Enhancing Plastic Yielding in Polyestercarbonate Glasses by 1,4-Cyclohexylene Linkage Addition

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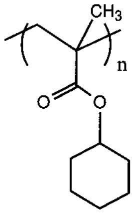
ABSTRACT: Two series of polyestercarbonates containing 1,4-cyclohexylene linkages were designed and synthesized to allow investigation of the relationship between local molecular motion and mechanical properties in these materials. Incorporation of increasing amounts of 1,4-cyclohexylene dicarboxylate linkages enhanced the segmental mobility in tetramethyl Bisphenol A polyestercarbonates and in Bisphenol A polyester carbonates. This effect is thought to be due to the cooperative conformational transition of neighboring cyclohexylene rings. The cooperative conformational transition is thought to be able to activate motions of the linked carbonate segments and further facilitate the molecular motions of neighboring segments. This motion is thought to favor shear yielding. The results are found to be consistent with this notion.

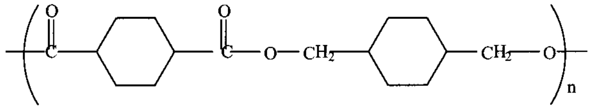
Introduction

The secondary relaxation behavior of polymers has been extensively studied over the past few decades since such relaxations have been generally correlated to macroscopic mechanical properties.[1−5] However, it is very difficult to identify unambiguously the molecular origin of the secondary relaxation in many polymers. This is because dynamic mechanical techniques, although able to directly access the mechanical relaxations, are unable to specify motions at the molecular level. Other techniques, such as dielectric relaxation and NMR, have been used to study movements of certain small molecular groups in solid polymers, and these techniques have been more successful at probing motion on the molecular level. However, the motions involving the specific groups studied using these techniques are not necessarily the same as or even related to the secondary relaxations observed mechanically. An example is the π-flip motion of phenylene ring observed by solid-state NMR in Bisphenol A polycarbonate.[2,6,7] In a recent study, it was found that the π-flip motion merely indicates chain motion but does not cause it.[8,9] Thus an effective way to ascertain the motional origins of the mechanical relaxations is to perform systematic DMA and NMR studies on polymers with tailored structures designed to elucidate such questions.

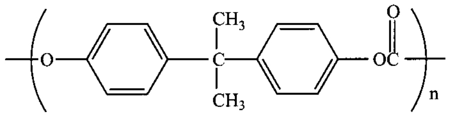
In recent years, Jho and Yee[10] have examined various molecular motions exhibited by individual groups in BPA−PC, and have concluded that these motions must be dynamically correlated to each other in order to induce the observed mechanical loss. Further, they concluded that the secondary relaxation in BPA−PC entails extensive cooperative motion, possibly involving as many as seven repeat units acting in concert. They argued that this type of cooperative motion produces a greater tendency for polymers to shear yield when under stress.[11] They further argued that the shear yield stress is lowered at a given temperature because large scale rearrangement is made easier by the intrinsic motional cooperativity of units.



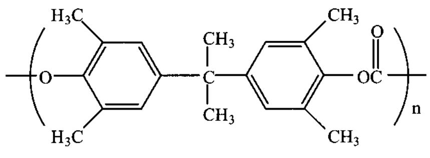
PCHMA [poly(cyclohexyl methacrylate)]



PCC [poly(1,4-cyclohexylenedimethylene 1,4-cyclohexylene dicarboxylate)]



BPA-PC (Bisphenol A polycarbonate)



TMBPA-PC (Tetramethylbisphenol A polycarbonate)

Figure 1. Molecular structures of PCHMA, PCC, BPA−PC, and TMBPA−PC.

The advantage of motional cooperativity is illustrated by the fact that not all polymers with low temperature secondary relaxations exhibit ductile behavior. Those polymers with secondary relaxations due to weakly correlated motions are found to be less likely to benefit from these relaxations. One example is poly(cyclohexyl methacrylate) (PCHMA) (Figure 1), which displays a sharp relaxation peak at −70 °C (DMS at 1 Hz). Heijboer[3] attributed this relaxation to the conformational transition of the cyclohexyl group, which involves a transition between the two most stable conformations of the cyclohexyl ring. The activation energy of this motion is similar to that of the isolated cyclohexane molecules, which suggests that the side groups are only weakly correlated with their surroundings. He found that even though the incorporation of the cyclohexyl

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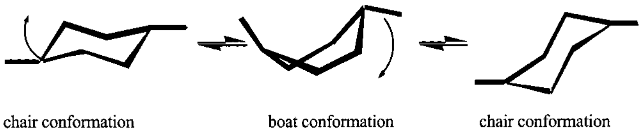


Figure 2. Conformational transition of the cyclohexylene ring.

group increased the strength of the secondary relaxation, there was not a corresponding increase in impact strength. Heijboer therefore concluded that only secondary relaxations due to the main chain are beneficial for impact resistance.[3,4] On the other hand, in a series of copolymers based on poly(ethylene terephthalate) (PET) and poly(1,4-cyclohexylenedimethylene terephthalate) (PCT), a systematic shift of the ductile−brittle transition to lower temperatures was observed in Izod impact tests as the cyclohexylene dimethylene group content in the copolymers was increased. A relaxation peak similar to that in PCHMA appears at about −70 °C (1 Hz). This relaxation has been identified with the conformational transition of the main chain cyclohexylene group by Chen et al.[12] (Figure 2). Moreover, they found that as the yield stress decreased, the craze stress increased with increasing cyclohexylene methylene content. They postulated that two neighboring cyclohexylene rings could cooperate in effecting a conformational transition, which would consequently lead to a higher overall level of chain mobility. This postulate was strongly supported by the NMR.[12]

The local molecular motions must couple with those on neighboring chains on some scale in order for such motions to influence macroscopic mechanical properties.[13] In this paper, we study how a particular local molecular motion, the conformational transition of the cyclohexylene ring, interacts with neighboring segments and further influences the mechanical properties exhibited by two series of polyestercarbonates. We further study the effect of changing the distance between the cyclohexylene rings and how this effects the mechanical properties.

Experimental Section

Chen et al.[12] showed that the conformational transition of cyclohexylene rings, which occurs at −70 °C (1 Hz), can enhance chain mobility in the PET/PCT copolymers. It will be very interesting and useful to see whether the local conformational transition of cyclohexylene units can couple with motions of other cyclohexylene units on the same chain and how it would influence the mechanical properties of these polymers. We designed two series of copolymers containing cyclohexylene groups. One of them contains Bisphenol A polycarbonate blocks. Bisphenol A polycarbonate is known for its prominent mechanical properties even at a very lower temperature, which has been correlated to its secondary relaxation occurring at −100 °C.[14] Recent study shows that the secondary relaxation of BPA−PC involves large scale cooperative motions.[11] Another contains tetramethyl Bisphenol A polycarbonate blocks, which are extremely brittle and have a secondary relaxation at 50 °C (1 Hz).[11] It is commonly thought the lack of mobility of TMBPA segments produces the brittleness. These segments are linked by cyclohexylene dicarboxylates. The block lengths were varied to study the effect on the ability of the cyclohexylene rings to couple to each other. We used DMA to measure the secondary relaxations of these copolymers, and compared the uniaxial tension behaviors between these copolymers under the influence of the conformational transition of cyclohexylene rings.

A. Synthesis. It would be optimum for testing properties of the copolymers if the block lengths are precisely controlled. However, it would take too much time and effort to synthesize large quantities of copolymers with precisely controlled block lengths for macroscopic mechanical tests. Moreover, the macroscopic behavior is probably not very sensitive to the regularity of the block length.[15] Consequently, the block copolymers were prepared using a relatively simple procedure, which included two steps. The first step was the synthesis of oligomers, and the second was the polymerization of the oligomers to form polymers with the desired structures.

1. Starting Materials. Phosgene was supplied as 20 wt % (1.9 M) solution in toluene from Fluka. *N*,*N*-Dimethyl-4-aminopyridine, triethylamine, thionyl chloride, 1,4-cyclohexylene dicarboxylic acid, *N*,*N*-dimethylaniline, and Bisphenol A (99+%) were supplied from the Aldrich Co. and were used without further purification. Tetramethyl Bisphenol A was kindly donated by the Bayer Co.

2. Synthesis of Monomers. (1) Synthesis of 2,2-Bis(4-chlorocarboxyphenyl)propane [BPA Bis(chloroformate)]. In a 2000 mL three-neck reaction flask with a mechanical stirrer was suspended 114.0 g (0.50 mol) of bisphenol A in 750 mL of a toluene solution containing 20 wt % phosgene (1.5 mol). The mixture was cooled to −40 °C using a dry ice/acetone bath. A solution of 121 g (1.0 mol) of *N*,*N*-dimethylaniline and 3.0 g (0.05 mol) of *N*,*N*-dimethyl-4-aminopyridine in 200 mL of toluene was added to the phosgene solution through a dropping funnel over 20 min. The reaction mixture became green in color as soon as DMA was added. The mixture was then allowed to gradually warm and kept for a further 12 h at room temperature. The mixture eventually turned light yellow, which indicated the completion of the reaction. The mixture was filtered to obtain a clear solution after the *N*,*N*-dimethylaniline hydrochloride salt was removed and washed with 200 mL of anhydrous toluene. Residual phosgene was removed under reduced pressure over a period of more than 2 h. The solution was then passed through a flash column, loaded with 600 mL of Davison 12 silica gel, with toluene as the eluant. The toluene was evaporated under vacuum and a light blue crystalline product was obtained. The crystals were purified by recrystallization from 400 mL of hexanes. This procedure produced 152 g of BPA bischloroformate with a final yield of 86%. The melt point was 93−94 °C. ^{1}H NMR (300 MHz, $CDCl_3$, ppm): 1.67 (s, 6H, *i*-PrCH_3); 7.13 (d, 4H, Ph*H*); 7.23 (d,4H, Ph*H*).

(2) Synthesis of 2,2-Bis(3,5-dimethyl-4-chlorocarboxyphenyl)propane [TMBPA Bis(chloroformate)]. The synthesis procedure for this compound is similar to that for BPA bischloroformate except that 2,2-bis(3,5-dimethylphenyl)propane was used as the starting material. The yield is 84%. The melting point was 106−108 °C. ^{1}H NMR (300 MHz, $CDCl_3$, ppm): 1.60 (s, 6H, *i*-PrCH_3); 2.20 (s, 12H, PhCH_3); 6.90 (s, 4H, Ph*H*).

(3) Synthesis of 1,4-Cyclohexylenedicarboxylic Chloride. In a 250 mL flask, 17.22 g (0.1 mol) of 1,4-cyclohexylene dicarboxylic acid was suspended in 59.49 mL (0.8 mol) of thionyl chloride. Two drops of *N*,*N*-dimethyl formamide were added to the solution, and the solution was refluxed for 3 h. Thionyl chloride was removed by distillation at 50 °C under vacuum. Needlelike white crystals formed after the solution was cooled. The crystals were recrystallized from *n*-hexane twice with a final yield of 80%. The melting point is 68−69 °C. ^{1}H NMR (300 MHz, $CDCl_3$, ppm): 2.77 (m, 2*H*, C*H*); 2.34 (m, 4H, CH_2); 1.60 (m, 4H, CH_2).

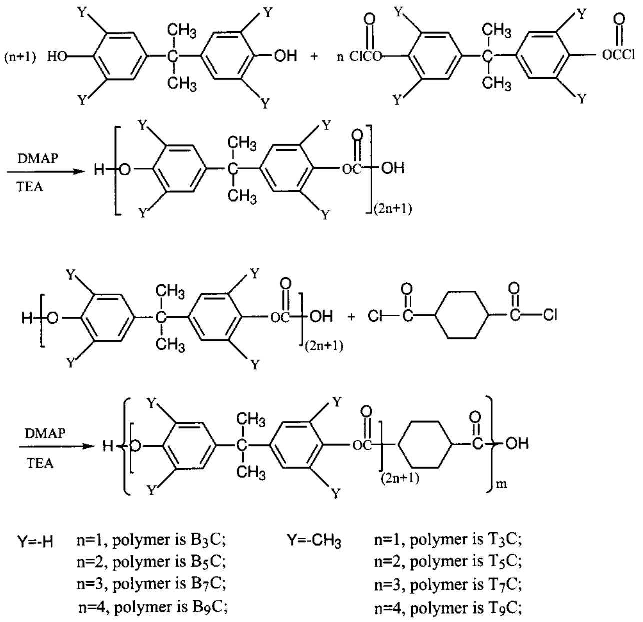
3. Synthesis of Polymers. (1) Tetramethyl Bisphenol A Polycarbonate. A 0.2031 g sample of cumylphenol (0.96 mmol), 14.20 g (50 mmol) of tetramethyl Bisphenol A, 12.63 g (125 mmol) of triethylamine, and 1.3 g of *N*,*N*-dimethyl-4-aminopyridine (DMAP) were dissolved in 400 mL of dichloromethane in a 1000 mL three-neck reactor, equipped with a mechanical stirrer and a dropping funnel. The mixture was then heated to 40 °C using a water bath. 20.45 g (50 mmol) of tetramethyl Bisphenol A bischloroformate dissolved in 120 mL of dichloromethane was added at a rate which was initially very high, but which became gradually lower toward the end of the reaction. The whole process took about 30 min.

The reaction mixture was then allowed to cool to room temperature, before being poured into a blender filled with a

polymers	Mn(*10^4)	PDI	T_g(°C)
BPA-PC	3.6	2.5	150
B_9C-PEC	3.6	2.3	158
B_7C-PEC	3.6	2.3	159
B_5C-PEC	2.8	2.6	161
B_3C-PEC	3.5	2.6	172
TMBPA-PC	3.1	1.6	200
T_9C-PEC	3.0	1.6	198
T_7C-PEC	3.6	1.5	197
T_5C-PEC	3.3	1.7	202
T_3C-PEC	3.2	1.8	206

Figure 3. Number average molecular weights (M_n), polymer dispersion index (PDI) measured by GPC with THF as solvent, and T_gs of copolymers measured by DSC at a rate of 10 °C/min.

Scheme 1



large quantity of 2-propanol. The polymer formed a white precipitate, which was filtered, and dried. The polymer was again dissolved in dichloromethane, precipitated from 2-propanol, filtered and dried. This step was repeated twice more in order to remove impurities thoroughly. The molecular weights and glass transition temperatures (T_g) are shown in Figure 3.

(2) Synthesis of 1,4-Cyclohexylene/Polycarbonate Block Copolymers. The polymers were synthesized using the solution condensation polymerization method. The oligomers were synthesized first by unbalanced feeding of Bisphenol A or tetramethyl Bisphenol A and their corresponding chloroformates. The oligomers were then coupled with 1,4-cyclohexylene dicarboxylic chloride to form the block copolymers. Cumylphenol was added to end-cap the polymers and to provide control of the molecular weight. The two series of copolymers obtained by this approach were denoted as B_xC and T_xC, where B represents Bisphenol A, T represents tetramethyl Bisphenol A, C represents 1,4-cyclohexylene dicarboxylate, and *x* represents the average number of repeat units within each block.

The synthesis procedure for B_xC and T_xC is shown in Scheme 1. The synthesis of B_3C is given here as it represents a typical method of preparation of these block copolymers.

Synthesis of B_3C Polyester Carbonate. A 15.99 g (70 mmol) sample of Bisphenol A, 12.36 g (35 mmol) of Bisphenol A chloroformate, 0.2820 g (1.33 mmol) of cumylphenol, 17.67 g (175 mmol) of triethylamine (TEA), and 1.71 g of *N*,*N*-dimethyl-4-aminopyridine (DMAP) were suspended in 400 mL of dichloromethane at 0 °C in a 1000 mL four-neck reaction flask equipped with a mechanical stirrer. The solution was then gradually warmed to room temperature and held at this temperature for about 30 min. Then 7.32 g (35 mmol) of 1,4-cyclohexylene dicarboxylic chloride dissolved in 50 mL of dichloromethane was added at a rate which was initially very high, but which became gradually lower toward the end of the reaction. The whole process took about 30 min.

The reaction mixture was poured into a blender filled with a large quantity of 2-propanol. The polymer formed a white precipitate, which was filtered and dried. The polymer was again dissolved in dichloromethane, precipitated from 2-propanol, filtered, and dried. This step was repeated twice more to further remove impurities. The molecular weights and glass transition temperatures (T_g) are shown in Figure 3.

B. Sample Preparation for Dynamic Mechanical Analysis (DMA). DMA was performed on thin film specimens of typical dimensions of 50 mm × 5 mm × 0.06 mm (length × width × thickness). The thin films were cast from 10% polymer solutions using 1,2-dichloroethane as the solvent. The polymer solutions were carefully cast on glass plates and kept at room-temperature overnight in order to evaporate most of the solvent. Then, the thin films were cut into strips with dimensions mentioned above and peeled from the glass plates. The sample strips were held under vacuum at 100 °C for at least 24 h for the removal of any remaining solvent or moisture. Prior to the DMA measurements, the samples were heated to 15 °C above their respective T_gs for 15 min and then quenched down to room temperature in order to erase the thermal history. All the film specimens were visually clear, suggesting the absence of crystallinity, which was also proven using X-ray diffraction.

C. Sample Preparation for Uniaxial Tensile Test. The dog-bone-shaped tensile bars for the polymers were compression molded at temperatures 70 °C above their respective T_gs, and quenched down to room temperature. The specimens were then polished using fine grit sand paper. Prior to the measurements, the specimens were placed back into the mold and heated to 15 °C above their T_gs for 15 min to erase any thermal and mechanical history, before being quenched back down to room temperature. The dimensions of the specimens were 50 mm × 3 mm × 3 mm (length × width × thickness). The gauge length was 12.7 mm.

Results and Discussion

Dynamic Mechanical Behavior. The sub-T_g secondary relaxation peaks are identified as β, γ, δ, etc., in the order of decreasing temperature below T_g, which is designated α. In general, the β-relaxation in polymer glasses forms a shoulder on the lower temperature side of the T_g and is actually due to structural relaxation rather than a specific motion because its appearance and shape depend on the thermal history of the specimen. On the other hand, the γ-relaxation is only slightly affected by the thermal history and is due, in the present cases, to motions of polymer segments as we demonstrate below. Its occurrence at a given temperature and frequency indicates that movement of some molecular group in the polymer chain is activated at that temperature and time scale.

The γ-relaxation behavior of B_xC and T_xC was studied with a home-built dynamic mechanical testing apparatus. We want to find out whether the motion of cyclohexylene groups couple with those of their neighbors. In Figure 4, the dynamic mechanical spectra (DMS) of BPA−PC, TMBPA−PC, and poly(1,4-cyclohexylenedimethylene 1,4-cyclohexylenedicarboxylate) (PCC) at 1 Hz are compared. BPA−PC has a broad γ-relaxation peak at −100 °C, whereas that of TMBPA−PC occurs at 40 °C. These low-temperature relaxations have been attributed to cooperative molecular motions involving several repeat units.[15] PCC has a prominent

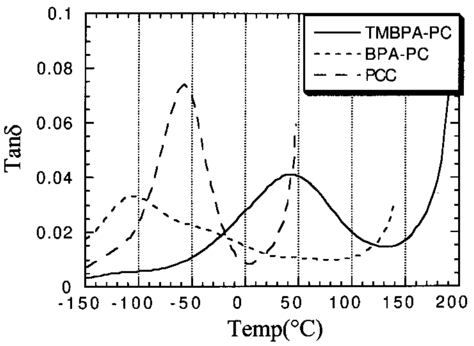


Figure 4. DMS (1 Hz) of BPA−PC, TMBPA−PC, and PCC.

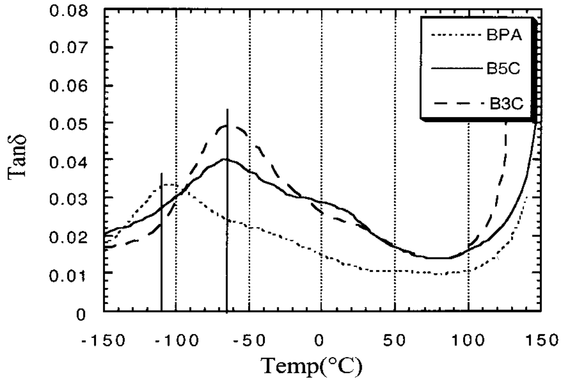


Figure 5. γ-relaxations from DMS of B_xC at 1 Hz.

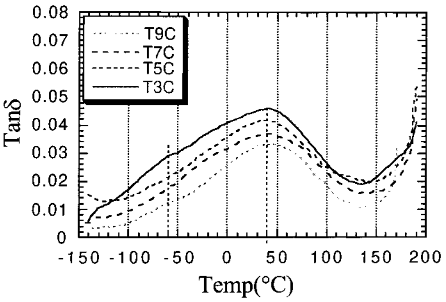


Figure 6. γ-relaxations from DMS of T_xC at 1 Hz.

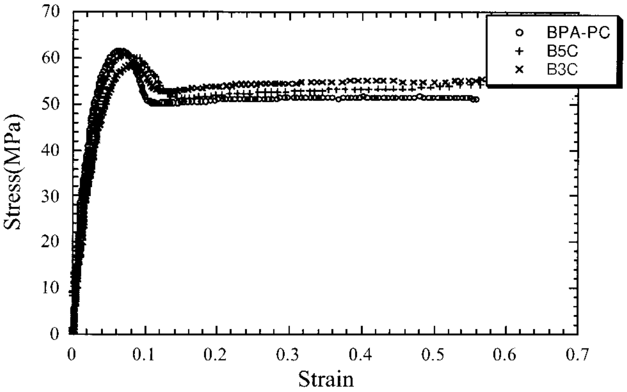


Figure 7. Stress−strain curve of B_xC at a strain rate of 2.62×10^{-3}/s at room temperature.

γ-relaxation peak at −60 °C. The study on PET/PCT and related copolymers by Chen et al. indicates that this relaxation originates from the chair−boat−chair conformational transition of the cyclohexylene ring.[12] If segmental motions of the polycarbonate blocks and the conformational transition of the cyclohexylene group are independent of each other in B_xC and T_xC, then the relaxation peaks for the copolymers could be produced by a simple superposition of the appropriate peaks. Otherwise, other relaxation mechanisms may be involved. The γ-relaxation peaks of B_3C and B_5C copolymers, as shown in Figure 5, appear at −65 °C. If these peaks are compared with the peak position of the γ-relaxation in PCC, it becomes apparent that the γ peaks of B_3C and B_5C copolymers are associated with the conformational transition of the cyclohexylene ring. The shapes and positions of the γ-relaxation of the B_3C and B_5C copolymers are quite dissimilar from those of Bisphenol A polycarbonate, which suggests that a different relaxation mechanism is involved in these copolymers. A separate relaxation peak corresponding to the Bisphenol A polycarbonate blocks cannot be found at −100 °C, which is perhaps not surprising if the motions of the Bisphenol A polycarbonate blocks are coupled with those of the cyclohexylene rings. It appears that the full relaxation of the BPA blocks can occur only if the conformational transition of the cyclohexylene ring has been "switched on". When the conformational transition of cyclohexylene rings is "frozen", the motions of Bisphenol A polycarbonate blocks are hindered, presumably until the block length equals or exceeds seven, at which point the native local motion of the BPA−PC segments[11] can resume without any help or hindrance from the cyclohexylene rings.

In Figure 6, all the T_xC copolymers exhibit a γ-relaxation peak at 50 °C, which is approximately the peak activation temperature for the local segmental motions in tetramethyl Bisphenol A polycarbonate[11] at the same frequency. The peaks become less symmetric with increasing cyclohexylene ring content. Specifically, the peaks become broader on the low temperature side, which implies that even though the tetramethyl Bisphenol A polycarbonate segmental motions have not been completely activated, the presence of the cyclohexylene ring increases the mobility of the polymer chain. However, even though the possibility of conformational transition of cyclohexylene rings exists, full relaxation occurs only when tetramethyl Bisphenol A polycarbonate block motions are activated at 50 °C. There is no relaxation peak at −65 °C, which indicates that the conformational transition of cyclohexylene ring can only gradually become free to move as its neighbors become more mobile. This is in accordance with our explanation for the onset of the γ-relaxation of B_xC copolymers, which required the activation of the higher temperature process. Thus, the conformational transition of the cyclohexylene ring couple with the kinematics of its neighboring segments. We also note that tetramethyl Bisphenol A segments as short as three could be activated. Likewise, in B_xC, as soon as the cyclohexylene ring was activated, the number of segments had no effect on the peak position.

It may also be noted that additional relaxation features were observed in the form of bumps on the main curves between the γ-relaxation peak and the α-relaxation peak (glass transition) for B_xC copolymers. This may be due to structural relaxation or some additional relaxation resulting from interaction between cyclohexylene units and B_x blocks. While we are not able to clearly establish the true origin of this relaxation, neither of the possible reasons detract from the argument about the interdependence of motions of the cyclohexylene ring and polycarbonate blocks in those copolymers.

Uniaxial Tension Behavior. In Figure 7, all of the B_xC copolymers show a stress−strain behavior typical

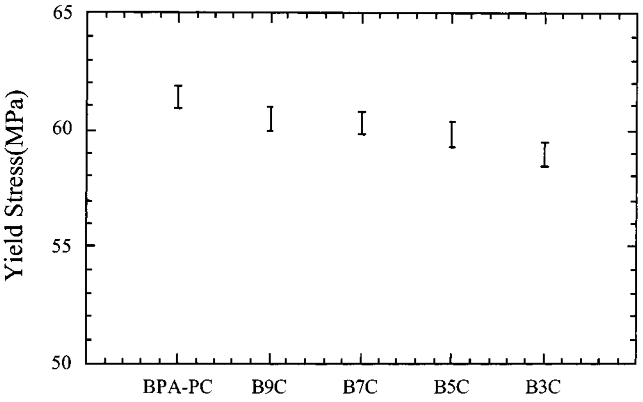


Figure 8. Yield stresses of B_xC at a strain rate of 2.62×10^{-3}/s at room temperature.

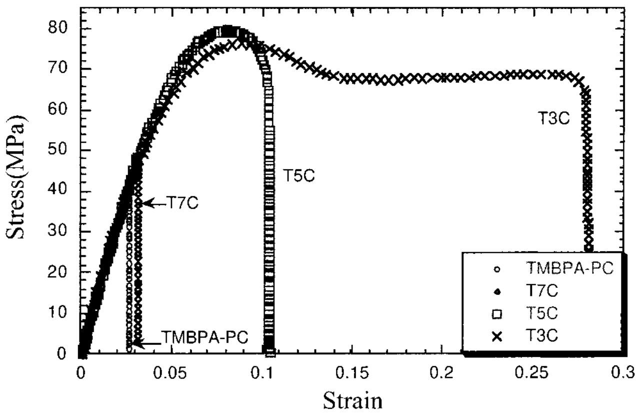


Figure 9. Stress−strain curve of T_xC at a strain rate of 3.28×10^{-4}/s at room temperature.

for ductile plastics. The yield stresses were determined from quadruplicate tests for each material (Figure 8), and decrease slightly with increasing cyclohexylene ring contents. By comparison, the yield stresses of terephthalate/Bisphenol A polyester carbonate(B_xt, where t represents the terephthalate linkage) increase with terephthalate content.[16] The difference between the terephthalate linkage and the cyclohexylene linkage is that the former has a planar structure and is therefore capable of only a π-flip motion instead of the conformational transition available to the cyclohexylene ring. The π-flip motion of the terephthalate can presumably occur without any cooperative from in-chain neighbors.

Tetramethyl Bisphenol A polycarbonate is an extremely brittle polymer at room temperature (Figure 9). With the incorporation of 20 mol % or more cyclohexylene linkages, the copolymers become ductile. The higher the cyclohexylene content, the more ductile the polymer becomes without significant changes in the molecular weight. Indeed, the copolymer was able to yield when the block length of tetramethyl Bisphenol A polycarbonate became equal to five. When the block length was further reduced to a value of three, the copolymer displays even more ductile behavior. On the other hand, the T_3t (t again represents terephthalate linkage) copolymer remained brittle (Figure 10), irrespective of the terephthalate content.

In Figure 9, we can see that T_7C is still brittle. This is probably because when the block is too long, the conformational transition of cyclohexylene rings can no longer facilitate the motion of all of the attached segments. Furthermore, as shown in Figure 6, the conformational transition of cyclohexylene linkages is always hindered by the immobile tetramethyl Bisphenol A segments at room temperature, even though tetramethyl Bisphenol A segmental motions can be activated at 50 °C. As shown later in this paper, the conformational transition of cyclohexylene linkages requires a large angular variation of bonds linking it to its neighboring groups. If the neighboring segments are too bulky and immobile, the conformational transition will be hindered because it is difficult to accommodate the angular variation. So for T_7C, which has the long tetramethyl Bisphenol A blocks, the incorporation of cyclohexylene linkages is not sufficient to make it ductile.

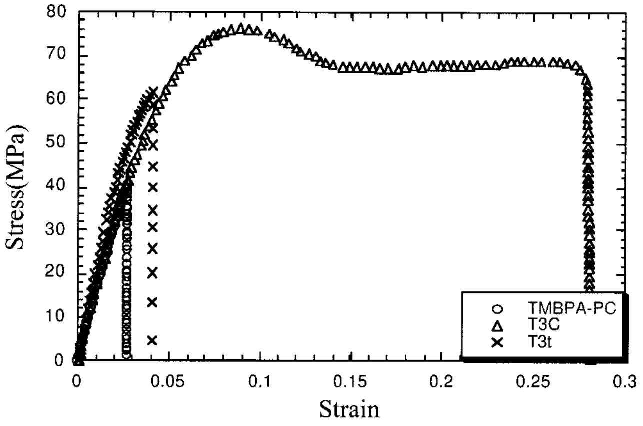


Figure 10. Stress−strain curve of TMBPA−PC, T3C, and T3t at a strain rate of 3.28×10^{-4}/s at room temperature.

It will be much clearer if we have yield stress data for T_xC at temperature above 50 °C. Nevertheless, we can expect that TMBPA−PC and T_xC would be able to yield, because the segmental motions were fully activated at such conditions. The yield stresses would decrease with the increasing cyclohexylene linkage contents.

The results presented suggest that the conformational transition of cyclohexylene rings is accompanied by an increased tendency for shear yielding in B_xC and T_xC copolymers.

Motions of Cyclohexylene Rings. We now wish to discuss a possible means by which the cyclohexylene linkage may interact with its neighbors to expedite the homogeneous shear yield process. The cyclohexylene ring has only two most stable conformations, which are the chair conformations. A transition occurs between these two conformations via an intermediate, twist boat conformation (Figure 2). The conformational transition was described in detail by Chen[11] and a point to be emphasized is that in order for the ring to go through the conformational transition, the neighboring ester groups must undergo a fairly large excursion. As shown in Figure 11 (simulated using Cerius2), in the model molecule, a *trans*-dimethyl 1,4-cyclohexylenedicarboxylate, the angle between bonds A and B, which are the bonds linking 1,4-cyclohexylene to its neighbors, changes from 180 to 114°. This angular variation during the transition consequently forces the neighboring, connected segments to translate and/or rotate, depending on what the cyclohexylene group at the other ends of the segments are doing, and on the shape and flexibility of the segments themselves. In B_xC, this movement overlaps with the intrinsic motions of Bisphenol A polycarbonate blocks, and enhanced segmental mobility is achieved around cyclohexylene units. Therefore, it

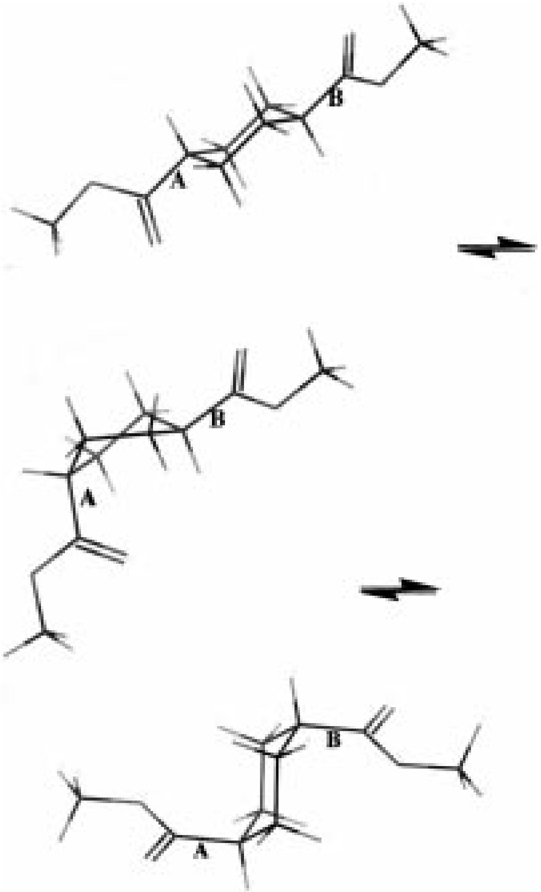


Figure 11. Conformational transition of dimethyl 1,4-cyclohexylenedicarboxylate.

is proposed that the enhanced mobility can facilitate relaxation of stresses built up by an imposed strain by helping polymer chains in sliding past each other, which will expedite homogeneous shear yield under stress and prevent early craze initiation. In a separate paper, we will show how the conformational transition of cyclohexylene rings will increase craze stress significantly.

Conclusion

The incorporation of 1,4-cyclohexylene dicarboxylate enables tetramethyl Bisphenol A polyester carbonates to exhibit ductile yield behavior while reducing the yield stress in Bisphenol A polyestercarbonates. It is proposed that this is possible due to the conformational transition of the cyclohexylene group being able to couple with the motions of neighboring segments, and this further facilitates the molecular motion of neighboring segments, which processes are advantageous for shear yield.

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