Polypropylene/(Polyethylene Terephthalate/Polybutylene Terephthalate) Blend Fibres: Phase Structure and Mechanical Properties

Abstract
This paper deals with conditions for preparation, phase structure and selected properties of polypropylene (PP)-polyester (PES) blend fibres. The dispersed polyester phase in PP matrix is represented by both polyethylene terephthalate (PET) and polybutylene terephthalate (PBT), as well as by their blends (PES). The phase morphology of the blend fibres with regard to the rheological properties of the components of PP/PES blend fibres was investigated by the microscopic method. The DSC analysis and wide-angle X-ray scattering (WAXS) methods were used to evaluate the supermolecular structure of blend fibres. The correlations between the rheological properties of polymer components and the phase structure & mechanical-physical properties of the PP/PES blend fibres are discussed in this paper. We found a positive effect of the PBT in the PES-dispersed phase on the rheological properties of the blend and mechanical properties of the PP/PES blend fibres.

Key words: polypropylene, polyethylene terephthalate, polybutylene terephthalate, phase structure of blend fibres, mechanical properties.

Introduction
The polymer blends are physical ‘alloys’ of the structurally different polymers which do not have any clearly identifiable covalent bonds between the macromolecule chains of components. The chemical interactions in the real polymer blends often lead to a certain share of the copolymers, e.g. between two kinds of polymers by a re-esterification reaction. Polymer blends permit new materials to be prepared without any synthesis of the new polymers. The properties of the blends depend on a decisive degree on the properties of individual components and on the blend phase structure. The mutual compatibility of polymers and blending conditions play a very important role in creating the phase structure of polymer blends [1]. The phase structure of the fibres based on a heterogeneous polymer blend is significantly affected by the rheological properties of components and adhesive interactions at the interphase [2, 3]. The deformation of polymer melt in both the spinning and drawing of fibres leads to deformation of the particles of the dispersed phase, and a polyfibrilar system of blend fibres is formed [4 - 6].

According to Taylor & Cox’s theory, the shape of the particles in the dispersed phase of immiscible polymers in the steady uniform shear or extensional flow can be expressed by two dimensionless parameters [7, 8]: the capillary number $Ca = \frac{\eta_m \gamma}{\sigma}$ and the viscosity ratio $K = \frac{\eta_d}{\eta_m}$, where $\gamma$ is the shear rate, $\sigma$ the interfacial tension, $\eta_d$ the viscosity of the dispersed phase, and $\eta_m$ the matrix viscosity. The deformation $\varepsilon$ is expressed by the length $l$ and diameter $d$ of the deformed particles:

$$\varepsilon = \frac{(l - d)}{(l + d)}$$

Competition between drop breakup and coalescence results in critical deformation $\varepsilon_{cr}$. For shear flow, the minimum of $\varepsilon_{cr}$ is within the range of $K$ value 0.1 - 1.0, and for elongation flow, it is within $K = 1 - 5$.

In any case, the deformation of the dispersed phase is expressed by the shear rate $\dot{\gamma}$, which is indirectly proportional to $K$ [9]. The experimental results from various authors concerning the polymer blends are not unambiguous in harmony with the theory. Some authors have found the maximum of the deformation of the dispersed phase to be of the same viscosity as that of the matrix [10].

Almost all polymer blends consisting of semi-crystalline polymers in solid state display a micro-phase, heterogeneous structure [11]. Some of these blends are miscible above melting temperature in molten state, and the miscibility of components in the amorphous region has also been noted after individual crystallisation. This specific structure has been observed in some polymer blends (PET/PBT, PET/PTT) [12, 13] and polyamide blends (PA6/PA66, PA6/PA10) [14]. In particular, polyamide blends have a structure and some positive properties which are interesting from the point of view of their application in PP/PES blends, and PP/PES blend fibres with PP as a majority phase, in comparison with pure PET.

The structure and properties of the PET/PBT blends have been studied, particularly in recent years. PET/PBT blends have a lower glass temperature $T_g$ [11] and a lower melting temperature $T_m$ [15, 16] relative to PET despite having the same crystallinity. The minority phase in the concentration range under 10 wt.% cannot create its own crystalline phase. The low content of copolymer in the blend which is formed above $T_m$ provides good deformability without the necking phenomena which is characteristic for PET. Further, PET/PBT blends exhibit one glass transition $T_g$ without regard to composition. This confirms the thermodynamic miscibility of the components in the amorphous state [11, 15]. Interphase interactions and adhesion between the crystalline phase of components, resulting from their miscibility in the amorphous phase, improves mechanical-physical properties such as the tenacity and modulus of elasticity of the PET/PBT blend (referring to individual components). It can be assumed that these positive properties of the PET/PBT blends result from the interactions of aromatic rings of terephthalic acid in polymer chains, whereas the flexibility of the butandiol part of the PBT enables the development of these interactions.

In this paper, the structure and selected mechanical properties of the fibres based
materials were prepared: 1. PET; 2. PBT; 3. PET/PBT blends; 4. PP/PET/PBT, without and with the PP/PES fibres: 16 wt.% of the PES dispersed phase in the PP matrix of the PP/PES fibres were blended and spun using a laboratory spinning plant with a one-screw extruder with a diameter of 30 mm. The content of the PES dispersed phase in the PP matrix of the PP/PES fibres was 8 wt.% and 16 wt.%, without regard to the content of compatibiliser in the FPA. The parameters of the spinning process were as follows: temperature 275 °C, spinneret with 48 holes, spinning speed 400 m/min⁻¹. The fineness of the undrawn multifilaments was $T_{dt} = 960$ dtex. The multifilaments were drawn using the laboratory drawing equipment. The drawing ratio was 1 : 3, the temperature 110 °C. The fineness of the drawn multifilaments was $T_{dt} = 320$ dtex x 48. The fibres were annealed without shrinkage at 110 °C for a period of 30 min.

Methods

Rheological measurements

The rheological properties of the components and polymer blends (FPA) were measured using a Göttfert N 6967 capillary extrusiometer with extruder $\phi = 20$ mm at 275 °C. The conditions of measurement were close to those in the spinning equipment, namely dynamic conditions in extruder before extrusion of the blend melt. We used the Newton & Oswald de Waele law to determine the basic rheological parameters: apparent viscosity $\eta = \dot{\gamma}$ and power law exponent $n = k \cdot \dot{\gamma}^m$, which characterises the Newtonian behaviour of the polymer melt, where $\tau$ is the shear stress, $\gamma$ the shear rate, $\eta$ the apparent viscosity, $n$ the power law exponent, $k$ the coefficient. The viscosity ratio $K = \eta_{dd}/\eta_m$ was calculated for a constant shear rate of 100 s⁻¹ ($K_{100}$) and 500 s⁻¹ ($K_{500}$), where $\eta_d$ is the viscosity of the dispersed phase and $\eta_m$ the matrix viscosity.

DSC measurement

A Perkin Elmer DSC 7 was used to evaluate the thermal properties of the PP/PES blend fibres and to estimate the supermolecular structure and mutual interaction of the components at interface. The heating rate was 10 °C/min.

Wide-angle scattering (WAXS) method

The structure of fibres was evaluated by the wide-angle X-ray scattering (WAXS) method. The investigations were carried out using a Seifert X-ray diffractometer. The diffraction patterns were registered for samples powder on a Hardy microtome within a range of angles from 4° to 40°. The pattern analysis was performed calculating a theoretical curve approximating the experimental data. The theoretical curve was constructed as a sum of functions describing a background scattering, an amorphous halo and crystalline or mesophase peaks. The parameters of component functions were found by minimising the sum of squared deviations of the theoretical curve from the experimental one. The minimisation was carried out by means of Rosenbrock’s method using the OptiFit computer program [17]. On the basis of the WAXS patterns, the parameters characterising the fibre structure were determined. The crystallinity index and the content of the mesophase were calculated as a ratio of the area under the crystalline or mesophase peaks to the total area.

Morphology of PP/PES fibres

The size and shape of the deformed particles of the PES dispersed phase were observed, after etching the PP matrix with xylene at 140 °C, and after separation of the PES microfibres. The light microscopy observation technique was used [18].

Mechanical properties of PP/PES blend fibres

Instron type 1122 was used to evaluate the mechanical properties (tenacity and elongation) of the PP/PES blend multifilaments. The initial length of fibres was 10 cm, and the clamp rate was selected according to the limiting time to break $\approx 20$ s (modified ISO 2062). The variation coefficients (from 30 measurements) of the tensile strength and elongation at break were used as a measure of the internal (structural) unevenness of the PP/PES blend fibres.

Results and discussion

Rheology of the polymer blends (FPA) and phase structure of the PP/PES blend fibres

The dependence of the melt viscosity on shear rate indicated in Figures 1. a-d and the rheological parameters in Table 1 reveal the differences in flow properties of the PP, PET, PBT polymers and their blends. The almost Newtonian flow is
characteristic of the PET & PBT polyesters and their blends, with lower melt viscosities in comparison with PP. The non-Newtonian flow of the PP melt, characterised by a low power-law exponent (n~0.53), causes a significant decrease in melt viscosity in dependence on the shear rate, approaching the viscosity of the polyester phase (Figure 1a). A particularly low viscosity at higher shear rate was obtained for FPA based on PP/PET blends with the majority of the PET phase (Figure 1d). The lower viscosity of the polyester FPA relative to PP enables the PP/PES blend fibres to be prepared with deformed particles of the dispersed PES phase (microfibres) in the PP matrix [6]. The deformation of the PES microphase into microfibres is also supported by the lower elasticity of the melt, resulting from its almost Newtonian flow (n ≈ 1), (Table 1). The dimensionless viscosity parameter $K$, which is responsible for shape of the dispersed phase particles, does not exceed 1 (Table 1). It leads to the assumption of the high deformation of the PES microphase in the extensional flow throughout the whole range of the deformation gradient. Particularly convenient conditions for deformation of the dispersed FPA are indicated for FPA based on the PP/PET and PP/PET/PBT blends with the lowest $K_{500}$ values (Table 1).

The LiE compatibiliser decreases the viscosity of the polyester phase, which is already at a small concentration (1.5 wt.%), and contributes to the greater deformation of the dispersed phase and the formation of the polyfibrillar structure of blend fibres (PES microfibres in PP matrix).

The final shape of the PES microfibres results from the deformation of the polymer melt in both the spinning and drawing processes. Drawing the fibres can significantly contribute to an increase in length of the original PES microfibres after spinning [6]. A qualitative analysis of the results obtained confirms these assumptions (Figure 2).

Supermolecular structure

A DSC analysis of the PET/PBT blends has shown that the majority component in the blend suppresses crystallisation from the melt of the minority component (up to 30 wt.%) below, by the DSC measurable value (Table 2). At the same time, the melting temperature of the majority component decreases with the increase of content of the minority polyester. From this point of view, the PES dispersed phase can significantly affect the supermolecular structure of the PP/PES blend fibres. The results of the DSC and WAXS analyses indicate the changes in supermolecular structure of both the PES dispersed phase and the PP matrix (Figures 3 & 4, and Tables 2 & 3).

The PET component in PP/PET blend fibres exhibits a clear melting temperature characteristic of the PET & PBT polyesters and their blends, with lower melt viscosities in comparison with PP. The non-Newtonian flow of the PP melt, characterised by a low power-law exponent (n~0.53), causes a significant decrease in melt viscosity in dependence on the shear rate, approaching the viscosity of the polyester phase (Figure 1a). A particularly low viscosity at higher shear rate was obtained for FPA based on PP/PET blends with the majority of the PET phase (Figure 1d). The lower viscosity of the polyester FPA relative to PP enables the PP/PES blend fibres to be prepared with deformed particles of the dispersed PES phase (microfibres) in the PP matrix [6]. The deformation of the PES microphase into microfibres is also supported by the lower elasticity of the melt, resulting from its almost Newtonian flow (n ≈ 1), (Table 1). The dimensionless viscosity parameter $K$, which is responsible for shape of the dispersed phase particles, does not exceed 1 (Table 1). It leads to the assumption of the

![Figure 1. Dependence of the melt viscosity versus shear rate at 275°C for PP, PET and PBT polymer components (a), for binary PET/PBT blends (b), for binary PET/PBT blends with 1.5 wt.% of LiE compatibiliser (c) and for ternary polymer blends PP/PET 30/70 and PP/PET/PBT 30/60/10, content of LiE ~ 1.5 wt.% (d).](image1)

![Figure 2. Microscopic evaluation of size and shape of the PES particles dispersed in the PP/PES fibres after separation of PP matrix. PET (a), PBT (b), PET/PBT 30/70 (c), PET/PBT 70/30 (d).](image2)

<table>
<thead>
<tr>
<th>Blend composition, %</th>
<th>n</th>
<th>$K_{100}$</th>
<th>$K_{500}$</th>
<th>$K_{100 LiE}$</th>
<th>$K_{500 LiE}$</th>
</tr>
</thead>
<tbody>
<tr>
<td>PET</td>
<td>100</td>
<td>0.93</td>
<td>0.35</td>
<td>0.67</td>
<td>0.36</td>
</tr>
<tr>
<td>PET/PBT 85/15</td>
<td>0.95</td>
<td>0.33</td>
<td>0.66</td>
<td>0.36</td>
<td>0.50</td>
</tr>
<tr>
<td>PET/PBT 70/30</td>
<td>0.93</td>
<td>0.36</td>
<td>0.69</td>
<td>0.29</td>
<td>0.53</td>
</tr>
<tr>
<td>PET/PET/PBT 50/50</td>
<td>0.96</td>
<td>0.41</td>
<td>0.83</td>
<td>0.37</td>
<td>0.73</td>
</tr>
<tr>
<td>PET/PET/PBT 30/70</td>
<td>0.97</td>
<td>0.42</td>
<td>0.85</td>
<td>0.38</td>
<td>0.79</td>
</tr>
<tr>
<td>PET/PET/PBT 15/85</td>
<td>0.96</td>
<td>0.41</td>
<td>0.81</td>
<td>0.39</td>
<td>0.77</td>
</tr>
<tr>
<td>PBT</td>
<td>100</td>
<td>0.97</td>
<td>0.39</td>
<td>0.79</td>
<td>0.36</td>
</tr>
<tr>
<td>PBT</td>
<td>0.97</td>
<td>0.39</td>
<td>0.79</td>
<td>0.36</td>
<td>0.66</td>
</tr>
<tr>
<td>PP/(PET/PBT) 30/70/0</td>
<td>0.58</td>
<td>0.41</td>
<td>0.44</td>
<td>0.36</td>
<td>0.43</td>
</tr>
<tr>
<td>PP/(PET/PBT) 30/60/10</td>
<td>0.75</td>
<td>0.37</td>
<td>0.53</td>
<td>0.36</td>
<td>0.53</td>
</tr>
</tbody>
</table>
melting temperature of the PET component decreases with the increase of the content of PBT in the PES phase. The heat effect of the minority PES component (PET, PBT) disappears at concentrations under 30 wt.% in the PES phase. The same behaviour was observed for the PBT component (Table 2). The melting enthalpy of the PES phase is higher with the higher content of the crystallising component (Table 2). Unmodified PP fibre exhibits a wide melting endotherm, with clear peaks at 151 °C and 162.5 °C. The endotherm is gradually changed with the PBT (PBT) content into one peak shape above 16 wt.% of the PES phase (Figure 3). The melting temperature of 154 - 156 °C is characteristic of the PP matrix with 8 wt.% of the PES dispersed phase, and reaches about 159 °C for 16 wt.% of the PES phase (Table 2). The results of the DSC measurements of the PP/PES fibres reveal that the PES phase can act as a nucleating agent, and induces a uniform crystalline structure (with regard to polymorphic forms and crystallite dimensions) of the PP matrix, with one melting temperature about 159 °C.

The peaks observed on the WAXS patterns result from the overlapping of two broader mesomorphic peaks (at 14.8° and 21.2°), seven crystalline peaks characteristic for the α modification of polypropylene ((110) at 14.09°, (040) at 16.84°, (130) at 18.44°, (111) at 21.22°, (041) at 21.8°, (060) at 25.2° and (220) at 28.7°), and one broader amorphous peak (Figure 4). On the basis of the analysis of the WAXS patterns, the structure of the PP/PES blend fibres which are formed consists of three phases: amorphous, mesophase and crystalline phase. The maximum crystallinity index and minimum mesophase index were found for the PP/PBT fibres and for the PP/PES fibres with higher PBT content. In this case, a near-perfect supermolecular structure of the PP matrix is formed, and a positive effect of PBT on the mechanical properties of the PP/PES blend fibres is expected. There is practically no difference in the mesophase and crystallinity content between blend fibres containing different concentrations of the PES (8 - 16 wt.%), (Table 3).

### Mechanical properties of the PP/PES-annealed blend fibres

The mechanical-physical properties of fibres depend on many variables in the molecular and supermolecular structure of the oriented polymer. The mechanical properties of the blend fibres are also affected by the properties of the individual polymer components, the phase structure of the blend fibres and both their internal and external unevenness. The phase structure of the blend fibres based on immiscible polymers depends on the microstructural properties of the polymer system. This is determined by the size and shape of the dispersed phase. The internal structure unevenness of fibres results from the higher probability of weak points in individual fibres, leading to lower tensile strength and its higher variation coefficients. The structural unevenness of fibres is very often connected with external-geometrical unevenness, which may be expressed e.g. by a coefficient of variation of the fibre diameter. From these aspects, the experimental results of the tensile strength and elongation of the PP/PES blend fibres can contribute to an estimation of the technological compatibility of the components in blend-oriented fibres, as well as to identification of the main structure parameters which influence the fibres’ mechanical properties.

The tensile strength of the ternary PP/PES blend fibres grows proportionally with the content of the PBT in the PES dispersed phase. The higher PES concentration...

---

**Table 2. Melting temperature $T_m$ and melting enthalpy $\Delta H_m$ (related to weight unit of the blend) of components of the PP/PES blend fibres, (* unmeasured value).**

<table>
<thead>
<tr>
<th>Comp. of FPA</th>
<th>PP</th>
<th>FPA=PET/PBT</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>$T_m$, °C</td>
<td>$\Delta H_m$, J·g$^{-1}$</td>
</tr>
<tr>
<td>8</td>
<td>0/100/0</td>
<td>154.1</td>
</tr>
<tr>
<td></td>
<td>0/100/0</td>
<td>154.1</td>
</tr>
<tr>
<td></td>
<td>0/30/70</td>
<td>156.6</td>
</tr>
<tr>
<td></td>
<td>0/70/30</td>
<td>156.5</td>
</tr>
<tr>
<td></td>
<td>0/100/0</td>
<td>157.5</td>
</tr>
<tr>
<td></td>
<td>0/30/70</td>
<td>160.3</td>
</tr>
<tr>
<td></td>
<td>0/100/0</td>
<td>159.3</td>
</tr>
<tr>
<td></td>
<td>0/30/70</td>
<td>159.8</td>
</tr>
<tr>
<td></td>
<td>0/70/30</td>
<td>157.3</td>
</tr>
<tr>
<td></td>
<td>0/30/70</td>
<td>159.0</td>
</tr>
<tr>
<td></td>
<td>0/70/30</td>
<td>159.6</td>
</tr>
<tr>
<td>16</td>
<td>0/100/0</td>
<td>159.3</td>
</tr>
<tr>
<td></td>
<td>0/30/70</td>
<td>159.8</td>
</tr>
<tr>
<td></td>
<td>0/70/30</td>
<td>157.3</td>
</tr>
<tr>
<td></td>
<td>0/30/70</td>
<td>159.0</td>
</tr>
<tr>
<td></td>
<td>0/70/30</td>
<td>159.6</td>
</tr>
</tbody>
</table>

**Figure 3. DSC thermograms of PP/PES blend fibres (16 wt.% of PES in PP matrix); a) Pure PP fibre (1), FPA = PET/PET 70/30 (2), PP/PET 30/70 (3), PET (4), b) FPA = PET (1), PET/PBT 70/30 (2), PET/PBT 30/70 (3), PBT (4).**

**Figure 4. Series of WAXS patterns for PP/PES annealed blend fibres (16 wt.% of PES) with dispersed phase: 1. PET, 2. PBT, 3. PET/PBT 30/70, 4. PET/PET 70/30, 5. PP/PET 30/70, 6. PP/PET 70/30.**
The improvement in the mechanical properties of the PP/PES fibres has also been found when the FPA based on binary or ternary (PP/PES) blend was used for blending with the PP before spinning. The influence of the LiE compatibiliser on the mechanical properties of these blend fibres was negligible (Table 6).

**Conclusions**

We investigated the influence of the composition and rheological properties of the final polyester additive based on PET and PBT on the phase & supermolecular structures and on mechanical properties of PP/PES blend fibres. The following conclusions can be drawn from our experimental results:

- The mutual interactions in the miscible PET and PBT blend in molten state leads to a higher melt viscosity (as well as a higher viscosity ratio $K$) at higher concentration of the PBT component. The compatibiliser, which was based on the Licowax E polyester montane wax, already decreases the viscosity of the PP/PES blends (as well as the viscosity ratio $K$) at a low concentration (1.5 wt.%).

- PP/PES blend fibres exhibit a polyfibrillar structure, with microfibres of the PES dispersed phase in the PP matrix at a high aspect ratio (high deformation $c$).

- The crystallinity of the PP matrix of the PP/PES blend fibres is proportional to the PBT content in the PES dispersed phase.

- The tensile strength of the PP/PES blend fibres is proportional to the PBT content in the PES dispersed phase,
Figure 6. Influence of ternary FPA and blend fibres composition on tensile strength T, elongation E and their coefficients of variation of the PP/PES blend annealed fibres; Composition of FPA: a) PP/PET/PBT (30/70/0), b) PP/PET/PBT (30/60/10).

Table 6. Influence of ternary FPA and blend fibres composition on tensile strength T, elongation E and their coefficients of variation of the PP/PES blend annealed fibres; Composition of FPA: a) PP/PET/PBT (30/70/0), b) PP/PET/PBT (30/60/10).

<table>
<thead>
<tr>
<th>Concentration of PES in blend fibres, wt.%</th>
<th>FPA</th>
<th>T, cN/dtex</th>
<th>CV_T, %</th>
<th>E, %</th>
<th>CV_E, %</th>
</tr>
</thead>
<tbody>
<tr>
<td>8</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>a</td>
<td>2.6</td>
<td>2.5</td>
<td>23.9</td>
<td>6.4</td>
<td></td>
</tr>
<tr>
<td>b</td>
<td>2.5</td>
<td>2.8</td>
<td>26.8</td>
<td>3.7</td>
<td></td>
</tr>
<tr>
<td>a with LIE</td>
<td>2.4</td>
<td>4.2</td>
<td>27.1</td>
<td>4.4</td>
<td></td>
</tr>
<tr>
<td>b with LIE</td>
<td>2.5</td>
<td>2.7</td>
<td>26.9</td>
<td>4.0</td>
<td></td>
</tr>
<tr>
<td>16</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>a</td>
<td>2.6</td>
<td>4.1</td>
<td>26.6</td>
<td>4.3</td>
<td></td>
</tr>
<tr>
<td>b</td>
<td>2.5</td>
<td>2.1</td>
<td>25.8</td>
<td>5.1</td>
<td></td>
</tr>
<tr>
<td>a with LIE</td>
<td>2.4</td>
<td>4.6</td>
<td>25.8</td>
<td>5.7</td>
<td></td>
</tr>
<tr>
<td>b with LIE</td>
<td>2.4</td>
<td>2.8</td>
<td>25.9</td>
<td>3.9</td>
<td></td>
</tr>
</tbody>
</table>

and also proportional to the viscosity ratio K of the PES dispersed phase to the PP matrix.

The compatibiliser Licowax E causes a significant reduction in the structural unevenness of the PP/PES blend fibres and an increase in their tensile strength, mainly at higher PES phase content (16 wt.%) in the PP matrix.

Acknowledgment

The support of the APVT (grant 20-011404) and VEGA (grant 1/2475/05) grant agencies is appreciated.

References


Received 21.10.2005 Reviewed 10.04.2006

Institute of Textile Engineering and Polymer Materials

UNIVERSITY OF BIELSKO-BIAŁA
Faculty of Textile Engineering and Environmental Protection

Besides traditional textile technologies, the Institute deals with modern trends in fibre technologies. Special emphasis is put on the analysis of the supermolecular structure and the properties of polymer materials.

The main research topics are:

- physical and chemical methods of examining fibre-forming polymers,
- studies of the physical micro-structure of fibres and polymers,
- studies of the nanostructure of molecular composites,
- effect of morphology on properties of homo- and copolymers,
- conductive polymers for textile purposes,
- studies of biopolymers for biomedical purposes,
- metrological examination of textiles from finished goods, and
design and examination of textiles and clothes.

The main experimental techniques available in the Institute are:

- WAXS (Wide-Angle X-Ray Scattering),
- SAXS (Small-Angle X-Ray Scattering),
- thermal analysis combined with mass spectrometry,
- FTIR, NIR-FTIR, FT-Raman spectoscopy,
- UV-VIS spectoscopy,
- SEM (Scanning Electron Microscopy),
- optical microscopy, with fluorescence and polarising devices,
optical microscope with device for thermo-optical analysis,
electrical measurements of polymer materials,
and Mechanical tests of fibres and other polymer materials.

Moreover the Institute offers complex analysis of yarns, and knitted, woven & nonwoven textiles, using state of the art systems.

More information:

www.itimp.ath.bielsko.pl
www.wiwiost.ath.bielsko.pl