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Recycled PET Packaging Materials of Improved Toughness— Importance of Devitrification of the Rigid Amorphous Fraction

Ferenc Ronkay, Dániel Gere, Emese Slezák, Edina Szabó, György Marosi, and Katalin Bocz*

Degradation, a common problem faced during the processing of recycled poly(ethylene terephthalate) (PET), leads to significant embrittlement of the products, as a result of which the material loses its applicability. Increased crystallization rate of the short chains of recycled PET and obstructed mobility of the amorphous phase are the main causes of enhanced brittleness. In this research, a straightforward method is proposed for improving the toughness of recycled PET products, namely the devitrification of the rigid amorphous phase by thermal annealing, which results in enhanced molecular mobility in the amorphous fraction, thereby promoting ductile deformation. The effects of thermal annealing conditions are comprehensively evaluated on the microstructure and macroscopic properties, i.e., impact resistance, of recycled PET films. The perforation energy value of the recycled PET film is found to increase to its threefold, reaching a value higher than 18 J mm⁻¹, as a result of 10 s thermal treatment at 120 °C. Differential scanning calorimetry, dynamic mechanical analyses, and thermally stimulated depolarization current measurements provide evidence for the devitrification of the rigid amorphous fraction under these conditions, which is the key to efficient enhancement in toughness.

F. Ronkay

Department of Innovative Vehicles and Materials GAMF Faculty of Engineering and Computer Science John von Neumann University Izsáki u. 10., Kecskemét H-6000, Hungary D. Gere Department of Polymer Engineering Faculty of Mechanical Engineering

Budapest University of Technology and Economics Műegyetem rkp. 3., Budapest H-1111, Hungary

E. Slezák, E. Szabó, G. Marosi, K. Bocz Department of Organic Chemistry and Technology Faculty of Chemical Technology and Biotechnology Budapest University of Technology and Economics Műegyetem rkp. 3., Budapest H-1111, Hungary E-mail: bocz.katalin@vbk.bme.hu

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1. Introduction

The World's plastic production is increasing every year, reaching 390.7 million tons in 2021.^[1] In the World (44%) and Europe (39%), most plastics are processed by the packaging industry.^[1,2] Regarding the production of plastic packaging material, poly(ethylene terephthalate) (PET) is the 3rd most processed type of plastic, along with polypropylene and polyethylene.^[1,3]

Due to their function, packaging materials have a very short lifetime (on average half a year), so they appear as waste within a short time.^[4–6] However, this waste fraction is much more valuable than sending it to a landfill, so efforts should be made to recycle it.^[7–9] Environmentally conscious producers already make their products from partially or completely recycled materials. However, only 8.3% of World plastics production was post-consumer recycled plastics in 2021.^[1] In line with social expectations and marketing considerations,

European Union directives and regulations (e.g., 2019/904, 2018/852, and 2022/1616) regulate usable plastic packaging materials and their recycling rates.^[2,10–14]

The applicability of PET, in contrast to most thermoplastic polymers, is usually characterized by the intrinsic viscosity (IV) instead of the melt flow index (MFI).^[15] Since IV characterizes the chain length of the polymer, it can also be used to determine the degree of degradation (thermal, hydrolytic, thermomechanical) during processing. Many researchers have already analyzed the degree of degradation during recycling and reprocessing.^[16-19] In earlier studies, the two-phase (crystalline and amorphous) model was generally used, and embrittlement was explained by the higher crystalline fraction resulting from the decreasing average molecular weight.^[20,21] Mancini and Zanin showed that during five consecutive recycling cycles, the crystallinity of PET increased by 50%, and its impact strength decreased by 70%.^[22] Unintended crystallization is therefore a serious problem for the applicability of products made from recycled PET material, e.g., in the case of thermoformed products the drop-test performance significantly reduces.^[23]

In addition to the crystalline proportion, the mechanical properties of PET are significantly determined by the number of

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interlamellar links between them. In the case of high tiemolecule density, lamellar shear precedes the fracture of tie molecules, and crazes are formed, thereby enhancing the ductility of the polymer. However, if the tie-molecule density is low, these linking molecular chains will break under the concentrated load, whereby macroscopically a brittle fracture will occur. The tie-molecule density depends on the molecular weight of PET. In the case of high-molecular-weight chains, the number of interlamellar linkages is greater. The tie-molecule density is also influenced by the ratio and structure of the crystalline fraction, which can be affected by crystallization conditions.^[24]

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In recent studies, the microstructure of PET is evaluated by applying a three-phase morphological model, distinguishing rigid (RAF) and mobile (MAF) amorphous phases besides the crystalline fraction (CRF).^[25,26] RAF is generally located on the amorphous-crystalline interface in basal lamellar planes and consists of chain segments that are coupled with both the crystalline and mobile amorphous fractions. These segments located on the crystal boundaries are hindered in their movement, which can significantly affect the mechanical properties. The RAF acts as a stress transfer point and causes an increase in the elastic modulus, similarly to the behavior of the crystalline phase.^[27] On the other hand, since the major source of ductility is segmental mobility, the restricted mobility of the amorphous phase changes the internal crack formation and crack propagation, resulting in a decrease in toughness.

In order to improve product quality, the molecular weight of the recycled material can be increased with chain-extender additives^[28-30] or by solid state polymerization (SSP),^[31,32] thereby reducing the crystallization rate. Impact resistance can also be increased with elastomer additives.^[33-36] However, these approaches cause a significant cost increase during recycling. In the present research, a more straightforward method is proposed for improving the toughness of recycled PET products, namely the devitrification of the RAF by thermal annealing, which may result in enhanced molecular mobility in the amorphous region, thereby promoting ductile deformation.

According to the study of Righetti et al.,^[27] during the isothermal crystallization of PET conducted at temperatures (T_c) lower than 215 °C, in the range where thermoforming of PET is typically performed, partial vitrification of the amorphous chains that are arranged near the crystal lamellae occurs from the beginning of the process. It is also reported that during the heating of PET crystals, grown at T_c below 215 °C, the devitrification of RAF begins immediately above the glass transition temperature (T_{a}) . At heating rates higher than the recrystallization kinetics, even complete mobilization of RAF is possible at temperatures lower than 215 °C.^[27] In the former research studies, conducted towards the analysis of the vitrification and devitrification of the RAF in PET,^[37,38] however, the impact on the mechanical properties has not been or barely discussed, yet it could be particularly significant in recycled materials.

The present study aims to evaluate the effect of thermal annealing conditions, causing devitrification of the RAF, on the microstructure and macroscopic properties, i.e., impact resistance, of recycled PET films.

2. Experimental Section

2.1. Materials

rPET film with a nominal thickness of 500 µm, which was extruded from 100% post-industrial PET waste (flakes from PET trays), was purchased from Pro-Form Kft. (Ecser, Hungary). For the used rPET film, an IV of 0.61 ± 0.01 dL g⁻¹ was measured using an RPV-1 type (PSL Rheotek) viscosimeter in 60:40% phenol-1,1,2,2-tetrachloro-ethane solution with 0.5 g dL⁻¹ concentration at 30 °C. T_{α} of 75.2 °C, the temperature of cold-crystallization (T_{cc}) of 130.3 °C (onset of T_{cc} : 123.6 °C), and T_m of 248.4 °C were determined by differential scanning calorimetry (DSC) (using a Setaram DSC 131 EVO apparatus, evaluating the first heating ramp at 5 °C min⁻¹, in N₂ atmosphere).

3. Methods

3.1. Sample Preparation

rPET film samples of 100 mm × 100 mm dimension were thermally annealed between two pre-heated molds made of aluminium in a Memmert Une 200-type drying oven set to 90, 120, 150, and 180 °C, respectively. The duration of the thermal treatment was 0, 10, 20, and 30 s, respectively. Then, the samples were allowed to cool down to room temperature without any constraint.

3.2. Characterization Methods

Instrumented Falling Weight Impact Test

The instrumented falling weight impact (IFWI) tests were performed on an Instron Ceast Fractovis 9350 impact tester. The following parameters were applied for the tests: impact energy: 196.6 J, impact velocity: 4.43 m s⁻¹, drop height: 1000.6 mm, total mass: 20.04 kg. Square specimens with $80 \text{ mm} \times 80 \text{ mm} \times 0.5 \text{ mm}$ dimensions were used. The tests were conducted at room temperature and at 50% relative humidity. 12 specimens were tested for each annealed film sample. The perforation energy was calculated with Equation 1.

$$E_{\rm p} = \frac{E_{\rm max}}{h} \tag{1}$$

where $E_{\rm p}$ is the perforation energy [J mm⁻¹], $E_{\rm max}$ is the total impact energy [J], and *h* is the thickness of the specimen [mm].

Wide-Angle X-ray Diffraction

Wide-angle X-ray diffraction (WAXD) analyses were performed using an X'pert Pro MPD (PANalytical Bv., The Netherlands) diffractometer using Cu Ka radiation operating at 40 kV/30 mA and a Ni filter. The diffraction patterns were collected at 25 °C over an angular range (2 θ) of 5 to 55° with a step size of 0.02° per step and a dwell time of 7 s per increment.

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Figure 1. Typical force-time curves recorded during IFWI tests and photographs of the broken specimens.

Differential Scanning Calorimetry

A Setaram DSC 131 EVO device was applied for the DSC test in an inert atmosphere. The temperature changed from 30 °C to 320 °C with a 20 °C min⁻¹ heating rate. The ratio of the crystalline fraction was calculated by Equation 2.

$$CRF = \frac{\Delta H_{\rm m} - \Delta H_{\rm cc}}{\Delta H_{\rm m}^0} \cdot 100$$
⁽²⁾

where CRF is the degree of crystallinity [%], $\Delta H_{\rm m}$ is the melting enthalpy of the sample specified to weight [J g⁻¹], $\Delta H_{\rm cc}$ is the specific cold crystallization enthalpy [J g⁻¹] and $\Delta H_{\rm m}^0$ represents the specific melting enthalpy of the 100% crystalline PET, which is 140 J g⁻¹.^[26]

The ratio of mobile amorphous phase (MAF) was calculated from the change in specific heat capacity measured at the glass transition by applying Equation 3:

$$MAF = \frac{\Delta c_p}{\Delta c_p^0} \cdot 100$$
(3)

where MAF is the ratio of mobile amorphous fraction [%], Δc_p is the specific heat capacity change of the sample attributed to glass transition [J (g K)⁻¹] and Δc_p^0 is the specific heat capacity change of 100% amorphous PET at the glass transition [0.405 J (g K)⁻¹].^[26] The ratio of the rigid amorphous phase (RAF) [%] was determined following Equation 4:

$$RAF = 100 - MAF - CRF$$
(4)



Figure 2. Perforation energy of rPET film as a function of annealing conditions.

Dynamic Mechanical Analysis

The dynamic mechanical analysis (DMA) was performed on a MetraVIB DMA tester. The tension measurement was executed

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Figure 3. WAXD patterns of rPET films before and after annealed at a) 90 °C, b) 120 °C, c) 150 °C and d) 180 °C.

on 8 mm wide specimens and the distance of the jaws was 20.08 mm. The samples were heated from 10 to 140 °C at 3 °C min⁻¹. No static force was applied, and the frequency of the dynamic strain was 10 Hz.

Thermally Stimulated Depolarization Current

A Setaram TSC II (Setaram, France) apparatus was used for the thermally stimulated depolarization current (TSDC) measurements. TSDC technique is a very sensitive method for investigating amorphous phases. It is able to separate overlapping events that cannot be distinguished, e.g., based on DSC measurements.^[39] Thus, TSDC is also suitable for examining mobile and rigid amorphous phases. In the present research, a simple measurement mode was used, which allowed us to compare the mobile amorphous phase of the annealed samples and the rPET. The first step of the heating program was to set the temperature to 100 °C (T_p) where the sample was polarized with 300 V polarization voltage (U_p). After 5 min of polarization, the sample was cooled to 0 °C (T_0) with a cooling rate (r_c) of 10 °C min⁻¹ and kept there for 1 min. In the next step, the chamber was heated to 120 °C ($T_{\rm f}$) with a 5 °C min⁻¹ heating rate ($r_{\rm h}$). During this last step, the depolarization current was measured and depicted in the function of temperature. For the evaluation, the maximum peak intensity of the first peak (α_1), corresponding to the mobile amorphous phase, was determined. The changes in the mobile amorphous phase were calculated by dividing α_1 of the samples with the α_1 of rPET.

3.2.1. Results and Discussion

3.2.1.1. Instrumented Falling Weight Impact Test: The impact fracture behavior of the rPET films was characterized using IFWI tests. Typical force-time curves obtained from the tests and photographs of failure specimens are presented in **Figure 1**. The non-treated rPET film breaks in a brittle manner, as indicated by one sharp peak in the force-time curve. In the corresponding photograph of the broken specimen, many cracks radiating from the center, and one curved crack in a circumferential direction can be seen. Thermal annealing of the rPET films had a substantial effect on the evolution of the force-time curve and the failure behavior of the rPET films. The force-time curves recorded after

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Figure 4. Crystallinity of rPET film as a function of annealing conditions.

annealing the rPET films for 30 s at 150 and 180 °C are very similar in shape, but the peaks tend to be significantly smaller and narrower with the higher annealing temperature. These specimens fail in a more brittle fashion, the circumferential crack forms a big hole in the center of the specimen, which allows the striker to pass through completely. In contrast, the force-time curves of the rPET films annealed at 120 °C indicate ductile fracture behavior; these films yield long before the maximum of the force-time curve is reached. The tip of the striker completely penetrates the specimen, forming a circular hole, but in this case, also a characteristic flap is formed as the center part of the specimen deforms in association with much stress whitening. The flap region of these specimens is assumed to undergo permanent plastic deformation.^[40] During testing, the film specimens treated at 90 °C for 30 s showed slightly variable behavior. The force-time curves recorded for the majority of the 90 °C annealed film specimens were similar to those of the non-treated rPET films but typically higher values in peak force were detected. The corresponding similar fracture pattern (shown in Figure 1) also indicates a similar failure mode for these annealed specimens. However, the minority of the 90 °C treated test specimens showed typical ductile fracture behavior, which was also accompanied by flap formation, more similar to the characteristics of the films treated at 120 °C.

In **Figure 2**, the evaluation of the perforation energy values of the rPET films can be seen as a function of annealing temperature and time, respectively. The relatively high standard deviation of the measured values is due to the lower sensitivity of the method on thin samples and is also attributed to the greater inhomogeneity of the recycled raw material-based films than usual. The brittle fracture was characteristic of the untreated rPET film; an average perforation energy of 5.95 J mm⁻¹ was measured for this reference sample. As low as 10 s of heat treatment resulted in a noticeable change in fracture toughness and energy absorp-

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tion capability of the rPET film. As a result of annealing at 90 °C or 120 °C, the perforation energy of the rPET films increased, while heat treatment at 150 °C or 180 °C caused a drop in perforation energy. The perforation energy value of the untreated rPET film increased to its threefold, reaching a value higher than 18 J mm⁻¹, as a result of 10 s thermal treatment at 120 °C. Under these conditions, the sharp crack tip characteristic for the untreated rPET films changed to blunted (see Figure 1). Based on the data presented in Figure 2, increasing the annealing time from 10 to 30 s has a noticeable effect on the impact behavior and corresponding perforation energy values of the rPET films. In the case of treatment at 90 °C a slight increase of the perforation energy can be observed with the longer annealing time, while at 120 °C or higher set temperatures, the perforation energy values moderately decrease with longer annealing. Overall, it can be established that annealing below or above the onset temperature of cold-crystallization of rPET (123.6 °C as determined by DSC) has opposite effects on the mechanical performance of the rPET film. The changes in perforation energy absorbing mechanism occurring due to thermal annealing at different temperatures must be traced back to different structural changes of the polymer.

3.3. Wide-Angle X-Ray Diffraction

WAXD analysis was carried out to characterize the crystalline structure of the rPET film samples before and after thermal annealing at different temperatures. In **Figure 3**a,b, diffused halo patterns can be seen, indicating a high amorphous proportion for the untreated rPET film and also for the films annealed at 90 °C or 120 °C. In the 2D WAXD patterns corresponding to the rPET films treated at 150 °C (Figure 3c) and 180 °C (Figure 3d), however, peaks indicating noticeable crystallization^[41,42] appear, the intensity of which increases with annealing time.

3.4. Differential Scanning Calorimetry

The crystallinity ratio of the rPET films was calculated from the first heating curve of DSC analyses. The crystallinity values (Figure 4) determined by DSC are in good accordance with the WAXD experimental results (Figure 3). Namely, rPET films annealed at 90 °C or 120 °C remained mostly amorphous (i.e., CRF < 10.0%) even after 30 s of treatment. At the annealing temperature of 150 °C, the crystallinity of the rPET films noticeably increased after 10 s, from 5.9 to 13.4%, while after 30 s of annealing, the degree of crystallinity reached the value of 20.9%. At the higher annealing temperature of 180 °C, the crystallinity of the rPET film reached a value higher than 20.0% right after 10 s of treatment and did not change significantly after the longer time period. The increased brittleness, as found by IFWI tests (Figures 1 and 2) is certainly related to the notable increase in the crystalline ratio of rPET samples after annealing at temperatures higher than T_{cc} . However, the increase in perforation energy and ductility of the rPET films when annealed at temperatures below T_{cc} is apparently not associated with their degree of crystallinity. Therefore, the changes in the amorphous region of the rPET films during annealing were also investigated based on the three-phase model.^[43]



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Figure 5. The ratio of a) mobile amorphous phase and b) rigid amorphous phase of rPET film as a function of annealing conditions.



Figure 6. Storage modulus of rPET film as measured a) at 30 °C and b) at 90 °C as a function of annealing conditions.

The changes in MAF and RAF as functions of annealing temperature and time are presented in Figures 5a,b, respectively. In the case of the films treated above T_{cc} , the increase in RAF (Figure 5b) follows a similar trend as that of the crystallinity (Figure 4). Apparently, treatment at 90 °C did not cause remarkable change neither in the crystalline nor in the amorphous phase; only a slight increase of the MAF can be observed after 30 s of annealing. It is assumed that the improvement in the energy absorption capability of the rPET films treated at 90 °C is not primarily caused by the RAF/MAF phase transition but is rather attributed to the strain relaxation of the molecular segments in the amorphous region.^[44] Nevertheless, annealing at 120 °C resulted in a noticeable increase of MAF from 69.1% to 89.0% and 86.4% after 10 s and 30 s, respectively (Figure 6a). In parallel, at this temperature of annealing, the ratio of RAF dropped from 25.0% to below 5.0% (Figure 5b), even though concurrently the crystalline ratio also slightly increased (Figure 4). This observation is in accordance with the findings of Righetti et al.^[37], i.e., structural relaxation of the RAF takes place near the crystallization temperature and thereby with concurrent development of the crystalline phase.

It was concluded that at 120 °C, a transition of the RAF to MAF occurs. Such devitrification of the amorphous fraction is proposed to be strongly related to the noticeable increase of the impact resistance, as observed by IFWI tests (Figures 1 and 2). In order to strengthen this assumption, i.e., the correlation of the RAF/MAF transition with the impact-damping properties, further analyses were performed to characterize the changes in the amorphous region.

3.5. Dynamic Mechanical Analysis

DMA measurements were performed to investigate the annealing-related changes in the stiffness and indirectly also on the phase structure of the rPET films. Graphs showing the evolution of storage modulus and tan delta as a function of temperature after annealing the rPET films at different temperatures are given as Supporting Information. Storage modulus



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values obtained at 30 °C and at 90 °C are plotted in Figure 6a,b, respectively. It can be seen in Figure 6a, that when exposed to an annealing temperature of 180 °C, the stiffness of the rPET material, as measured at 30 °C, monotonically increases with annealing time. The increasing stiffness is certainly related to the rapidly increasing crystallinity (see Figure 4) of the rPET polymer during annealing at 180 °C. Also, the RAF has a similar behavior to that of the crystalline phase and acts as stress transfer, producing an increase in the elastic modulus.^[27] At lower annealing temperatures of 150 °C, 120 °C, and 90 °C, however, after 10 s of annealing, a decrease of the storage modulus can be observed compared to the initial value (Figure 6a). At annealing temperatures close to or higher than T_{cc} (at 120 °C and 150 °C, respectively), after a longer annealing time of 30 s, the storage modulus values start to increase again. This phenomenon indicates structural change of the rPET material during the time period of annealing at 120 °C and 150 °C and is in accordance with the changes in the phase structure as proposed based on the DSC results. Namely, during the first period of the treatment (10 s), the devitrification of RAF to MAF occurring around 120 °C dominates, which is accompanied by a decrease in stiffness. In parallel, since at this temperature crystallization also occurs at a slower rate, after 30 s the increase in crystallinity already reaches a level that results in an increase of the inherent storage modulus. Similar annealing time-dependent behavior was described for PET/elastomer blends during post-crystallization at 150 °C.^[35]

As can be seen in Figure 6b, noticeably increased storage modulus values were measured at 90 °C for the rPET samples after 10 or 30 s annealing at 180 °C, and after 30 s annealing at 150 °C as well. The increased stiffness at elevated temperatures is generally associated with increased crystalline ratio and renders these materials suitable for use at elevated service temperatures compared to the amorphous polymer form.^[45,46] The experimental results presented in Figure 6b are in good agreement with the outcomes of Slezák et al.,^[35] who revealed that a noticeable improvement in the high-temperature storage modulus (measured at 90 °C) occurs when a crystalline ratio of approx. 18% is exceeded. At lower annealing temperatures of 90 °C or 120 °C, such an effect on the high-temperature modulus could not be detected (Figure 6b), as we did not see any indication of a noticeable change in the crystalline structure either measured by WAXD (Figure 3) or DSC (Figure 4).

Figure 7 presents the area of the $tan\delta$ peak, obtained by integration of the loss factor curve between 60 and 170 °C, as a function of annealing time at different temperatures. Based on the results presented in Figure 7, at annealing temperatures of 150 °C and 180 °C, reduction of the bulk amorphous structure can be suggested, which is in accordance with the increased crystallinity proportion of these rPET films (see also in Figures 2 and 3), and as proposed based on the increased storage modulus values as well. When a lower annealing temperature of 120 °C was applied, even a slight increase in the loss factor peak area was measured. This observation also indicates the increase of the MAF,^[47,48] in accordance with the DSC results (Figure 5a). 90 °C as annealing temperature had only negligible effect on the area of the tan δ peak, which confirms the assumption that not phase structural changes but rather strain relaxation effect explains the improved impact behavior of the rPET films after heat treated at 90 °C.



Figure 7. Peak area of the loss factor curve as a function of annealing conditions.

3.6. Thermally Stimulated Depolarization Current

In the TSDC curves of the rPET films, two current intensities were identified (**Figure 8a**); α_1 corresponding to the mobile amorphous fraction and α_2 corresponding to the rigid amorphous phase.^[49]

In Figure 8b, the evolution of the α_1 peak intensity can be seen as a function of annealing conditions. The higher is this intensity the higher is the ratio of MAF in the rPET film specimen.^[50] According to the experimental results presented in Figure 8b, the annealing temperature of 90 °C does not have a significant effect on the structure of the amorphous fraction; only a slight increment of the MAF can be assumed after 30 s of treatment. This small change in the amorphous region is probably of little relevance to the impact energy absorbing capability of the rPET films, which after annealed at 90 °C, showed noticeable ductile behavior (see Figure 1). The thermal treatment above T_{cc} (as measured at 150 °C and 180 °C) resulted in a decreased ratio of MAF (decreased α_1 peak intensities) that is certainly associated with rapid crack propagation and thus increased brittleness of the rPET films, as also evidenced by IFWI tests (Figures 1 and 2). In contrast, the increased α_1 peak intensity detected for the rPET film after thermally annealed at 120 °C (Figure 8b) further confirms that annealing at 120 °C allows noticeable RAF to MAF devitrification, and consequently, a significant toughening effect can be exerted, as revealed by IFWI test results (Figures 1 and 2).

4. Conclusions

This research study draws attention to the significant effect of RAF on the impact resistance of recycled PET films, which are particularly prone to embrittlement. The segments of RAF chains are located at the boundary of crystallites, which have increased



Figure 8. a) TSDC curve of untreated rPET film and b) α_1 peak intensity as a function of annealing conditions.

growing rates in PET recyclates of shorter molecular chains. The RAF segments are hindered in their movement, acting as stress concentration sites and decreasing toughness.

This work shows that thermal annealing close to $T_{\rm cc}$ (at 120 °C) is an effective way to devitrificate RAF and increase segmental mobility of the amorphous phase, which is the primary source of ductility. The RAF to MAF relaxation occurring at 120 °C was evinced by DSC, DMA, and TSDC methods. As a result, the brittle failure, characteristic of the non-treated rPET film, changes to ductile fracture behavior accompanied by outstanding energy absorption capability.

IFWI tests of rPET films showed improved toughness after treatment at 90 °C (i.e., between $T_{\rm g}$ and $T_{\rm cc}$), even though noticeable RAF to MAF devitrification could not be detected in this case. Therefore, the increased fracture toughness of rPET films annealed at 90 °C was mainly attributed to strain relaxation in the amorphous region. It was also found that at annealing temperatures above $T_{\rm cc}$, crystallization and devitrification simultaneously start to occur, but with increasing annealing time crystallization overcomes molecular relaxation, which increases the rigidity and brittleness of the rPET films.

Thermal annealing, performed close to the onset temperature of cold-crystallization, is proposed as a simple and cost-effective way to compensate for the deterioration of the mechanical properties of recycled PET films and thus to expand the market potential of extruded rPET products. This approach can also indirectly contribute to an increase in the recycling rate of PET waste.

Supporting Information

Supporting Information is available from the Wiley Online Library or from the author.

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Conflict of Interest

The authors declare no conflict of interest.

Author Contributions

Conceptualization: F.R., G.M., and K.B.; Methodology: E.S., E.Sz., and D.G.; Investigation: E.S., E.Sz., and D.G.; Evaluation: E.S., E.Sz., D.G., F.R., and K.B.; Visualization: E.S., E.Sz., D.G., F.R., and K.B.; Writing, E.S., D.G., F.R., G.M. and K.B. All authors reviewed the manuscript.

Data Availability Statement

The data that support the findings of this study are available from the corresponding author upon reasonable request.

Keywords

impact resistance, recycled PET film, thermal annealing, three-phase model

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