

Structure and Properties of β -Polypropylene Reinforced by Polypropylene Fiber and Polyamide Fiber*

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Abstract Matrix/fiber composites of β -form isotactic polypropylene (*i*PP) matrix and α -*i*PP or PA6 fibers were prepared by laminating technique under different preparation temperatures. The mechanical properties and interfacial morphologies of these composites were studied by tensile test, optical microscopy and scanning electron microscopy, respectively. The experimental results show that the tensile yield load and tensile modulus of β -*i*PP/PA6 matrix/fiber systems increased significantly at the expense of elongation at break. These mechanical properties show essentially no dependence on the sample preparation temperature. On the other hand, the mechanical properties of *i*PP matrix/fiber single polymer composites depend strongly on the sample preparation temperature. At low sample preparation temperature, *e.g.*, 172 °C, the solid α -*i*PP fiber induces α -*i*PP crystallization, leading to the formation of α -*i*PP transcrystalline layer around the fiber. This results in a remarkable increment of the tensile yield load and tensile modulus. The elongation at break is also much better than that of the *i*PP/PA6 matrix/fiber system. It reflects a better interfacial adhesion of the single polymer composite compared with the *i*PP/PA6 composite. At higher sample preparation temperature, *e.g.*, 174 °C or 176 °C, the partial surface melting of the oriented fiber allows interdiffusion of *i*PP molecular chains in the molten fiber and matrix melt. The penetration of matrix chains into the molten *i*PP fiber results in some *i*PP molecular chains being included partially in the recrystallized fiber and the induced β -transcrystalline layers. This kind of configuration leads to an improvement of interfacial adhesion between the fiber and matrix, which causes a simultaneous increase of the tensile yield load, tensile modulus and elongation at break of β -*i*PP.

Keywords: Polypropylene; Composite; Mechanical property; Interfacial morphology.

INTRODUCTION

Isotactic polypropylene (*i*PP) is a typical polymorphic polymer with at least four basic crystalline modifications (including the monoclinic α -form, trigonal β -form and orthorhombic γ -form) and a mesomorphic smectic form^[1–9]. Its α form is thermodynamically most stable and can be obtained simply through melt crystallization. The β -phase is metastable and can only be obtained through some kinds of specific procedures, such as crystallization in thermal gradient^[10, 11], crystallization of the sheared or elongated melts^[12–18] or crystallization at the presence of nucleation agents^[19–30]. Among these methods, the addition of β -nucleating agent is the most accessible and effective method to obtain high β -*i*PP content material^[31]. The β -*i*PP exhibits some excellent performance characteristics such as the improved elongation at break, impact strength. Another most intriguing property of the β -*i*PP rests on the ability to produce microporous thin films through plastic deformation caused

* This work was financially supported by the National Natural Science Foundation of China (Nos. 50833006, 21274009 and 51221002).

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Received October 13, 2013; Revised December 13, 2013; Accepted December 24, 2013

doi: 10.1007/s10118-014-1417-x

by volume contraction during β to α transformation^[32–34]. Therefore, it has also received much attention from scientific research as well as industrial applications^[6, 12, 13, 35, 36].

It should be pointed out that, except for ductility and impact strength requirement, desired modulus and tensile strength are also necessary in some practical applications. For thermoplastic polymeric materials, their composites with various fibers provide an efficient way for improving the tensile strength and modulus. Consequently, the thermoplastic matrix/fiber composites have been paid more and more attention^[37–40]. The properties of the fiber/matrix composites depend essentially on the used materials. There are, however, other factors that influence the properties of the composites remarkably^[41, 42]. For example, the fiber induced interfacial morphology of the matrix will affect the stress transfer from the matrix to the fiber and influence the final properties of the composites^[43]. Also a better fiber/matrix interfacial adhesion can strengthen the effect of fibers^[43–49]. This leads to extensive studies on the fiber/matrix interfacial adhesion. To enhance the fiber/matrix interfacial adhesion, polar functional groups are most frequently introduced into the composites to improve the compatibility between the fiber and matrix. Actually, improvement of interfacial bonding is highly expected for self-reinforced single polymer fiber/matrix composites. This kind of composite exhibits also ideal recyclability since it is composed of same polymeric materials. Therefore, there is an increasing interest in the field of self-reinforcement. Up to now, a number of investigations have been conducted on such self-reinforced polymer composites, which include polyethylene fiber reinforced polyethylene^[50–56] and polypropylene fiber reinforced polypropylene composites^[57]. While the polyethylene single-polymer composite has prepared *via* specific processing, the *i*PP homogenous composite was prepared by compression-molding the non-oriented *i*PP thin films of *ca.* 80 μm in thickness together with the highly oriented *i*PP fibers, which were end-fixed on a glass slide, at 170 °C for 5 min. It was confirmed that the used procedure has molten only the *i*PP matrix material but not the constrained fibers. In this case, α -form *i*PP transcrystalline layers in the vicinity of *i*PP fibers have been produced. The prepared single *i*PP composite shows improved mechanical properties. In our previous work, *i*PP single polymer composites with β -form *i*PP transcrystalline layers were prepared through partial surface melting of the *i*PP fibers^[58–62]. In this configuration, the non-molten *i*PP fibers can reinforce the generated β -form *i*PP matrix. Moreover, it was confirmed that *i*PP chains in the matrix must interdiffuse with oriented chains of partially molten fiber to generate β -*i*PP nuclei^[63]. Taking into account that some of the molecular chains will be included partially in the fiber and partially in the matrix, excellent interfacial bonding and therefore superior mechanical properties are expected for the thus prepared single polymer composites.

In the present work, single polymer composites of β -*i*PP reinforced with α -*i*PP fibers were prepared. The mechanical properties of the composites at different preparation temperatures were investigated and compared with the *i*PP/PA6 matrix/fiber composites. The results reveal an effective reinforcement of the *i*PP single polymer composites due to the synergetic effect.

EXPERIMENTAL

Materials and Sample Preparation

The isotactic polypropylene (*i*PP) matrix material employed in this work was a commercial grade *i*PP, GB-2401, with melt flow index of 2.5 g/10 min, $M_w = 437900$, and melting temperature of 170 °C, produced by Yanshan Petroleum and Chemical Corp., China. For producing β -*i*PP matrix, β -nucleation agent was used which has the trade name of WBG-II produced by Guang Dong Weilinna Functional Materials Co., LTD. The mixing of nucleating agent (NA) and *i*PP matrix was carried out by a home-made blender. The β -*i*PP thin films of *ca.* 0.2 mm in thickness were prepared by compression-molding the *i*PP/NA mixture at 200 °C under 75 kg/cm². The obtained β -*i*PP films have a melting temperature of about 154 °C as measured by DSC.

*i*PP fibers with a diameter of about 18 μm were produced using the same GB-2401 *i*PP with a homemade melt-spinning device. Spinning temperature was 200–240 °C. Resultant *i*PP fibers were then subjected to a cold-draw procedure at 110 °C with a draw ratio of 4. Polyamide (PA6) fibers used in this study, having a diameter of about 25 μm , were kindly supplied by Tianjin Polytechnic University. To remove the sizing agents

on the fiber surfaces, the fibers were rinsed for 4 h in a distillation flask with acetone and then dried in a vacuum oven at 40 °C for 5 h.

The β -iPP matrix/fiber composites either with iPP fibers in α -form or with PA6 fibers were prepared by laminating method. iPP or PA6 individual fibers were tightly fixed on a metal frame to form fiber felts. The distance between the fibers in the felts was estimated to be *ca.* 150 μ m. The fiber felts were then placed between two β -iPP films to form a matrix/fiber/matrix sandwich structure. The sandwich samples were heated to a temperature above the melting point of β -iPP matrix for 15 min and cooled to room temperatures. For comparison under exactly same sample thermal histories, pure β -iPP samples without fibers have also subjected the same procedures as the fiber/matrix composites. The specimens for tensile test were cut into dumbbell-shaped bars. The tensile testing specimen had the dimension of 30 mm $\times$ 4 mm $\times$ 0.2 mm.

Characterization

Standard tensile test was performed using an America Instron 1185 tensile testing machine at a constant cross-speed of 100 mm/min. All mechanical properties were measured under room temperature at *ca.* 25 °C and the average value was obtained from over five specimens for each condition.

For optical microscopy observation, an Axioskop 40 A Pol Optical Microscope (Carl Zeiss) was used. All optical micrographs presented in this paper were taken under crossed polarizers.

A JSM-6701F field emission scanning electron microscope was used to study the interfacial morphology of the composites. The interfacial adhesion between fiber-matrix was evaluated by the fracture surface morphologies of the composites. Prior to scanning electron microscopy (SEM) observations, all fracture surfaces were sputter-coated with gold, and then observed under an acceleration voltage of 5 kV.

RESULTS AND DISCUSSION

Mechanical Properties of the β -iPP Laminate and its Composites

The deformation behavior of the iPP and PA6 fibers was first checked. Representative stress-strain curves of α -iPP and PA6 fibers are presented in Fig. 1. One can clearly see that except for a slight lower toughness, the PA6 fibers exhibit a much better mechanical properties than the α -iPP fibers. The elongations at break are 90% for PA6 fibers and 110% for α -iPP fibers. The tensile strengths at break are about 310 MPa and 168 MPa for PA6 and α -iPP fibers, respectively. The tensile moduli of the PA6 and α -iPP fibers are 2019 MPa and 1355 MPa. According to these results, a better reinforcement effect of the PA6 fiber toward β -iPP matrix may be expected. Therefore, the mechanical deformation behavior of the composites composed of β -iPP matrix and PA6 or α -iPP fibers is compared.

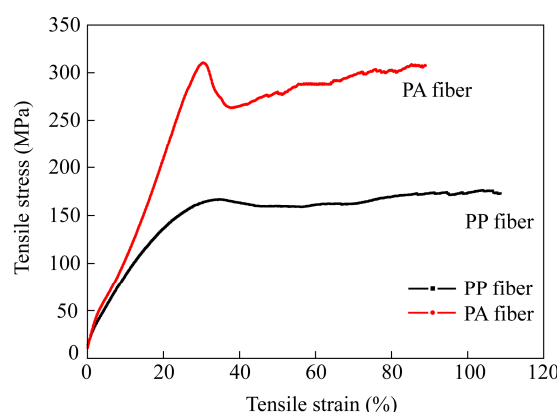


Fig. 1 The tensile stress-strain curves of iPP and PA fibers (a bundle of fibers)

Figure 2 shows the stress-strain curves of the matrix/fiber composites with β -iPP matrix and PA6 or α -iPP fibers prepared at different compression temperatures. For a direct comparison, the stress-strain curves of the

β -iPP matrixes subjected the same preparation procedures as the matrix/fiber systems are also presented in Fig. 2. The characteristic data obtained from the mechanical tests on various composites are summarized in Table 1. We can see that the mechanical properties of the *i*PP single polymer composites depend strongly on the preparation temperature, while those for the β -iPP matrix and *i*PP/PA6 matrix/fiber system changed slightly. For the samples compressed at 169 °C, see Fig. 2(a), increase in tensile strength is evident for the composites either with *i*PP or with PA6 fibers. For example, the tensile modulus of *i*PP single composite changed from 531 MPa for β -iPP laminate to 992 MPa. An even higher increment is seen for the *i*PP/PA6 matrix/fiber composite, e.g., increased from 531 MPa to 1115 MPa. This rests on a higher strength of PA6 fibers compared to the α -iPP fibers. On the other hand, the elongation at break of the matrix/fiber systems decreased remarkably. The elongation at break of the *i*PP single polymer composite is only about 25% of the β -iPP laminate, while that of the *i*PP/PA6 matrix/fiber composite shows a decrease of more than 90% compared to the β -iPP laminate. Now the *i*PP/PA6 composite has an elongation at break of only 60%. It is even lower than the PA6 fibers. This is normal since the increase in tensile modulus is mostly at the expense of the toughness. When the sandwiched laminates were compressed at 171 °C, see Fig. 2(b), only small changes in the mechanical properties of the β -iPP matrix and *i*PP/PA6 matrix/fiber system were observed. However, for the *i*PP single polymer composite, the elongation at break has been increased intensely from 151% to 542% with the tensile modulus reduced from 992 MPa to 704 MPa. Nevertheless, it is still much higher than the β -iPP laminate (704 MPa versus 549 MPa). With further increase of compression temperature to 173 °C, both the elongation at break and tensile modulus of β -iPP laminate cut down a little bit, reflecting the influence of thermal history on the mechanical properties of the polymeric materials. The *i*PP/PA6 composite exhibits still the highest strength among the studied systems,

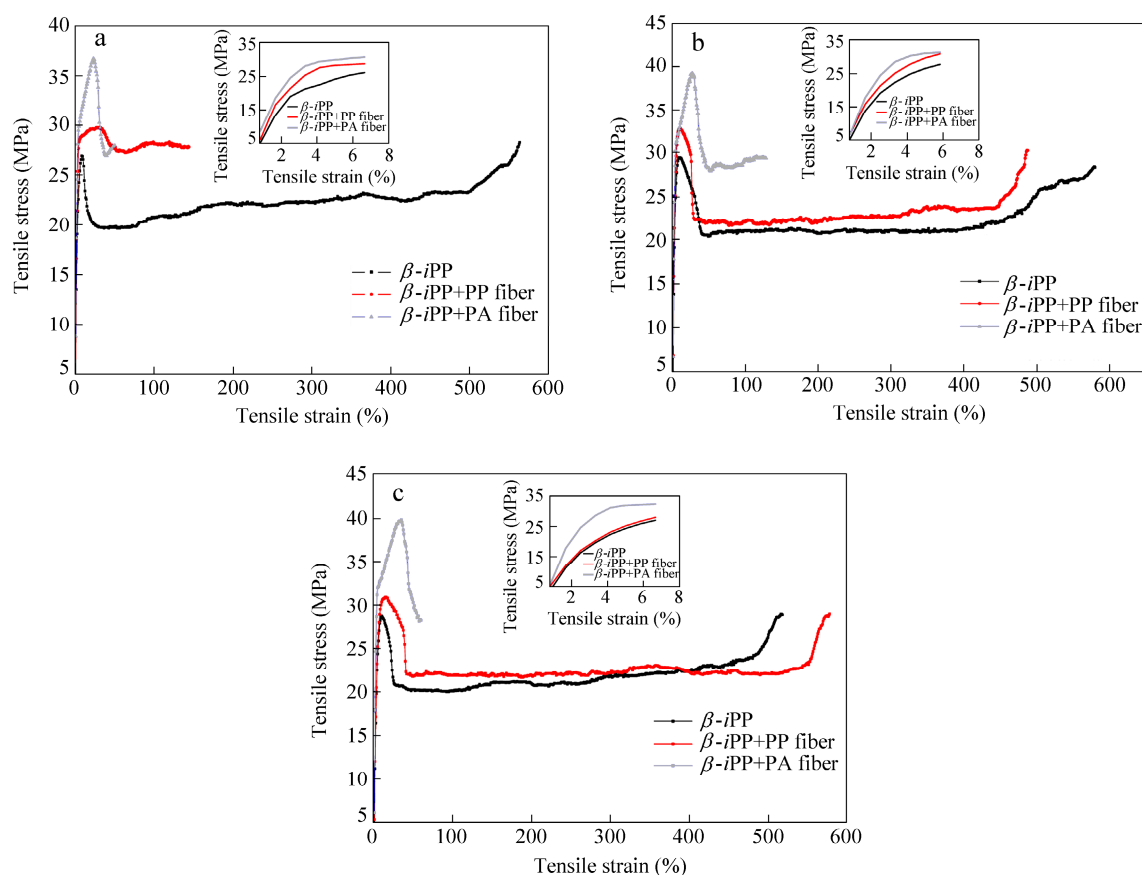


Fig. 2 The tensile stress-strain curves of β -iPP and its matrix/fiber composites prepared at (a) 169 °C, (b) 171 °C and (c) 173 °C (The enlarged insets show the linear region of the stress-strain curves.)

even though abated slightly. At the same time, an improvement in elongation at break was seen. However, it is still poorer than that of the PA6 fibers. The most intriguing result is found for the *i*PP single polymer composite. One can see that the elongation at break and tensile modulus are improved simultaneously with respect to the β -*i*PP laminate.

Table 1. Characteristic data of the β -*i*PP laminate and its composites obtained from tensile tests

Samples	Preparation temperature	Properties		
		Tensile yield load (N)	Elongation at break (%)	Tensile modulus of elasticity (MPa)
β - <i>i</i> PP	169 °C	21.73 ± 1.2	621.17 ± 61.0	531.41 ± 51.1
	171 °C	22.51 ± 1.9	566.15 ± 15.1	549.08 ± 46.5
	173 °C	20.19 ± 1.2	511.00 ± 7.3	517.13 ± 39.4
β - <i>i</i> PP + <i>i</i> PP fiber	169 °C	25.77 ± 1.8	151.96 ± 10.7	992.08 ± 42.8
	171 °C	27.82 ± 0.42	542.26 ± 46.0	704.70 ± 2.2
	173 °C	27.64 ± 0.7	582.82 ± 6.3	535.12 ± 50.2
β - <i>i</i> PP + PA fiber	169 °C	31.49 ± 1.2	61.95 ± 12.1	1157.77 ± 86.7
	171 °C	32.13 ± 2.2	65.11 ± 19.2	1052.96 ± 127.9
	173 °C	32.57 ± 0.78	69.24 ± 8.3	1101.73 ± 62.9

Interfacial Structures of the β -*i*PP/Fiber Composites

From the above discussion, it is clear that the influence of *i*PP and PA6 fibers on the property of β -*i*PP laminates is different. It is well known that the mechanical properties of fiber reinforced polymer composites are greatly related to the fiber-polymer interfacial adhesion and interfacial morphology of the composites^[45, 64, 65]. Taking these into account, the different tensile behavior of the *i*PP/PA6 composites with respect to the *i*PP single polymer composites should be mainly related to the different fiber-polymer interfacial adhesion. On the other hand, the different tensile behavior of *i*PP single polymer composites prepared at different temperatures may be related to the different interfacial morphologies. In order to illustrate these differences, interfacial morphologies of the composites were studied.

Figure 3 displays the optical micrographs of the *i*PP single composites prepared at various temperatures. For the samples compressed at 169 °C, see Fig. 3(a), the *i*PP fiber can be easily recognized. There is a distinct boundary between the *i*PP matrix and its same polymer fiber. This indicates that the α -*i*PP fiber is not molten. The solid *i*PP fiber initiates the transcrystallization of the melt *i*PP matrix and generates thin transcrystalline layers surround the fiber. This implies the high nucleation ability of the *i*PP fiber toward its *i*PP polymer matrix. In the areas away from the *i*PP fiber, instead of spherulitic morphology, the *i*PP grows in microcrystallite form, indicating the high nucleation efficiency of the used nucleation agent. Selective melting at 158 °C demonstrates that the *i*PP matrix is composed of β -*i*PP crystals. On the contrary, the transcrystalline layers consist mainly of α -*i*PP crystals. It is acceptable under the consideration that α -*i*PP fiber causes the homoepitaxial crystallization of *i*PP melt in its α -form. There are also some fan shaped β -*i*PP crystals in the transcrystalline layers. These fan shaped β -*i*PP crystals can be triggered either by the nucleation agent or by the fiber if partial melting happened in some places. The exact origin of the formation of fan shaped β -*i*PP crystals cannot be given at moment. For the samples compressed at 171 °C, see Fig. 3(b), the location of the *i*PP fiber can still be identified. However, the outline of the *i*PP fiber cannot be distinguished completely. This may indicate the occurrence of melting or at least partial melting of *i*PP fiber. In this case, the whole sample including the interfacial layer is composed of fully β -*i*PP crystals. Similar morphology was observed for the sample compressed at 173 °C, see Fig. 3(c). The formation of β -*i*PP interfacial layer is in good accordance with our previous studies, which indicate that melting or partial melting of the *i*PP fiber encourages the crystallization of *i*PP in its β -form^[59, 61–63].

Figure 4 shows the optical micrographs of the β -*i*PP/PA6 matrix/fiber composites prepared at different temperatures. Since the PA6 fibers cannot be molten at all of the chosen temperatures, the PA6 fibers can be clearly visualized in each of the optical micrographs. As in the case of *i*PP single polymer composites, the existence of β -*i*PP nucleation agent leads to formation of *i*PP microcrystallites. In the vicinity of the PA6 fiber,

no evident transcrystalline layers can be observed. This originates from the poorer nucleation ability of PA6 fiber toward *i*PP compared with the used nucleation agent. Selective melting of the samples at 158 °C demonstrates that, in the *i*PP/PA6 matrix/fiber systems, the *i*PP crystallizes essentially in its β -form, indicating again a higher nucleation ability of the used β -*i*PP nucleation agent.

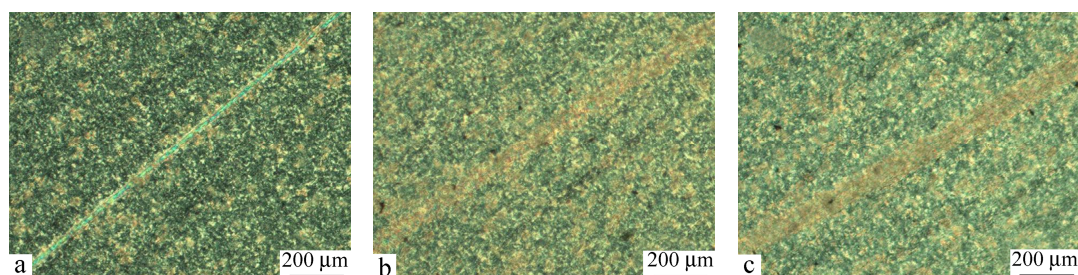


Fig. 3 Optical micrographs of *i*PP single polymer composites prepared by compression molding at (a) 169 °C, (b) 171 °C and (c) 173 °C and then cooled to room temperature

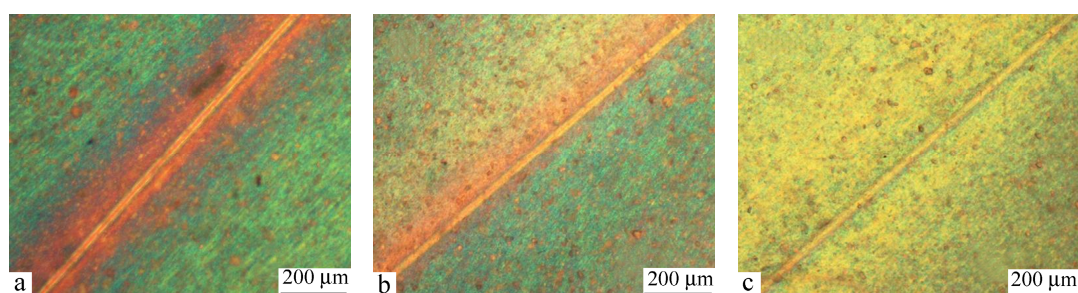


Fig. 4 Optical micrographs of the β -*i*PP/PA6 matrix/fiber composites prepared at (a) 169 °C, (b) 171 °C and (c) 173 °C and then cooled to room temperature

The above experimental results illustrate the different crystallization behavior of *i*PP around the *i*PP and PA6 fibers. Due to the lower nucleation ability of PA6 fiber toward *i*PP matrix with respect to the used nucleation agent, the PA6 fibers have no impact on the crystallization of the *i*PP matrix. Therefore, the *i*PP in the *i*PP/PA6 composites crystallizes in its β -form under all preparation conditions used. The crystallization of *i*PP in its single polymer composites depends strongly on the preparation temperature. At relatively lower preparation temperature, *e.g.* 169 °C, the α -*i*PP fiber kept in solid form. The solid α -*i*PP shows higher nucleation ability toward its *i*PP matrix than the used nucleation agent and therefore induces transcrystallization of the *i*PP matrix. The higher nucleation ability of α -*i*PP fiber on its *i*PP matrix relies on the same chemical structure and perfect crystallographic matching, which enables the occurrence of homoepitaxy. As a result, instead of β -*i*PP crystals induced by the nucleation agent, α -*i*PP transcrystals are produced by the α -*i*PP fiber. At elevated preparation temperatures, *e.g.* 171 °C or 173 °C, melting or at least surface melting of the α -*i*PP fiber takes place. However, a complete relaxation of the highly oriented *i*PP molecular chain in the fiber has not been reached. When crystallization later ensued, the oriented molecular chains previously included in the fiber are able to transform into crystalline self-nuclei and initiate the crystallization of the *i*PP matrix. Taking the fact that oriented *i*PP melt encourages the formation of its β -form crystals into account, transcrystalline layer of β -*i*PP crystals surrounding the *i*PP fiber was obtained. These different interfacial morphologies provide us some clues regarding the different tensile behavior of the composites.

For the *i*PP/PA6 systems, the β -*i*PP crystals are grown from the nucleation agents. The growth of β -*i*PP crystals in the vicinity of PA6 fiber will be stopped when impinged with the fiber. Under such circumstance, weak interfacial adhesion between the PA6 fiber and β -*i*PP matrix is expected. It is the weak interfacial adhesion that gives rise to the underperformance in reinforcing the β -*i*PP matrix. This is supported by the scanning

electron microscopy (SEM) observation on the fracture surface morphologies of the samples. Figure 5 shows the SEM micrographs of the fracture surfaces of the PA6/*i*PP fiber/matrix composites. For the sample compression-molded at 169 °C and 171 °C, see Figs. 5(a) and 5(b), the PA6 fibers have been pulled out from the composites with fairly clean surface. There exist also interspaces between the fiber and matrix. This indicates that the combination of PA6 fiber with the *i*PP matrix is indeed much poor. For the sample compression-molded at 173 °C, even though the PA6 fiber has not been pulled out, it breaks away from the matrix. This may imply an improved surface bonding between the PA6 fiber and the *i*PP matrix. It is nevertheless still not much strong. As a result, similar mechanical behavior of the PA6/*i*PP fiber/matrix systems prepared at different temperatures is observed.

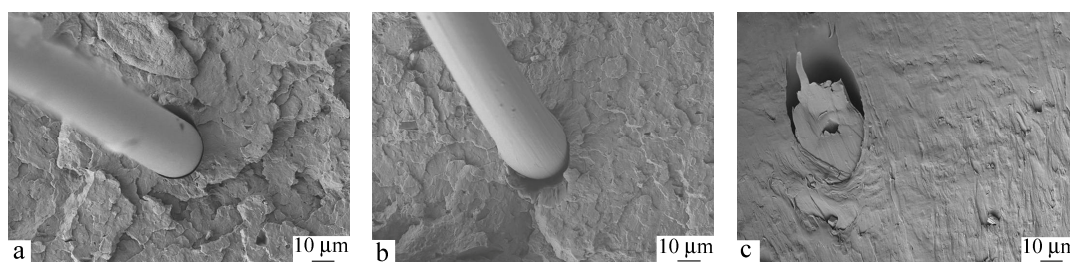


Fig. 5 SEM micrographs showing the fracture surface morphologies of the PA6/*i*PP fiber/matrix composites prepared at (a) 169 °C, (b) 171 °C and (c) 173 °C and then cooled to room temperature

For the *i*PP single polymer composites, taking the same *i*PP materials in each part of the composite into account, the connection between the *i*PP fiber and matrix is expected to be better than that of the PA6/*i*PP sample. The different mechanical properties should be related to the different interfacial morphologies caused by the character of the *i*PP fibers during sample preparation. This is supported by the SEM observation on the fractural surface morphology as shown in Fig. 6. It can be seen that the *i*PP fiber in the sample compression molded at 169 °C is also pulled out from the matrix, see Fig. 6(a). However, the *i*PP fiber is well connected with the *i*PP matrix. No interspace has been created during tensile test. Moreover, the surface of the *i*PP fiber is not as smooth as that in the *i*PP/PA6 matrix/fiber systems. This indicates an improved interfacial adhesion between the *i*PP matrix and fiber, which leads to an increment of tensile modulus from 533 MPa for β -*i*PP laminate to 943 MPa for the single *i*PP composite. At the same time, the elongation at break of single *i*PP composite is evidently higher than that of the PA6/*i*PP composite (144% versus 50%). These results can be explained in following way. At 169 °C, the *i*PP fiber was not molten. The α -*i*PP fiber induces therefore the α -*i*PP transcrystallization of the supercooled *i*PP matrix in the vicinity of the fiber, as illustrated in Fig. 7(a). On the other hand, the used nucleation agent encourages the β -*i*PP crystallization of the matrix away from the fiber. In this kind configuration, the α -*i*PP fiber together with the transcrystalline α -*i*PP layers are responsible for the enhanced Young's modulus and tensile strength, while the β -*i*PP matrix will response for the ductility. When the sample was compression molded at temperatures higher than 171 °C, the *i*PP fibers are not pulled out from the matrix, see Figs. 6(b) and 6(c). This implies an even better interfacial adhesion between the *i*PP fiber and matrix. The improved interracial adhesion should be related to the partial surface melting of the *i*PP fibers, as demonstrated by the formation of β -*i*PP transcrystalline layers surrounding the *i*PP fibers. In a previous work, it is suggested that the matrix chains are involved in the β -nucleation event and that partial surface melting of the oriented fiber is essential to the penetration of matrix chains into the oriented material prior to nucleation. In this case, some *i*PP molecular chains will be included partially in the recrystallized fiber and its induced β -transcrystalline layers, as illustrated in Fig. 7(b). It is these penetrating molecular chains that give rise to the excellent interfacial connection between the *i*PP fiber and matrix. Consequently, the elongation at break of the *i*PP single polymer composite has been increased remarkably, *e.g.*, from 144% to 487% for samples made at 171 °C. Meanwhile, the surface melting of the fiber results in an attenuation of the α -*i*PP fiber and causes a

decrease in tensile modulus. Therefore, a greater degree surface melting at 173 °C brings about a further decrease in tensile modulus. Nevertheless, the tensile yield load, the tensile modulus and the elongation at break of the single *i*PP composite prepared at 173 °C increased simultaneously compared to the β -*i*PP laminate, indicating an efficient modification of the β -*i*PP.

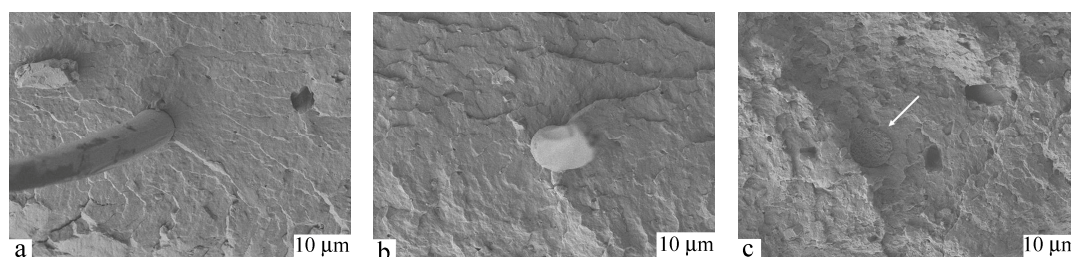


Fig. 6 SEM micrographs showing the fracture surface morphologies of the *i*PP single polymer composites prepared at (a) 169 °C, (b) 171 °C and (c) 173 °C and then cooled to room temperature

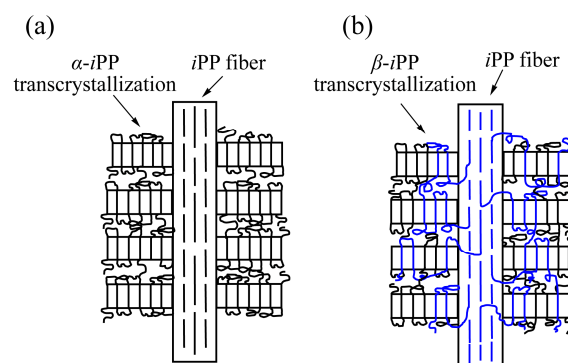


Fig. 7 Schematic representation of transcrystallization layers in the *i*PP fiber/matrix single polymer composites prepared at (a) 169 °C and (b) 171 °C

CONCLUSIONS

Matrix/fiber composites of β -*i*PP with α -*i*PP or PA6 fibers were prepared by compression molding at different temperatures. The tensile properties and interfacial morphologies of the obtained composites were studied. The tensile test results show that tensile yield load and the tensile modulus of β -*i*PP/PA6 matrix/fiber systems increased remarkably at the expense of elongation at break. This is normal for polymer composites. Moreover, the mechanical properties of β -*i*PP/PA6 matrix/fiber composites do not depend on the sample preparation temperature. For the *i*PP matrix/fiber single polymer composites, it was found that their mechanical properties depend strongly on the sample preparation temperature. The *i*PP single polymer composite compressed at 169 °C behaves like the *i*PP/PA6 matrix/fiber system with increased tensile modulus and decreased elongation at break, even though the elongation at break is much better than that of the *i*PP/PA6 matrix/fiber system. The *i*PP single polymer composite compressed at 171 °C shows evident increase in tensile modulus with only slight decrease of the elongation at break. Further increase in sample preparation temperature, *e.g.*, at 173 °C, improvement of both tensile modulus and elongation at break compared to the β -*i*PP was identified. The different mechanical performances of *i*PP single polymer composites originate from the different interfacial adhesion. At lower sample preparation temperatures, solid α -*i*PP fiber induces α -*i*PP crystallization, leading to the formation of α -*i*PP transcrystalline layer around the fiber. This results in an improved interfacial adhesion of the single polymer composite compared with the *i*PP/PA6 composite. At higher sample preparation temperatures, the partial surface melting of the oriented fiber allows interdiffusion of *i*PP molecular chains in the molten fiber and

matrix melt. The interpenetration of matrix *i*PP chains and molten fiber *i*PP chains results in some *i*PP molecular chains being included partially in the recrystallized fiber and the induced β -transcrystalline layers. This kind of configuration leads to an even better interfacial adhesion between the fiber and matrix, which causes a simultaneous increase of the tensile yield load, tensile modulus and elongation at break β -*i*PP.

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