

Fracture toughness and fracture mechanisms of PBT/PC/IM blends

Part V Effect of PBT-PC interfacial strength on the fracture and tensile properties of the PBT/PC blends

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To investigate the effect of PBT-PC interfacial strength on the fracture toughness and toughening mechanisms of the PBT/PC system, a series of PBT/PC blends with different content of *in situ* formed PBT-PC copolymers were made by melt blending. The *in situ* copolymer was separately prepared via reactive blending of the PBT and PC in the presence of a transesterification catalyst in a twin-screw extruder for a few minutes. The reactive extrudate (RE) was studied using a DSC and the existence of the PBT-PC copolymer in the RE was confirmed. Microstructure characterizations of the PBT/PC/RE blends revealed that the domain sizes of the PBT and PC decrease and the PBT-PC interfacial strength increases with the RE content. Compared with the PBT/PC blend, all the PBT/PC/RE blends have higher yield strength, elongation at break as well as tensile modulus. The quasi-static fracture tests show that fracture toughness of the blends increases with the RE content. Since the highest toughness was obtained with the blend having the highest RE content (7.5%), it is not certain at this stage whether adding more than 7.5% RE will further improve the fracture toughness. The impact toughness of the PBT/PC/RE blends was found to decrease with the increase of the PBT-PC interfacial strength, which confirms the failure mechanisms proposed in the Part-4 of this series. © 2003 Kluwer Academic Publishers

1. Introduction

In the Part-1 to -4 of this series [1–4], the fracture behavior and toughening mechanisms of the PBT/PC system were studied. The possible relationships between the morphology and mechanical property of the blends were discussed and proposed. In the study of the rigid-rigid PBT/PC blends without rubbery modifier [3], a synergetic effect was obtained in the blends containing 40% to 50% PBT. The quasi-static fracture toughness of the blends was much higher than that of the neat PC or PBT. Meanwhile, both the modulus and the yield stress showed positive deviation from the rule of mixtures. On the other hand, the blends containing more than 60% PBT showed both poor tensile properties and low fracture toughness. Toughening mechanism study

disclosed that the PBT-PC interfacial adhesion in the toughened blends were stronger than those PBT-rich blends and, the observed debonding-cavitation at the PBT-PC interface was a key process that ensures massive plastic deformation to take place and prevent the blends from fast unstable failure. It was further proposed in the Part-3 of the series [3] that the strength of the PBT-PC interface plays a critical role in determining the fracture toughness of the PBT/PC system. Neither excessively strong nor very weak interfacial bonding strength is beneficial to the fracture resistance of the system. There might be an optimum interfacial bonding strength, which would render the polymer blends with the highest toughness as well as improved yield strength and tensile modulus. To verify this assumption,

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a systematic investigation on the fracture behavior of the PBT/PC blends with same composition but different interfacial adhesion strength is necessary.

To change the interfacial adhesion strength of an immiscible or a partially miscible blend, compatibilizers are often used. Research shows that compatibilizer tends to concentrate at the interface of the blend components, reduces the interfacial tension, prevents coalescence and strengthens interfacial adhesion [5–7]. Most compatibilizers are either block copolymers or graft copolymers, which can be commercial copolymers or formed *in situ* during reactive melt blending of the polymer components through interfacial chemical reactions or exchange reactions. As the commercial copolymers are usually expensive and suitable copolymers for a particular blend system may not be always available, the *in situ* copolymer formed during reactive blending is regarded a better candidate for interface modification.

Extensively research [8–11] has demonstrated that PBT/PC is a partially miscible blend. For the above-mentioned reason, the PBT-PC copolymer can be used to alter the PBT-PC interfacial adhesion strength. In the present study, we attempted to produce the PBT-PC copolymers through the well-known PBT-PC transesterification [12, 13] by extrusion of PBT and PC in the presence of a transesterification catalyst, antimony oxide (Sb_2O_3) [13] with a high shear rate screw configuration. The products of the reactive extrusion were expected to have large amount of PBT-PC copolymers, which were then used as a compatibilizer in the preparation of the PBT/PC blends. With addition of different amounts of the *in situ* copolymer into the PBT/PC blends, the PBT-PC interfacial adhesion strength should change with the copolymer content. To prove the existence of the PBT-PC copolymer, the product from the reactive extrusion was tested using DSC. The microstructure, particularly the PBT-PC interfacial structure, of the blends was examined using both SEM and TEM. Mechanical properties, including tensile, quasi-static and impact fracture toughness, of the blends with various interfacial conditions were carefully measured and compared. The relationships between the mechanical properties and the interfacial boundary condition were discussed.

2. Experimental work

The Poly(butylene terephthalate) (PBT, VALOX 315) and Polycarbonate (PC, CALIBRE 300) used in the present study were supplied by the G.E Plastic and the

Dow Chemical Company, respectively. The pellets of the two components were dried in an air-circulating oven at 120°C for 5 hours prior to compounding.

The PBT/PC blends were made using a two-step extrusion method. In the first step extrusion, a mixture of PBT and PC pellets was compounded in the presence of a transesterification catalyst, antimony oxide (Sb_2O_3). The weight ratio of the PBT : PC : Sb_2O_3 was kept as 40 : 60 : 0.5. The reactive blending was performed in a twinscrew co-rotating extruder (Werner and Pfleiderer ZSK-30, $L/D = 29/1$, where L is the total barrel length and D is the barrel bore diameter). Screw speed and barrel temperature were predetermined, as shown in Table I. The total resident time of the PET and PC inside the extruder was about 8 minutes, which was achieved by double extrusions. The first-step compounding was designed to produce the PBT-PC copolymers *in situ* through the PBT-PC transesterification. The resultant PBT/PC/ Sb_2O_3 blend was semi-transparent. The pellets of the blend had a very thick transparent skin and a small milk-white core, which means that the crystallization rate of the blend is low. Due to the quenching in the water bath, the skin layer of the extrudate could not crystallize and remain in transparent amorphous state. The PBT/PC/ Sb_2O_3 blend obtained in the first-step compounding is designated, hereafter, as RE component. The RE component would be used later as an interface modifier to alter the PBT-PC interfacial strength in the final PBT/PC/RE ternary blends.

In the second step extrusion, the PBT/PC/RE blends, consisting of 40 : 60 by weight of PBT and PC with increasing levels of RE component, were prepared. The composition of the blends and the processing parameters were list in Table I. After extrusion, the PBT/PC/RE blends were dried at 120°C for 5 hours and then injection molded into dog-bone bars for tensile tests (ASTM D-638) and large rectangular plates for essential fracture work tests (ESIS protocol, version 5, 1997).

The occurrence of the PBT-PC transesterification was verified by means of differential scanning calorimetry (DSC). The melting and crystallization behavior of the PBT/PC binary blend and PBT/PC/RE ternary blends were tested under various heating-cooling cycles. For each cycle, a sample of ~10 mg was encapsulated in an aluminum pan and scanned from 25 and 280°C. The heating rate was 10°C/min. At the end of the heating process, sample was kept at 280°C for 5 minutes and then quenched down to temperature below zero before another heating process.

The tensile property of the blends was tested at ambient temperature following the ASTM D-638 with a

TABLE I Compositions and processing conditions of PBT/PC blends

Material	Composition	Code	Processing temperature (°C)				Die	Screw speed (rpm)
			Zone 1	Zone 2	Zone 3	Zone 4		
PBT/PC/ Sb_2O_3	40/60/0.5	RE	240	260	265	280	240	16
PBT/PC/RE	40/60/0		240	250	260	250	230	17–19
PBT/PC/RE	40/60/1.5		240	250	260	250	230	17–19
PBT/PC/RE	40/60/3.0		240	250	260	250	230	17–19
PBT/PC/RE	40/60/5.0		240	250	260	250	230	17–19
PBT/PC/RE	40/60/7.5		240	250	260	250	230	17–19

Sintech D/10 Universal Tensile Machine at a crosshead speed of 10 mm/min. The quasi-static fracture toughness was obtained in terms of the specific essential work of fracture using the double-edge-notched-tensile (DENT) technique, following the European Structure Integrity Society (ESIS) Test Protocol for Essential Work of Fracture Version 5, 1997. The geometry of the specimen was $3.2 \times 39 \times 156$ mm. The details on the sample preparation and measurement can be found in one of our earlier papers [14].

The morphology of the blends was revealed by transmission electron microscopy (TEM) and scanning electron microscopy (SEM). Ultra-thin sections for TEM analysis were cut using a Leica Ultracut-R Ultramicrotome and both the thin sections perpendicular or parallel to the mould filling direction (MFD) were prepared. The thin sections were stained with RuO_4 and examined using a JEOL 100 CXII TEM. The specimens for SEM were prepared by fracturing the blend samples in liquid nitrogen and the SEM observations were performed on a JEOL 6300 SEM. More information on TEM and SEM sample preparation can be found in Ref. [3].

3. Results and discussion

3.1. DSC study

The evidence of the PBT-PC transesterification was obtained through our DSC study. Fig. 1 displays a group of DSC curves obtained after different numbers of heating cycles for the PBT/PC (40/60) binary blend without catalyst. The curves clearly demonstrate that the melting temperature (T_m) of the crystalline phase of the blend shifted to lower temperatures with increase of heating cycle numbers, in other words, accumulative annealing time at 280°C . This decreasing trend of T_m reflects that the integrity of the PBT crystals decays with prolonged melt holding time. A comparison among the four DSC curves in Fig. 1 further suggests that this decay is most probably caused by the transesterification between the PBT and PC. As can be seen from Fig. 1, after the sample was annealed at 280°C for 10 minutes, a large exothermic peak was observed on the third-run DSC curve at about 150°C and, at the meantime, the T_m shifted down to 210°C , which is 17°C lower than the T_m on the first-run DSC curve. Further increase of the annealing time at 280°C to 15 minutes, a very

broad exothermic peak was found on the fourth-run DSC curve at about 157°C , which is 7°C higher than that on the third-run DSC curve, meaning that the recrystallization of the blend became more difficult. The T_m on the fourth-run DSC curve is the lowest among the four melting temperatures. It is about 198°C and 29°C lower than the first-run T_m . The melting process in the fourth DSC run, in fact, started before the completion of the exothermic process. The above-mentioned experimental evidences strongly suggest that the transesterification between the PBT and PC occurred during the DSC test and the degree of the transesterification is higher after a longer annealing time at 280°C . The products of the transesterification, i.e. PBT-PC copolymers, have significant influence on the recrystallization and melting of the PBT/PC blend.

The occurrence of the PBT-PC transesterification is further confirmed by the facts that addition of a transesterification catalyst, Sb_2O_3 , into the PBT/PC system would accelerate the decreasing trend of the T_m and make the exothermic process to occur in an earlier stage. Fig. 2 shows a group of DSC curves obtained with the PBT/PC/ Sb_2O_3 blend. For comparison, the first-run DSC curve of the PBT/PC blend without Sb_2O_3 is placed at the bottom of the figure. Similar to the PBT/PC blend, the T_m of the crystalline phase of the PBT/PC/ Sb_2O_3 blend decreases with the increase of the annealing time at 280°C . However, the T_m decreases much faster for the PBT/PC/ Sb_2O_3 than that for the PBT/PC blend, as can be seen in Fig. 3. Moreover, it is noted that the third-run DSC curve of the PBT/PC/ Sb_2O_3 shows the similar characteristics as the fourth-run DSC curve of the PBT/PC blend. These observations suggest that the transesterification catalyst Sb_2O_3 can effectively accelerate the PBT-PC exchange reaction. The effect of the catalyst becomes even more evident in the fourth DSC run of the PBT/PC/ Sb_2O_3 blend, on which neither endothermic nor exothermic peak can be seen. In fact, the fourth-run DSC curve in Fig. 2 has the characteristics for a typical immiscible binary amorphous polymer blend with two glass transitions at 86°C and 127°C . Since the PC content is higher (60 phr) than the PBT (40 phr) in the blend, the higher one (127°C) is the T_g of the remaining amorphous PC and the lower T_g (86°C) is most probably the glass transition temperature of the PBT-PC copolymers. On the

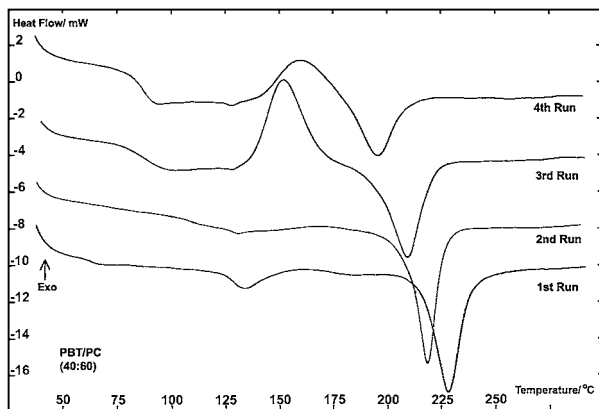


Figure 1 DSC thermograms for the PBT/PC binary blends after different accumulative annealing time at 280°C .

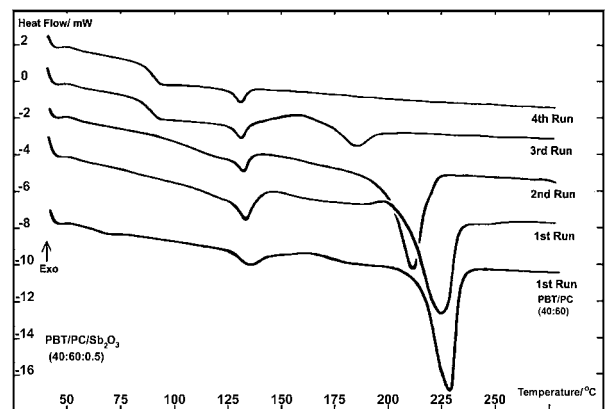


Figure 2 DSC thermograms of the PBT/PC/ Sb_2O_3 blends after different accumulative annealing time at 280°C .

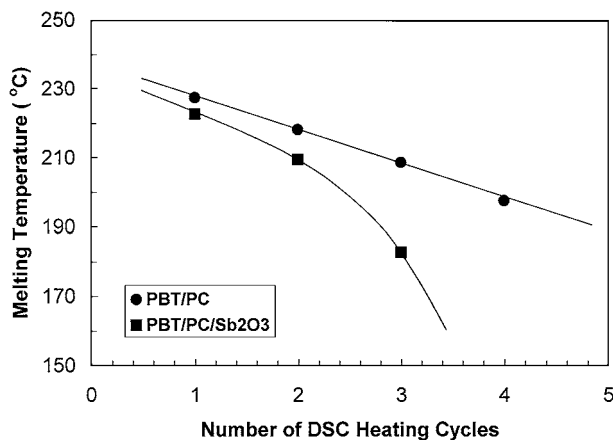


Figure 3 Variation of melting temperature of the PBT/PC and PBT/PC/Sb₂O₃ blends after different accumulative annealing time at 280°C.

other hand, no exothermic and endothermic process implies that the PBT-PC copolymers formed through the PBT-PC transesterification are unable to crystallize under the present DSC testing condition and remain as amorphous polymers. This supposition is supported by the semi-transparent appearance of the RE component, which is a PBT/PC/Sb₂O₃ blend, mixed in an extruder for 8 minutes.

Given the above-discussion, some conclusions may be drawn. After a prolonged annealing at 280°C, transesterification between the PBT and PC will occur and PBT-PC copolymers will form. The degree of the transesterification depends on the annealing time at high temperature and use of catalyst. Since the copolymers are unable to crystallize under the present testing condition, the exothermic peak on the third- and fourth-run DSC curves in Fig. 1 cannot be attributed to the crystallization of the copolymer. Thus, the peak is most likely a result of the PBT crystallization, which cannot complete during the cooling processing from 280°C. This result indicates that the PBT-PC copolymer must have a great hindrance to the PBT crystallization, because the PBT crystallization is very fast and, without interference, it will complete during cooling. As demonstrated by the second-run DSC curve in Fig. 1, no exothermic peak is found, meaning that the PBT crystallization has completed during the cooling process after the first DSC run. The PBT-PC copolymers formed in the first DSC run was probably in a very limited amount, which could not stop the PBT recrystallization, but, decreased the T_m of the blend from 227°C to 218°C. The glass transition of the PBT-PC copolymer was found on all the third- and fourth-run DSC curves in both Figs 1 and 2 at about 86°C, which is higher than the T_g of PBT (~74°C) and lower than the T_g of PC (~127°C). Noticeably, the T_g of the PBT-PC copolymer does not change with the annealing time at 280°C.

Fig. 4 is the first-run DSC curves obtained with the PBT/PC/RE ternary blends having different contents of the RE component. A common feature of the curves, except the curve for the blend without RE component, is the exothermic peak at about 130°C (as indicated by the arrows). Considering that the PBT-PC copolymer in the RE component could not crystallize under the present testing condition, the observed exothermic peak is most

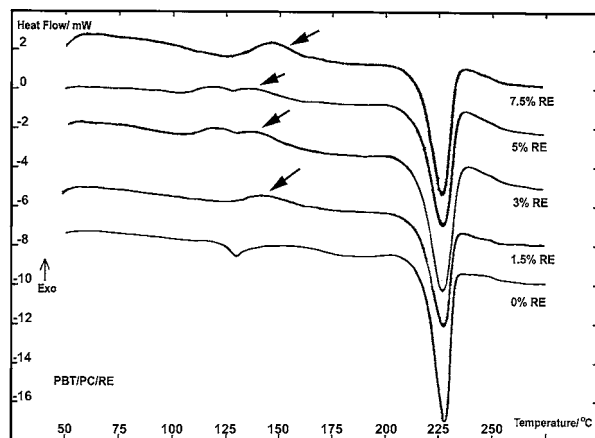


Figure 4 First-run DSC thermograms for the PBT/PC/RE ternary blends with different RE contents.

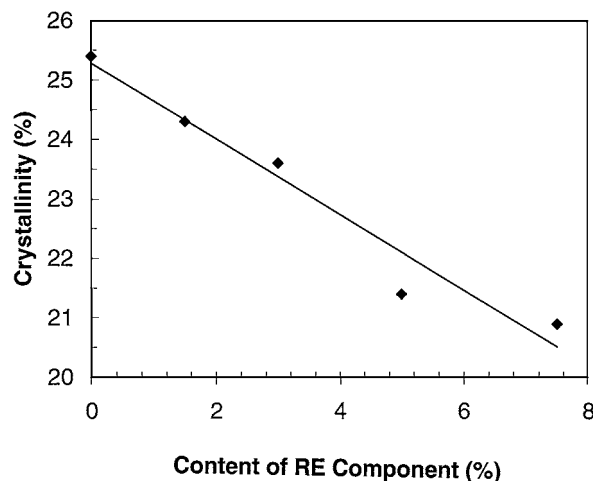


Figure 5 TEM micrographs of the PBT/PC binary blend and the PBT/PC/RE ternary blends.

probably caused by the crystallization of the PBT portions, which could not crystallize during cooling because of the hindrance from the PBT-PC copolymer. Since the area of the exothermic peak increases and the crystallinity of the PBT/PC/RE blends decreases with the RE content, refer to Fig. 5, the influence of the RE component on the PBT crystallization must increase with the RE content. On the other hand, the added RE component is anticipated to form a third phase located at the interface between the PBT and PC phases, because of its PBT-PC copolymer nature. If this proposition is correct, then, the crystallization of the PBT portions near the PBT-PC interface will be most severely influenced. In other words, a larger exothermic peak at high RE content means a stronger interaction between the PBT and RE component. The experimental results in Fig. 4 show that the RE component can act as a compatibilizer to improve the compatibility between the PBT and PC, which may lead to various morphology, interfacial strength and mechanical property changes, as will be discussed in the following text.

3.2. Morphology study

Fig. 6a–e are the TEM micrographs of the PBT/PC/RE blends with different contents of RE component. The common feature of the micrographs is a PBT-PC co-continuous structure, where the dark domain is the PC

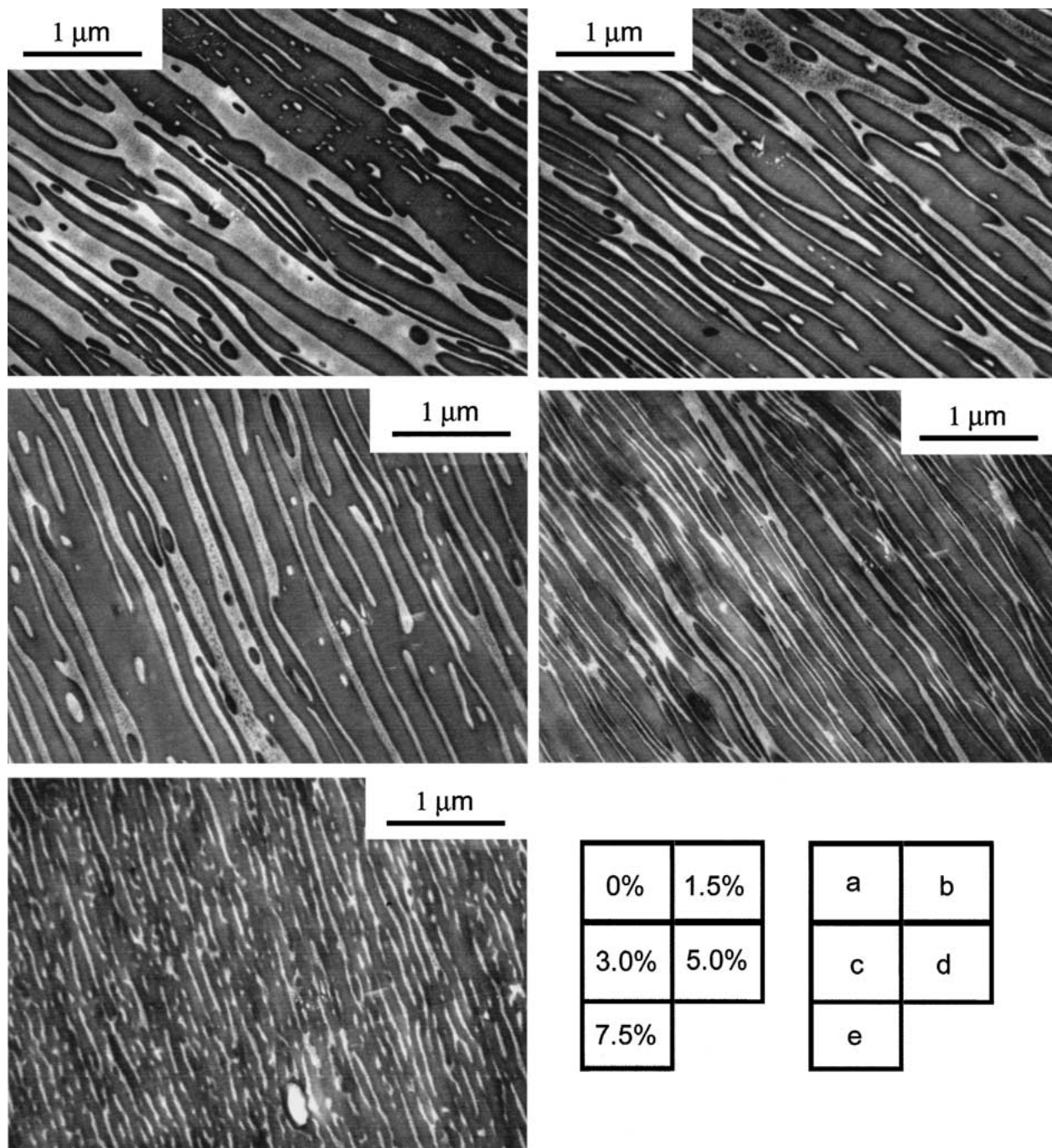
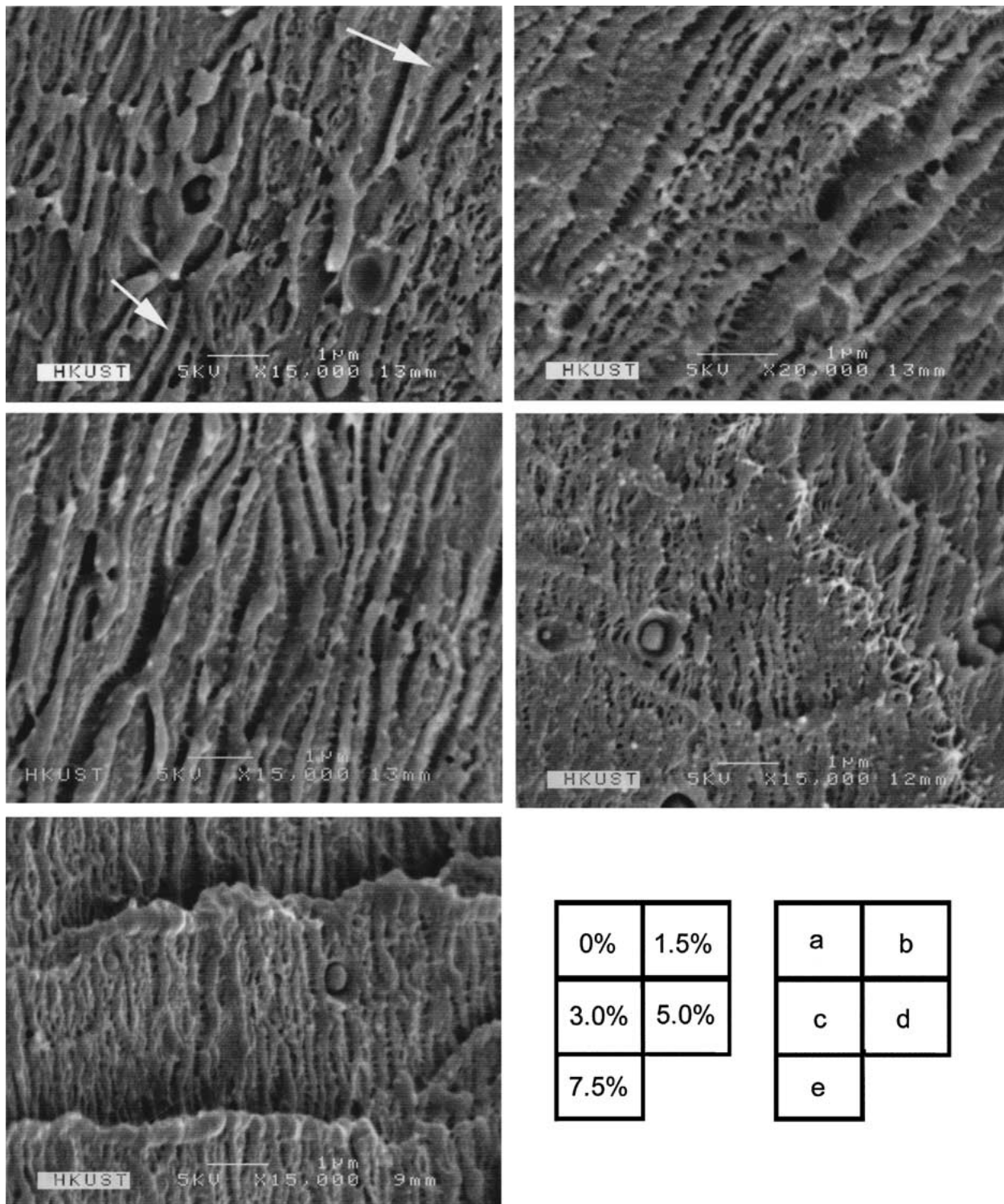


Figure 6 SEM micrographs of PBT/PC binary blend and PBT/PC/RE ternary blends.

phase because it has higher affinity with RuO_4 than the PBT [3]. A comparison of the micrographs shows that addition of the RE component would reduce the domain size of the PBT (gray phase) and PC, especially at higher RE contents. For example, the PBT domain width of the blend without RE component was about $0.2\text{--}0.4\ \mu\text{m}$, refer to Fig. 6a. When 5% of RE component was added, it was reduced to the range of $0.02\text{--}0.1\ \mu\text{m}$, refer to Fig. 6d. At 7.5% RE content, the PBT domains show a tendency to break into fine particles with a rather uniform width of approximately $0.05\ \mu\text{m}$. It is also noted that the variation in domain width decreased with RE content, too. A narrower domain width distribution was achieved at higher RE contents. Since the size of the dispersed phase is proportional to the interfacial tension for many immiscible polymer blends [15, 16], the observed domain size reduction reflects

that the interfacial tension between the PBT and PC was reduced and the coalescence of the domains was suppressed due to the addition of the RE component. It suggests that the *in situ* formed copolymers in the RE component has formed an interphase between the PBT and PC, which stabilizes and prevents the dispersed droplets from coalescence.

The SEM study on the fracture surfaces of the blends with and without the RE component supports the conclusions derived from the TEM study. Moreover, as can be seen in Fig. 7a–e, the interfacial structure of the PBT/PC/RE blends, which cannot be seen in the TEM micrographs in Fig. 6a–e, has now been clearly disclosed by the SEM micrographs. Evidently, not only domain size, but also the microstructure of the PBT-PC interphase changes with the RE content. For the PBT/PC binary blend (0% RE), a few fibrils connecting the PBT



0%	1.5%	a	b
3.0%	5.0%	c	d
7.5%		e	

Figure 7 Variation of yield Stress and tensile modulus of the PBT/PC binary blend and the PBT/PC/RE ternary blends against RE contents.

and PC domains were found at the interface, as indicated by the arrows in Fig. 7a. However, when the RE component was incorporated in the PBT/PC blend, large numbers of fibrils were found at the PBT-PC interface region, as shown by Fig. 7b to e. Particularly, at 5% and 7.5% RE contents, a knitting-like structure was found on the fracture surfaces of the ternary blends (refer to Fig. 7d and e), indicating that the *in situ* formed PBT-PC copolymer in the RE component much locate at the interface between the PBT and PC.

Combining the results from the DSC and microscopic studies, we may conclude that the RE compo-

nent made by the reactive extrusion contains certain amount of PBT-PC copolymers formed through the PBT-PC transesterification catalyzed by Sb_2O_3 . When the copolymer is blended with the PBT and PC, it forms an interphase bridging the PBT and PC domains. As a result, the compatibility of the PBT and PC is improved, resulting in a finer and more uniform microstructure in the ternary blends. The copolymer interphase is expected to alter the PBT-PC interfacial strength, which, in turn, will change the mechanical properties of the ternary blends, as will be discussed in the next section.

3.3. Mechanical properties

3.3.1. Tensile properties

One of the advantages of rigid-rigid polymer blends is that the toughness of the blends may be improved without sacrifice of other important mechanical properties. For the PBT/PC system under investigation, we have reported in the Part-3 of this series [3] that the fracture toughness of the PBT/PC blends can be enhanced with modulus and yield strength being slightly improved as well, so long as the interfacial adhesion between the PBT and PC domains is strong. However, it was also observed that when the PBT-PC interface was weak, the yield strength of the blends would be lower than that of the neat PBT or PC. Obviously, the PBT-PC interfacial adhesion strength plays a critical role in determining the tensile properties of the blends.

Given the fact that the interfacial boundary condition of the PBT/PC/RE ternary blends changes remarkably with the RE component, as demonstrated in the previous sections, the tensile properties of the blends should vary with the RE content. Figs 8 and 9 display the variation of the yield stress, tensile modulus, elongation at break and tensile stress of the ternary blends against the RE content. It is evident that addition of the RE component to the PBT/PC blend did not cause any nega-

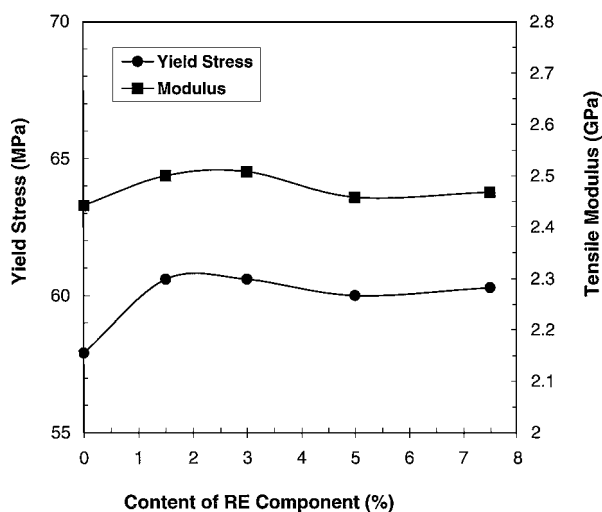


Figure 8 Variation of elongation at break and tensile strength of the PBT/PC binary blend and the PBT/PC/RE ternary blends against RE contents.

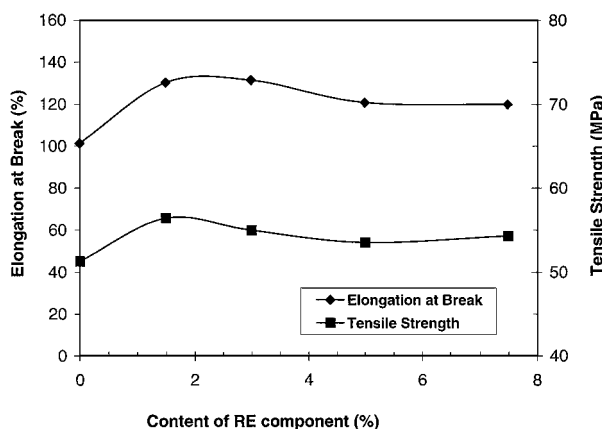


Figure 9 Variation of Young's moduli of PBT/PC (40/60) binary blend and PBT/PC/RE ternary blends against RE contents.

tive impact on the properties; instead, improvement of the four parameters was achieved. A maximum can be clearly identified on all four curves at RE content between 1.5% and 3%. The property enhancement is undoubtedly a result of the improved PBT-PC interfacial strength owing to the addition of the RE component. The decrease after the maximum suggests that there is an optimum RE content for the tensile properties and adding more RE component than the optimum amount will cause a reduction from the maximum. This is most likely due to the intrinsic low strength and poor ductility of the RE component. It was observed during the preparation of the RE component that the melt viscosity of the RE component was very low and the color of the RE component became light yellow after two extrusions in the presence of antimony oxide. In fact, the RE component was too brittle to be successfully injection molded into test bars for tensile test. It is believed that the RE component has a low molecular weight and a wide molecular distribution, due to the thermal degradation caused by long time extrusion, presence of catalyst and high shear stress of the screws [16].

3.3.2. Quasi-static fracture toughness

Under quasi-static fracture condition, we have demonstrated [3] that the fracture toughness of the PBT/PC blends containing 40% or 50% PBT was significantly higher than that of the neat PC and the PBT-rich blends. Toughening mechanism study revealed that [3] the debonding-cavitation at the PBT-PC interface plays a key role in the toughening of the PBT/PC system. For the PBT/PC blends containing 40 or 50% PBT, the interfacial adhesion between the PBT and PC is relatively strong. Under quasi-static loading, the PBT crazes formed in front of the crack tip, where tri-axial tension dominants, are stabilized by the well-bonded neighboring PC domains and unable to develop into harmful cracks. This stabilizing process prevents the sample from fast, unstable fracture and enables the sample to sustain with a higher loading. With further increasing the loading, the tri-axial tensile stress in front of the crack will reach the PBT-PC interfacial debonding strength, thus, debonding-cavitation between the PBT and PC domains will occur, which, in turn, releases the high plastic constraint at the crack tip and makes the plastic deformation in the PBT and PC domains possible. The large volume plastic deformation will blunt the running crack and absorb large amount of energy, leading to increased fracture toughness.

The core of the proposed toughening mechanism is obviously the PBT-PC interfacial strength and debonding-cavitation. When the interfacial strength is too low, the applied load cannot be effectively shared by the blend components, the PBT craze stabilization by the PC domains is impossible and, moreover, the weak PBT-PC interface will form the easy path for the crack to running through. When the interfacial strength is excessively high, however, the debonding-cavitation process may not be able to take place before the stress level at the crack tip reaches the fracture strength of material in front of the crack. The consequence is that the material will remain in a highly plastic constrained condition

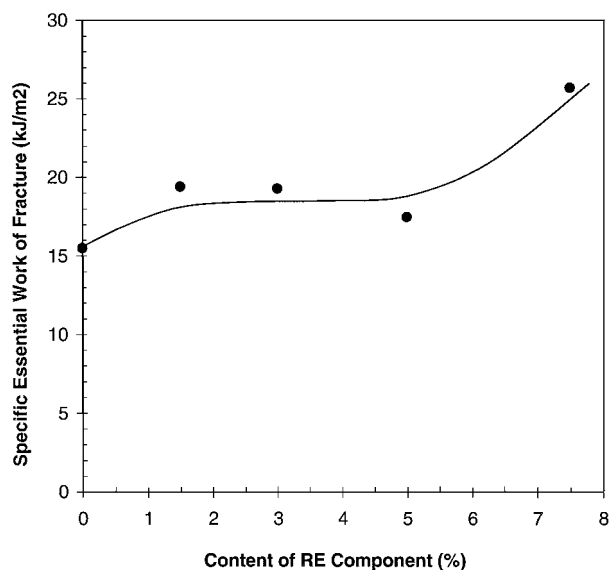


Figure 10 Variation of the specific essential work of fracture for the PBT/PC binary blend and the PBT/PC/RE ternary blends against RE contents.

and, eventually, unstable fracture of the material takes place without much plastic deformation. Based on this toughening model, there must be an optimum interfacial adhesion strength for the PBT/PC blends to gain the highest fracture toughness.

In the present study, the PBT-PC interfacial adhesion strength is expected to increase with the content of the RE component. This expectation has been confirmed by both the electronic microscopy and the tensile tests. The fracture toughness results of the PBT/PC/RE ternary blends having different interfacial strength, as shown in Fig. 10, demonstrate that the specific essential work of fracture of the blends increases with the increase of the RE component, in other words, PBT-PC interfacial adhesion strength. The highest fracture toughness, 25.8 kJ/m^2 , is achieved with the blend with 7.5% of the RE component, which is 60% higher than that of the PBT/PC without added RE component (15.5 kJ/m^2) and about 5 times of the fracture toughness of the neat PC tested under identical condition [3]. On the other hand, it is also noted that the optimum RE content for the highest specific essential work of fracture has yet to be determined. As suggested by the curve in Fig. 8, the increase of the specific essential work of fracture with the RE content is not significant at RE content of 5% or below. Marked improvement in fracture toughness is seen at 7.5% of RE addition and, at this stage, it is not certain whether adding more than 7.5% RE component will further improve the fracture toughness and at which RE loading the fracture toughness will start to decrease.

3.3.3. Impact strength

Due to the very high deformation rate of impact test, the impact strength of the PBT/PC blends, which have high fracture toughness under quasi-static condition, was found to be lower than that of the neat PC and PBT. The strong PBT-PC interface, consisting of rigid PBT-PC copolymer, was proposed in the Part-4 of this series to have a negative effect on its impact strength [4].

This is because PC is a stress-state sensitive material and undergoes brittle fracture when it is subjected to a plane-strain condition whereas the PBT is a strain-rate sensitive material and tends to have higher rigidity and lower fracture toughness when deformation rate is high. When the PBT/PC blends having co-continuous microstructure are tested under high strain-rate impact, the PBT domains become more rigid and brittle. They will impose a higher plastic constraint on the nearby PC domains if the interfacial bonding between the two components is strong. The high plastic constraint from the PBT domains will maintain the PC domains in a plane-strain dominant condition, which will promote the PC domains to break in brittle mode. In the quasi-static condition, because of the low deformation rate and, most importantly, because of the debonding-cavitation at the PBT-PC interface, the plastic constraint on the PC domains is released before brittle fracture takes place. This debonding process cannot happen in impact test because the PBT-PC copolymer at the interface is too rigid to complete the debonding-cavitation process.

Based on the failure mechanism discussed in the above text [4], one would predict that the impact toughness of the PBT/PC blends would decrease with the increase of the PBT-PC interfacial strength. This prediction has been proven to be true by the present study. The impact toughness of the PBT/PC/RE ternary blends, shown in Fig. 11, decreases almost linearly from 7.25 kJ/m^2 to 5.75 kJ/m^2 when the RE content increases from 0% to 7.5%, although the quasi-static fracture toughness of the blends have been improved substantially within this RE content range. This observation clearly suggests that the key to improve the impact strength of the PBT/PC system is to reduce the strain rate sensitivity of the PBT and release the plastic constraint on the PC. One way, probably, is to introduce a small amount elastomer into the PBT phase and another is to introduce a more ductile interphase between the PBT and PC, so that debonding-cavitation may be completed to release the plastic constraint. Some results of our attempt on introducing a reactive elastomer

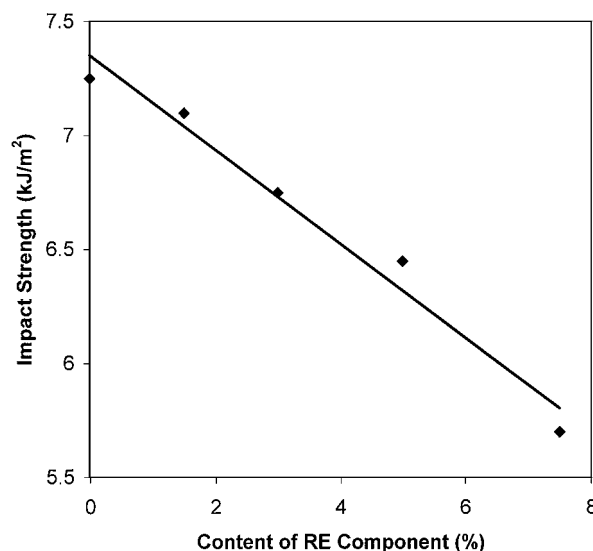


Figure 11 Variation of Charpy impact strength of the PBT/PC binary blend and the PBT/PC/RE ternary blends against RE contents.

into the PBT and the evaluation on the mechanical and fracture properties of the (PBT/elastomer)/PC blends will be reported in a coming paper.

4. Conclusions

PBT/PC blends with different interfacial adhesion strength were made by melt blending of the PBT and PC together with the *in situ* formed PBT-PC copolymers. The enhanced interfacial adhesion can effectively transfer the applied stress from one phase to the other, reduce the flaws in the material and result in improved the yield strength, elongation at break and Young's modulus.

Quasi-static fracture toughness tests of the PBT/PC/RE blends demonstrated that the PBT-PC interfacial strength has strong influence on the fracture behavior of the blends. Generally, the specific essential work of fracture of the blends will increase with the copolymer content, i.e. interfacial strength. In the present study, the highest fracture toughness was achieved at 7.5% RE content. Since this is the highest RE content used in this work, it is not certain whether addition of more than 7.5% RE will further improve the fracture toughness or at which RE content, the fracture toughness will start to decrease with the RE content increase.

Impact toughness of the PBT/PC/RE blends was found to decrease linearly with the RE content. This observation confirms the failure mechanisms proposed in the Part-4 of this series, where the brittle failure of the PBT/PC blends with relatively strong interfacial strength was attributed to the high deformation rate, high plastic constraint and no debonding-cavitation, or other mechanisms, to release the constraint.

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