballistic Test for Armor[5] requires the use of a 0.05 mm thick aluminum witness plate to record complete penetrations. The projectile may actually be stopped by the material but if a fragment or spall punctures the witness plate, the penetration is considered complete.

The Lexan® 121 was prone to brittle failure resulting in a large amount of spalling and cracking. The Lexan® 141 and 161 by contrast had only one brittle failure and some minor cracking that could be related to previous tests or visible defects located near the point of impact. This difference in the failure mechanism leads to the selection of Lexan® 161 to blend with the polysulfone and polyetherimide even though the V_c for the Lexan® 121 was 10 m/s greater for the 3.2 mm plaques. The difference in V_c for the 1.6 mm plaques of the polycarbonates is insignificant. The Lexan® 161 was chosen over the Lexan® 141 because there was concern at that time that the repeated processing of the polymer by blending and then molding could reduce the molecular weight and lead to a reduction in physical properties. Since there was no difference in the V_c and insignificant difference in the V_{50} between the materials, it was believed that it would be best to use the higher molecular weight material.

The testing of the polysulfone plaques resulted in a significant number of impacts causing spalling or cracking in the samples. With the polyetherimide samples, every impact caused radial cracking and spalling.

In figure 7, it can clearly been seen that in the 3.2 mm thick samples, that an initial increase in V_c occurs with 5% loading with either polysulfone or polyetherimide. This initial increase is then followed by what appears to be a linear decrease in V_c with increasing loading of polysulfone or polyetherimide. The same results are found in the 1.6 mm thick samples for polyetherimide. The values of the V_c for 1.6 mm thick polycarbonate and polysulfone are almost the same and the V_c of the polysulfone blends are scattered about those values.

The 3.2 mm thick processed polycarbonate plaques had only one impact that caused cracking and the 1.6 mm thick plaques had a few impacts that appeared to tear the material rather than to crack it.

The polycarbonate-polysulfone blends showed general increasing brittleness, with increasing concentration of polysulfone. The 5% blend behaved in a similar manner to the processed polycarbonate. The 10% blend had increased cracking and an incident of spalling in each thickness. The 3.2 mm thick plaques of the 20% polycarbonate-polysulfone showed about the same amount of cracking as the 10% blend, but the cracking appeared to be less severe and there were no cases of spalling. The 1.6 mm thick plaques actually went against the trend with no brittle failures.

The polycarbonate-polyetherimide blends showed increasing brittleness which resulted in additional cracking and spalling with increasing polyetherimide concentration .

A comparison between V_c and V_{50} for each sample is shown in figure 9. For the 3.2 mm thick samples the calculated V_c is slightly higher in value than the V_{50} but in the 1.6 mm thick plaques, the plot of the velocities are scattered about a $V_c = V_{50}$ line.

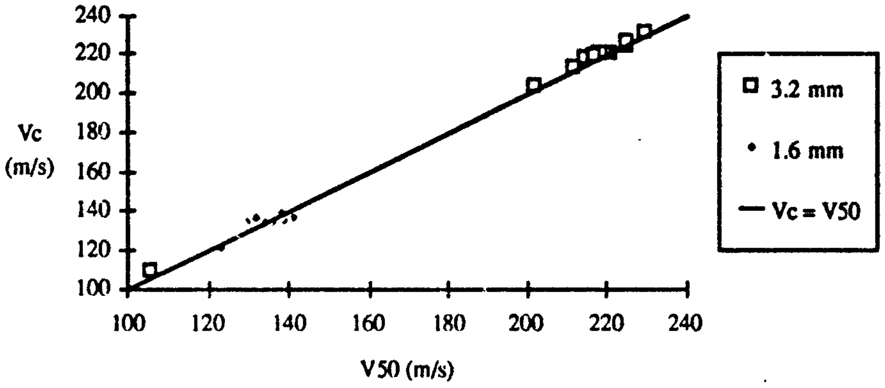


Figure 9: V_c Versus V_{50} for Polycarbonate-Polysulfone and Polycarbonate-Polyetherimide Blends.

Impact Testing. The results from both the Izod and Charpy testing for the neat polymers show that there is no significant difference in impact strength between samples cut longitudinal or transverse to the flow direction or with different impact test methods. As seen in figure 10, the polysulfone blend samples show a change in impact strength at 10% loading. Half of the sample sets at 10% polysulfone reflect a lower impact strength and a brittle failure mechanism.

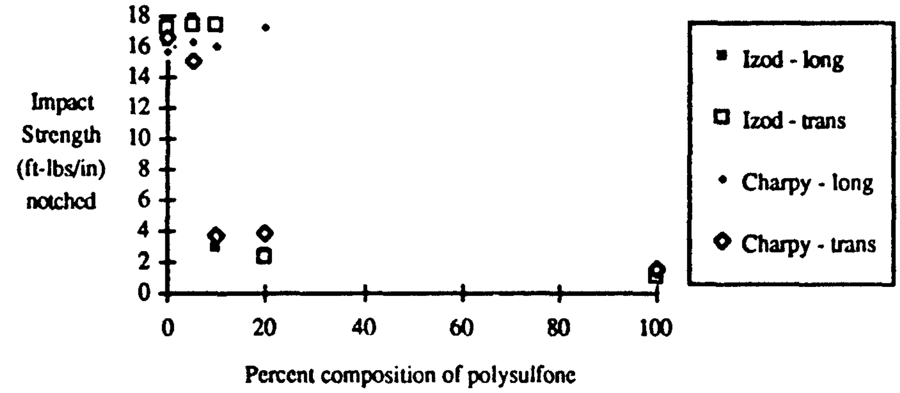


Figure 10: Impact Strength Versus Percent Composition of Polysulfone Blends.

At 20% polysulfone, only the Charpy sample set cut longitudinally to the flow direction still maintains high impact strength. Figure 11 shows that the polyetherimide blends show a large drop in the impact strength with polyetherimide levels of 20%.

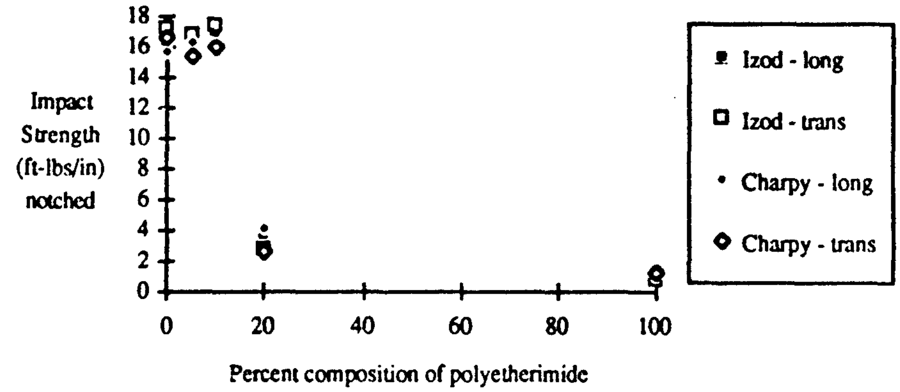


Figure 11: Impact Strength Versus Percent Composition of Polyetherimide Blends.

The information shown in figures 10 and 11 can be misleading. The blended materials can fracture by two different types of failure mechanisms; ductile and brittle failure. A bimodal distribution occurs with impact strengths of approximately 3.5 ft-lb/in for brittle fracture and 16 ft-lb/in for ductile fracture. According to the ASTM Standard D256-84 if a sample set experience different types of failure or failure mechanisms, then they shall be considered a departure from standard and shall be grouped by type of failure and the average reported for each group along with the percent of samples failing by that manner. It becomes confusing to graphically display in one chart the entire result for the impact samples while showing impact strength, percent composition, test type, and flow direction in the sample. Figures 10 and 11 show only the average for the majority result. The percentage of ductile failure versus percent composition is shown in figures 12 and 13.

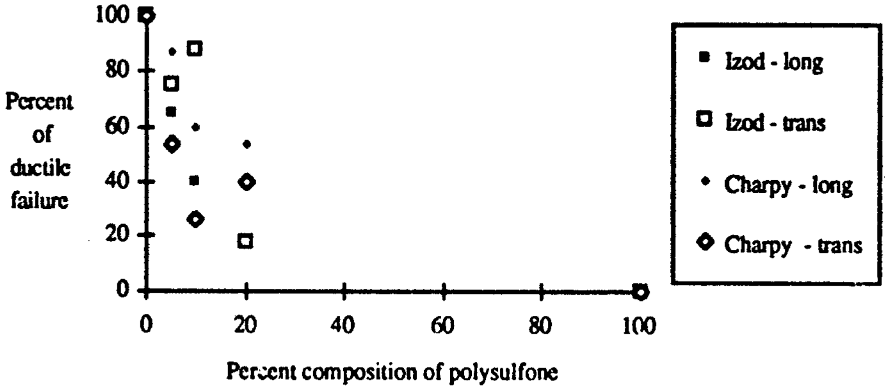


Figure 12: Percent of Ductile Failure Versus Percent Composition of Polysulfone Blends.

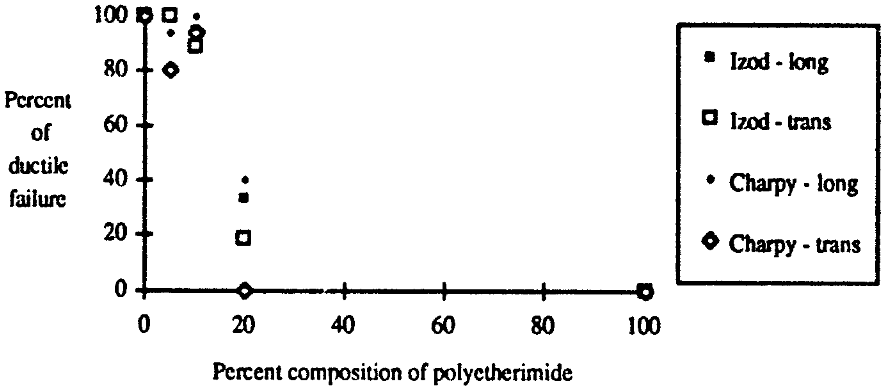


Figure 13: Percent of Ductile Failure Versus Percent Composition of Polyetherimide Blends.

These graphs reflect with greater clarity the changes in impact behavior with varying composition. The polysulfone blend shows a significant decrease in ductile failures with five percent polysulfone. Increasing polysulfone concentration causes even fewer ductile failures, with the plot in figure 12 resembling an exponential decay curve. The behavior of the polyetherimide blends shown in figure 13 reveals a high level of ductile failure, up to

10% polyetherimide. At 20% polyetherimide, there is a sharp drop in the number of ductile failures.

Tensile Testing. The results of tensile testing are shown in figures 14 through 21. In analysis of the yield strength for the polysulfone blends, it appears that the difference in thickness is significant while the difference in orientation is essentially insignificant within a blend composition. The thicker samples had a consistently higher yield strength for the 0.05 in/min strain rate; however, this relationship does not hold completely true for the 5.0 in/min strain rate samples. For the yield strength of the polyetherimide blends, the difference in thickness and orientation is insignificant within a blend composition.

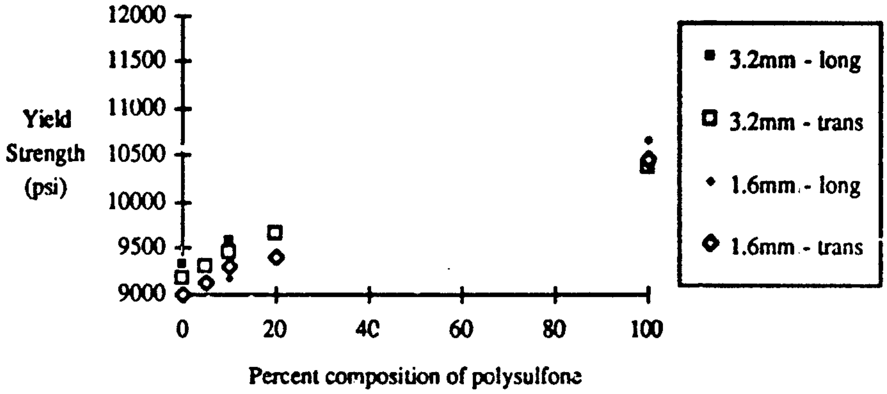


Figure 14: Yield Strength Versus Percent Composition of Polysulfone Blends at 0.05 in/min Strain Rate.

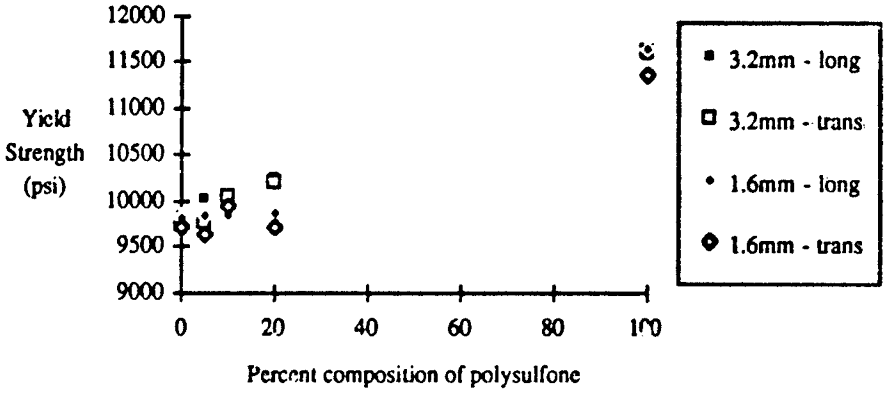


Figure 15: Yield Strength Versus Percent Composition of Polysulfone Blends at 5.0 in/min Strain Rate.

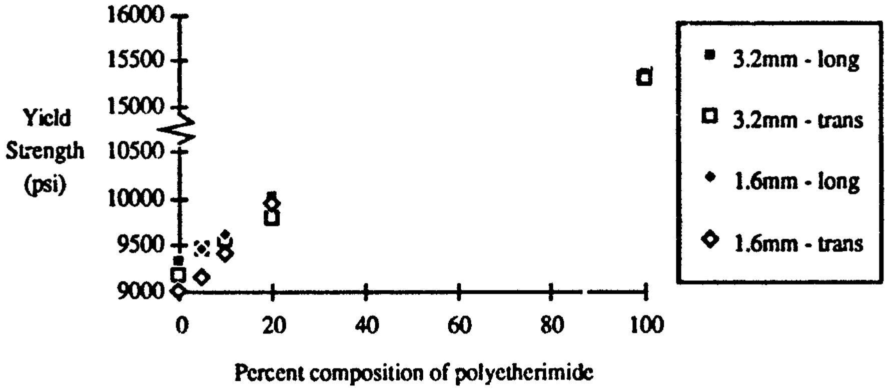


Figure 16: Yield Strength Versus Percent Composition of Polyetherimide Blends at 0.05 in/min Strain Rate.

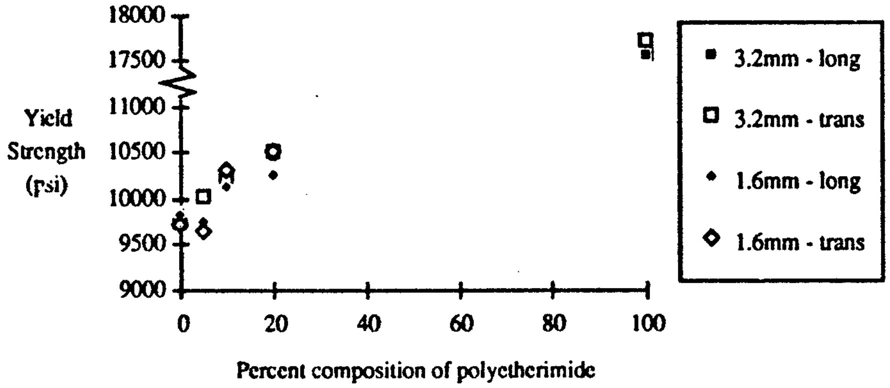


Figure 17: Yield Strength Versus Percent Composition of Polyetherimide Blends at 5.0 in/min Strain Rate.

The rate of testing is significant for both the polysulfone and polyetherimide blends. The average values of the yield strength for the 5.0 in/min strain rate were consistently 500 to 600 psi greater than for the 0.05 in/min strain rate.

Taking the results for the yield strength versus percent composition for a given strain rate as a whole, it is clear that the yield strength increases with increasing composition of polysulfone or polyetherimide if an average value is used for each composition. This increase appears to be linear but an extrapolation to 100 percent polysulfone or polyetherimide underestimates the measured value except for the polysulfone blends at 0.05 in/min strain rate, which overestimates the measured value. This indicates that a line describing the yield strength versus percent composition across the entire compositional range would either be curved or discontinuous. The amount of change in yield strength for a given change in composition is much greater in the polyetherimide blends, which is not surprising since neat polyetherimide has a much greater yield strength than neat polysulfone.

The yield energy, or the amount of energy input in a sample at the yield point, is the energy required to plastically deform a sample. Figures 18 to 21 show the yield energy versus composition for the polysulfone and polyetherimide blends. It can be noted that thicker samples have consistently higher values; longitudinal samples generally had higher values than transverse samples at 0.05 in/min strain rate; and rate of testing had no effect. The yield energy appears to be constant across the composition of the blends to 20% polysulfone or polyetherimide. Since the yield energy for the neat polysulfone and polyetherimide are greater than for the blends, it can be assumed that some change occurs at a higher composition to raise the yield energy to the higher level. These data have an average standard deviation of ±2.2 ft-lb/in^2, which is quite large compared to the spread of the average values of the sample sets. While analysis of the data has been made, it should be noted that no true significance can be given to effects of thickness, orientation, rate, material, or composition.

The break tensile strength and energy were also determined but had such a high variance that no meaningful observations could be made. The failure mechanism for break requires a crack to initiate then propagate through the sample. The initiation of a crack requires a large amount of strain energy and is highly variable. If a preexisting defect is present, then the sample will fail at a lower load, elongation, and energy because a crack did not have to initiate. These factors lead to the high variance in the break data.

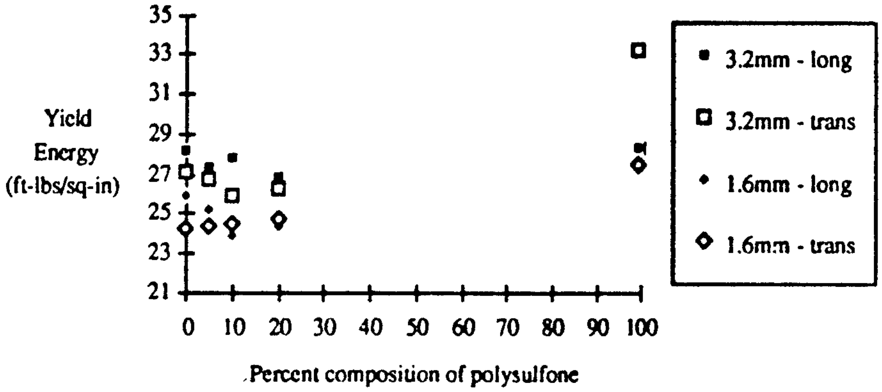


Figure 18: Yield Energy Versus Percent Composition of Polysulfone Blends at 0.05 in/min Strain Rate.

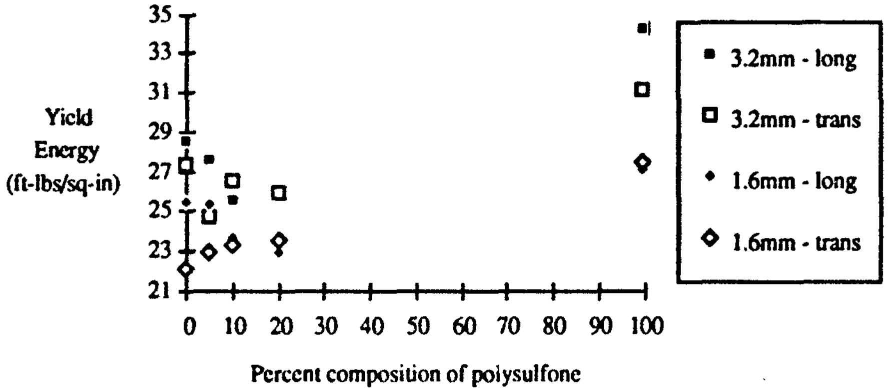


Figure 19: Yield Energy Versus Percent Composition of Polysulfone Blends at 5.0 in/min Strain Rate.

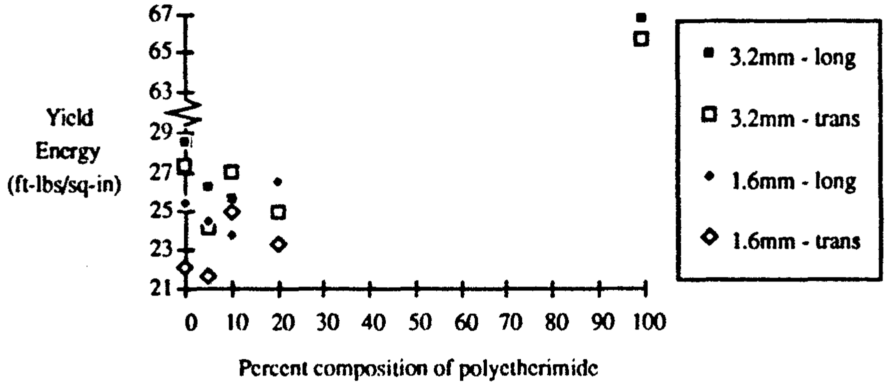


Figure 20: Yield Energy Versus Percent Composition of Polyetherimide Blends at 0.05 in/min Strain Rate.

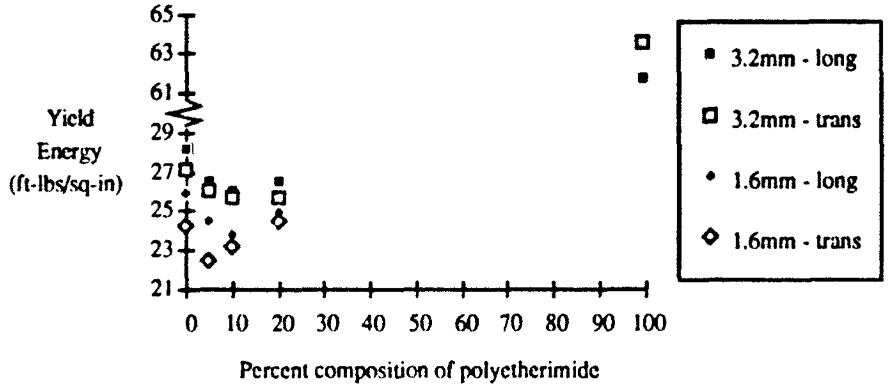


Figure 21: Yield Energy Versus Percent Composition of Polyetherimide Blends at 5.0 in/min Strain Rate.

CONCLUSIONS

It is obvious from the DSC results and visible appearance that the blends of polycarbonate with polysulfone and polyetherimide are immiscible and results from the molecular weight determination give no evidence that any intermolecular reactions are occurring. The polysulfone and polyetherimide are too stable to undergo any chain-scission reaction at temperatures which are low enough to prevent the degradation of the polycarbonate.

The effect of the repeated processing of the polymers did not significantly affect the molecular weight of the polymers but did affect the appearance of the polymers by a discoloration of the polycarbonate.

From the DMA data analysis, a linear relationship exists between the low temperature loss peak area and the ln(frequency). Other (possibly linear) relationships exist between the percent composition and both the V_c and the yield strength.

The Izod and Charpy impact test data show a bimodal distribution of impact strength for the blends. The average impact strength and percent of ductile failures decrease with increasing composition of polysulfone and polyetherimide.

The lack of transparency precludes the use of these materials for eye protection; however, both sets of blends exhibit good to fair ballistic impact properties up to at least 20% polysulfone or polyetherimide and the tensile data of the blends actually show an improvement in tensile strength.

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