Effect of the molecular weight on the crystallinity of PCL-b-PLLA di-block copolymers

Laura Peponi^{1*}, Ivan Navarro-Baena^{1,2}, J. Eduardo Baez¹, José M. Kenny,¹ Angel Marcos-Fernández¹

¹Instituto de Ciencia y Tecnología de Polímeros, ICTP-CSIC, Juan de la Cierva, 3 28006-Madrid, Spain.

²University of Perugia, Strada di Pentima 4, 05100 Terni (Italy), Italy

Laura Peponi – lpeponi@ictp.csic.es

Iván Navarro-Baena - ivannavarro.baena@gmail.com

J. Eduardo Baez – jebaez14@yahoo.com.mx

José M. Kenny – kenny@ictp.csic.es

Ángel Marcos-Fernández – amarcos@ictp.csic.es

*corresponding autor: Laura Peponi - ICTP-CSIC, calle Juan de la Cierva, 3, 28006 Madrid.
lpeponi@ictp.csic.es, Tel: +34 91 258 74 24; Fax: +34 91 564 48 53.

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ABSTRACT

A systematic study on the synthesis and characterization of di-block copolymers based on aliphatic polyesters such as poly(L-lactic acid) (PLLA) and poly(ϵ -caprolactone) (PCL), performed varying both the length of the blocks as well as the relative content of each block in the copolymers, is reported. The block-length and the molecular weight of each synthesized PCL-b-PLLA di-block copolymer were analyzed by nuclear magnetic resonance spectroscopy. The influence of the block length and of the amount of each block on the thermal properties and the morphology, was evaluated by differential scanning calorimetry and by small angle X-Ray scattering experiments. In particular, the correlation between the molecular weight of each block and its amorphous and/or crystalline structure was obtained, evidencing that the crystallization of the PLLA block was not influenced by the presence of PCL and depends mainly on its molecular weight but the crystallization of PCL is strongly interfered by the crystallization of PLLA. In particular PLLA blocks are amorphous for short lengths (≤ 672 g/mol, that means ≤ 9.3 LA repeat units) and start to crystallize for molecular weight ≥ 964 g/mol, that means ≥ 13.4 LA repeat units.

1. Introduction

The interest on polymers for the biomedical and pharmaceutical sectors has increased widely in the recent years. New polymeric materials have been developed to provide a broad spectrum of applications such as regenerative surgery, drug delivery systems, orthopaedics and coronary stents, among others [1-6]. In this context, aliphatic polyesters such as poly(L-lactic acid) (PLLA), poly(glycolic acid) (PGA) or poly(ϵ -caprolactone) (PCL) have demonstrated to be useful for these applications [6-9]. In particular, PCL and PLLA are two semi-crystalline polymers that are easy to process, biodegradable and biocompatible [10]. These properties make them good candidates for its use in the mentioned applications and, in addition, they can also be used to replace materials from non-renewable or not environmentally-friendly sources [11]. PLLA is crystalline at room temperature, melts around 180 °C, shows good mechanical properties such as great stiffness and strength, and also has good optic and gas barrier properties. All this characteristics make PLLA a good choice in biomedical applications as well as to substitute petroleum-based polymers like PET and PP in packaging applications [12]. However, PLLA is a brittle material, which represents an important handicap in certain applications. On the other hand, PCL has a melting temperature around 60 °C and a glass transition temperature well below room temperature, thus at room temperature it behaves as a tough plastic. Copolymers [13-15], blends [16] and composites [17] with PLLA and PCL have been developed in order to take advantage of the improved synergic properties offered by both polymers. Actually, for this integration strategy, copolymerization represents a better route in comparison with blending due to the immiscibility of the homopolymers [18]. Moreover, copolymers based on PLLA and PCL show shape memory properties [19-21]. Jeon et al. have reported the synthesis and mechanical, thermal and degradability characterization of both homopolymers and tri-block, multiblock and random PCL-PLLA copolymers [22]. Castillo et al. reported that the nucleation process and the morphology of both blocks depends on the percentage of them in the copolymer [23]. In particular they discovered that both copolymer blocks are partially miscible depending on their molecular weight, decreasing the nucleation rate and the melting temperature of PLLA block by increasing the percentage of PCL. Maglio et al. investigated on PLLA-PCL-PLLA tri-block copolymers with high molecular weight and they concluded that the two blocks can be considered immiscible [24]. The degradation rate of PLLA-PCL copolymers has been reported by Zhao and col., who have developed materials with adjustable degradation time depending on the composition of the copolymer [5].

The aim of this paper is to systematically study di-block copolymers of PCL and PLLA focusing on the effect of the molecular weight of the blocks on the crystallization behaviour. In fact, it is known, that the degree of crystallinity and thermal properties of these homopolymers depends on molecular weight [25-28]. Thus, in copolymers where there are interactions between both blocks, the molecular weight of each block plays an important role on their thermal behaviour [22, 23]. To understand these phenomena, a wide series of well-defined PCL-b-PLLA di-block copolymers were synthesized, varying the length of the blocks as well as the relative content of each block. Nuclear magnetic resonance (^1H NMR) and MALDI-TOF were used to characterize the chemical structure of the synthesized copolymers. Their thermal properties were studied by calorimetry (DSC) and their phase separated structure by small angle X-Ray scattering (SAXS). To our knowledge, such a systematic study on PCL-b-PLLA di-block copolymers has not been previously reported.

2. Experimental

2.1 Materials

L-lactide (L-LA) was kindly donated by Purac Biomaterials. ϵ -Caprolactone (CL), 1-butanol and tin(II) 2-ethylhexanoate ($\text{Sn}(\text{Oct})_2$) were purchased from Aldrich. All materials were used without further purification.

2.2 Synthesis of poly(ϵ -caprolactone)s (PCL)

The polymerization was carried out in bulk in a 100 ml round bottom flask previously dried. ϵ -caprolactone (CL, 50 g, 0.438 mol) and 1-butanol (117.4, 54.0, 25.9 or 10.1 mmol) were charged in the flask, 0.1% wt of $\text{Sn}(\text{Oct})_2$ respect to the CL monomer was added and a condenser with a calcium chloride tube was fitted. The mixture was then heated with magnetic stirring in an oil bath at 130 °C for 24 hours. After the reaction time, when still hot, a vacuum pump was connected to the flask, and the residual CL and 1-butanol removed. Finally, four different PCL monols of theoretical molecular weight 500, 1000, 2000 and 5000 named respectively BuPCL500-OH, BuPCL1000-OH, BuPCL2000-OH, BuPCL5000-OH were obtained. The main characteristics of the different synthesized PCL in terms of molecular weight and thermal behavior are reported in the Supporting Information and listed In **Tables 1_{SI}** and **2_{SI}**.

2.3 Synthesis of poly(ϵ -caprolactone)-*b*-poly(L-lactide) (PCL-*b*-PLLA)

Di-block copolymers were synthesized in bulk by using the following procedure: PCL monol and L-lactide in the appropriate amounts to obtain approximately 5 grams of di-block were charged in a 15 ml round bottom flask previously dried, and 0.1 wt % of Sn(Oct)₂ respect to the amount of LA monomer was added. The flask was stoppered and heated in an oil bath at 180 °C for 3 h. After the reaction time, when still hot, a vacuum pump was connected to the flask, and the residual LA removed. After cooling the reaction mixture, the resulting di-block copolymer, when solid, was dissolved in chloroform and precipitated in cold methanol. In some cases, a white precipitate in high yield was obtained and the precipitate dried under vacuum. In other cases, when the precipitate was obtained in relatively low yield, the filtered solvent was removed and the remaining material, either a waxy solid or a liquid, was isolated and dried under vacuum. Finally, when the resulting di-block copolymer was a liquid, it was stored under vacuum. The main characteristics of the synthesized di-block copolymers in terms of calculated molecular weight, number of repeated units [29] and the percentage of each block are reported in the Supporting Information and summarized in the **Table 3_{SI}**.

2.4 Characterization

Solution ¹H NMR spectra were recorded at room temperature in a Varian Unity Plus 300 instrument using deuterated chloroform (CDCl₃) as the solvent. Spectra were referenced to the residual solvent protons at 7.26 ppm.

Differential scanning calorimetry (DSC) was performed in a Mettler Toledo DSC822e instrument. Samples were sealed in aluminum pans. Two scans (0–200 °C and -90–200 °C) were performed with a heating rate of 10 °C/min and a cooling rate of 10 °C/min between runs under nitrogen purge. The melting points (T_m) and crystallization temperatures (T_c) are given as the maximum and minimum of the endothermic and exothermic transition, respectively. Data reported were taken from the second scan.

SAXS measurements were taken at beamline BM16 at the European Synchrotron Radiation Facility (Grenoble, France). Samples were melted and sealed in DSC aluminium pans, placed within a Linkam hot stage and heated at 10°C/min while the SAXS spectra were recorded. Calibration of temperature gave a difference of approximately 7°C between the temperature reading at the hot stage display and the real temperature at the sample. Detector calibration was done with silver behenate (AgC₂₂H₄₃O₂), and the long period L was calculated as $2\pi/q$ (q is the scattering vector).

3. Results and discussion

3.1 Synthesis and characterization of the chemical structure of the PCL-b-PLLA di-blocks

In order to make a deep analysis on the influence of the chemical structure on the properties of PCL-b-PLLA di-block copolymers, different series of well-defined PCL-b-PLLA di-block copolymers were synthesized varying both the molecular weight of the blocks as well as the relative content of each block in the block copolymers.

Four PCL monols of different molecular weight were prepared by bulk polymerization of ϵ -CL using 1-butanol as initiator and $\text{Sn}(\text{Oct})_2$ as catalyst. Yield was quantitative. Analysis by ^1H NMR, allowed for the determination of the actual molecular weight of the monols. An example of the ^1H NMR is reported in Fig. 1_{SI} of the Supporting Information.

Two methods could be used to perform the calculation: a) the ratio of the signal e of the CL units to the signal D of the initiator (M_{nI} in Supporting Information, Table 1_{SI}); b) the ratio of the signal a' of the terminal CL units to the signal a of the internal CL units (M_{nII} in Supporting Information, Table 1_{SI}). Both methods resulted quite precise, with a difference of around 1% between the values for the shorter PCL, and of 4 and 6% for the longer PCLs, where the small signal of the terminal groups results in a higher error. Because signal D is higher in intensity than signal a' , M_{nI} values were considered more precise and used in the calculations for the next synthetic step. In addition, ^1H NMR showed a negligible amount of residual unreacted ϵ -CL, less than 0.3 % by weight in the final product.

The analysis of the thermal properties shows the high level of crystallinity achieved by all the PCL monols, with similar values and above 58%. However, the melting point of monols is very different, increasing with the PCL molecular weight, as expected: (data reported in Table 2_{SI} of the Supporting information) [30].

For the two longer PCL the melting point above 50 °C indicates well developed crystals, whereas for the two shorter PCL, the appearance of several maxima and the lower melting point indicates less perfect crystals. In the case of nBuPCL500-OH, melting point is so low that it remains as a very soft paste at room temperature. With the so obtained PCL monols, several series of PCL-b-PLLA di-block copolymers were prepared, with theoretical PCL:PLLA weight contents of 15:85, 20:80, 30:70, 50:50, 70:30, 80:20 and 85:15. As already indicated in the experimental section, in some cases two fractions with different block composition were obtained for certain di-block copolymers, and finally, a collection of 35 di-block copolymers was isolated, see Table 3_{SI} reported in the Supporting Information.

The molecular weight of each block and the actual PCL:PLLA weight ratio on the isolated di-block copolymers were calculated from ^1H NMR spectra [31]. In **Fig. 1**, the spectrum for copolymer P4 is shown as an example indicating the most relevant peaks for the PCL block as well as for the PLLA block.

Fig. 1. ^1H NMR spectrum of P4, (15_{PCL}-85_{PLLA}) di-block copolymer. The general chemical structure of the copolymer is reported on top. (SS= Spinning Side bands).

Characteristic signals for polymerized caprolactone and polymerized lactide were observed. Multiplet from 5.05-5.25 ppm is assigned to methine proton of polymerized lactide (*f*), whereas the small signal at approximately 4.35 ppm is related to terminal LA units (*f'*, as shown in the inset on the right side). Multiplet from 4.08-4.18 ppm is due to protons *a* from CL units linked to LA units, while triplet at 4.05 ppm is related to protons *a* from CL units linked to another CL unit plus protons *A* from initiator. No signal for terminal CL molecules (at 3.74 ppm) from unreacted PCL monol were detected for this sample, but it appeared in some copolymers with high PCL content. The multiplet between 2.34-2.44 ppm is due to the proton *e* from CL units linked to a LA unit, while the triplet at 2.30 ppm is related to the proton *e* from CL units linked to other CL units. For the rest of the spectrum, multiplets at 1.66 ppm and 1.39 ppm were related to the CL protons *b*, *d*, plus initiator protons *B*, and CL protons *c* plus initiator protons *C* respectively, and multiplet at 1.56 ppm to the LA methyl protons *g*. The triplet at 0.9 ppm is due to the initiator proton *D*. No rests of unpolymerized lactide (at approximately 5.03 ppm) and unreacted ϵ -caprolactone (at approximately 2.635 ppm) were detected. Moreover, by ^1H NMR analysis, the absence of trans-esterification of PLLA onto PCL chains was confirmed.

As for the PCL monols calculations, the ratio of the signals *D* and *e* was used to calculate the real PCL molecular weight on the di-block, and once calculated, the ratio of the signal *a* of CL units to the signals

($f+f'$) of LLA units allowed for the calculation of the length for the PLLA block and the PCL:PLLA weight ratio. The calculated weight ratio is fairly close to the theoretical one, except for some copolymers, where the actual PCL content is significantly lower than expected from feed ratio, probably due to the loss of species of lower molecular weight produced during precipitation of the copolymer (see Supporting Information, table 3_{SI}). The length of the PCL block in the copolymers is similar to the theoretical one in the series with low PCL content and correspondingly long PLLA blocks. At 50, 70 and 80% PCL content, the corresponding PLLA block becomes short and for some of the copolymers, two fractions were isolated: a precipitate with PCL molecular weight much higher than theoretical, and one material isolated by evaporation of the solvent with a PCL molecular weight closer to the value of the starting PCL monol. For copolymers with 85% PCL content, the non precipitated fraction of copolymer P25 was not isolated, and for copolymers P26 and P27, the precipitated fraction had a PCL molecular weight close to theoretical, whereas the fraction isolated by solvent evaporation was composed of species of lower molecular weight. For some of the copolymers with 80 and 85% content of PCL and correspondingly short PLLA block, a significant amount of unreacted PCL monol was detected. Importantly, we finally obtained a collection of 35 PCL-b-PLLA di-block copolymers with a well characterized chemical structure with different relative content of the blocks and different block lengths in order to study their crystallization properties. MALDI-TOF experiments (data not shown) confirmed that the ring-opening polymerization proceeded correctly.

3.2 Thermal properties of the PCL-b-PLLA di-blocks

The thermal properties and the crystalline nature of PCL-b-PLLA di-blocks were studied by DSC and the results in terms of glass transition temperature (T_g), melting point (T_m), enthalpy of crystallization (ΔH_m) and gravimetric crystallinity fraction X_c are summarized in Table 4_{SI} of the Supporting Information.

DSC thermograms corresponding to the second heating scan for four different examples of the synthesized PCL-b-PLLA di-blocks, corresponding to a copolymer where only the PCL is crystalline, a copolymer where only the PLLA block is crystalline and two copolymers where both blocks are crystalline, are reported in the Supporting Information, in Fig. 2_{SI}.

Initially, the crystallinity of the PLLA blocks was studied. Although there is a broad interval of values reported for the melting enthalpy of pure PLLA 100% crystalline [32-33], the most commonly used value, taken by us for calculations, is close to 93 J/g [34-37].

As shown in **Fig. 2**, PLLA blocks are amorphous for short lengths (≤ 672 g/mol, that means ≤ 9.3 LA repeat units) and start to crystallize for longer length values (≥ 964 g/mol, that means ≥ 13.4 LA repeat units). These results can be compared with the ones reported in the literature [29]. Once the PLLA blocks are able to crystallize, the crystallinity increases very fast with the increase on the PLLA molecular weight irrespective of the length or the relative content of the blocks, and already above 1872 g/mol, at about 26 LA repeat units, the crystallinity of the PLLA is higher than 50%, reaching an apparent plateau with a maximum crystallinity of approximately 65%. This value is similar to that of some PCL-b-PLLA di-block copolymers found in literature with PLLA length on the same range studied here (un-filled circles in **Fig. 2**) [32].

Fig. 2. Degree of crystallinity vs molecular weight of the PLLA blocks. (Filled squares represent the experimental points; unfilled circles data taken from Ref. 32). Line is just a guide for the eye.

It is clear that, at the temperatures where PLLA blocks crystallize, PCL blocks are amorphous and have no effect on PLLA crystallinity except for the restriction imposed on the chain end of the PLLA block that could have an influence on the minimum length needed for the PLLA blocks to crystallize. This minimum length for PLLA block crystallization is similar or slightly higher than for PE-b-PLLA di-block copolymers (with very incompatible blocks) where PLLA blocks crystallize already at 713 g/mol [34], and it is lower than for PEO-b-PLLA di-block copolymers (with partially miscible blocks) where PLLA blocks still do not crystallize at a length of 1000 g/mol [38].

A similar curve is found for the melting point value of the PLLA crystals (**Fig. 3**). At short PLLA length, the crystals are small and imperfect and melt at low temperature starting at 99 °C for a PLLA length of 964 g/mol. With the increase in PLLA length, the melting point of the crystals increases rapidly reaching an almost constant value slightly below 170 °C for the longest PLLA block. Comparing with data found in literature for PCL-b-PLLA di-block with PLLA length on the range studied here, the

values are similar to those reported here in some cases (un-filled squares in **Fig. 3**, taken from reference 23) and slightly higher in others (un-filled circles in **Fig. 3**, taken from reference 32).

Fig. 3. Melting point vs molecular weight of the PLLA blocks. (Filled squares represent the experimental points; un-filled circles represent data taken from ref. 32; un-filled squares represent data taken from ref. 23).

Compared with other double-crystalline block copolymers, the melting points of the PLLA blocks are similar to the values found for very incompatible PE-*b*-PLLA di-block copolymers [34], and much higher than for partially miscible PEO-*b*-PLLA di-block copolymers [39].

This is the first time that such a systematic study on the effect of PLLA length on its crystallinity when being part of a di-block copolymer has been carried out, as well as it is the first time that the minimum length for PLLA block crystallization has been determined with such accuracy. In addition, when studying these double-crystalline di-block copolymers, it is usual to relate the melting point of the PLLA block with the PLLA content on the copolymer [23]. In a recent review describing all the reported double-crystalline di-block copolymers, all the graphs are systematically presented plotting the melting point of the block with the highest melting temperature (PE, PLLA,...) as a function of the block content but not of its length [40]. We have here demonstrated in our PCL-*b*-PLLA di-block copolymers that PLLA content has a little effect on the PLLA block melting point and that melting point of the PLLA block depends almost entirely on the PLLA block length with little interference from the PCL block. This has been already stated by Pensec et al. [32], that synthesized just seven of these di-block copolymers, all of them with crystalline PLLA block, and therefore they could not determine the minimum length for PLLA crystallization. To understand PCL block crystallization, it is better to divide

the data in two sets: a) crystallization when the PLLA block is amorphous, see **Fig. 4a**, and b) crystallization when the PLLA block is crystalline, see **Fig. 4b**.

Fig. 4. Degree of crystallinity vs molecular weight of the PCL blocks when the PLLA block is amorphous (a), when the PLLA block is crystalline (b). In the inset of figure 6a a magnification of the graph is presented. Lines are just a guide for the eye.

When PLLA blocks are unable to crystallize (**Fig. 4a**), PCL is amorphous up to a length of 851 g/mol, and needs a minimum block length of 1133 g/mol to start to crystallize. Prepared PCL monols used for di-block synthesis were crystalline even at a molecular weight of 619 g/mol, whereas in di-blocks, the mobility restriction imposed by the amorphous PLLA block at one end of the PCL chain produces an increase in the minimum length for the PCL blocks to crystallize. Once this minimum length is reached, crystallinity is very high, in between 60% and 70% approximately, and when PCL block length increases, crystallinity is slightly reduced and very soon, at 1664 g/mol of PCL molecular weight, remains almost constant at around 55%.

On the other hand, when PLLA is able to crystallize (**Fig. 4b**), PLLA crystals are produced before PCL crystallization and greatly hamper its ordering except for some of the di-block copolymers with 50% content of each block. In four of these 50_{PCL}-50_{PLLA} copolymers, although PLLA crystallizes, PCL crystallinity is high (i.e. 53% for P13 with 28% of PLLA crystallinity, 24-25% for P14 with 45-48% of PLLA crystallinity and 45% for P15 with 56% of PLLA crystallinity). The length of the PLLA blocks ranges from 964 g/mol for P13 to 2701 g/mol for P15, and it seems that because of the high content of PCL blocks and the small size of the PLLA crystals (melting point of the crystals for these copolymers are the lowest for the synthesized di-blocks in this work) due to the short length of the PLLA blocks, the crystallization of the PCL blocks is not so much reduced. For the rest of the di-blocks where PLLA is

able to crystallize, PLLA crystallization greatly hampers PCL crystallization and the PCL block needs a molecular weight above 2000 g/mol to crystallize. From this point, PCL crystallinity increases with PCL molecular weight following an exponential trend, and at molecular weights above approximately 6500 g/mol, the maximum crystallinity of around 55% is reached. Although there are not more experimental points above this PCL molecular weight, based on the results found for PCL di-blocks with amorphous PLLA blocks, it is expected that the maximum PCL crystallinity has been already achieved and from this point a plateau would continue the curve. Melting point of the PCL crystals, however, is not much affected by PLLA crystallinity.

Fig. 5. Melting point vs molecular weight of the PCL block when PLLA block crystallizes.

As it can be seen in **Fig. 5**, melting point steadily increases with PCL molecular weight irrespective of the PLLA crystallinity, approaching the melting point of pure PCL at the highest length of the PCL block.

3.3 SAXS characterization of the PCL-*b*-PLLA di-blocks

The SAXS experiments corresponding to the same samples reported in the Supporting Information for the DSC analysis (figure 2_{SI}), are reported on **Fig. 6**. This means a copolymer where only the PCL is crystalline, a copolymer where only the PLLA block is crystalline and two copolymers where both blocks are crystalline. For lamellar systems it is well known that the long period, L , of the phase separation can be calculated directly from the curve Iq^2 vs. q , while for other morphologies a calculation of L from the I vs. q curve gives more realistic results. In this research we have not specifically investigated the copolymer morphology is present, but it is well known that PCL and PLLA

homopolymers crystallize with a lamellar morphology, and this is the reported morphology for their di-block copolymers [38-41] and other double-crystalline di-block copolymers [42, 43]. In addition, the representation Iq^2 vs. q highlights and makes it easier to detect the maximum, especially when using automated macros for the processing of the high amount of data generated in a synchrotron experiment.

Fig. 6. Scattering curves (Iq^2 vs. q) from the SAXS experiments for: a) P25:85_{PCL}-15_{PLLA}, where only the PCL block is crystalline; b) P3:15_{PCL}-85_{PLLA}, where only the PLLA block is crystalline; c) P20:70_{PCL}-30_{PLLA}, where both blocks are crystalline; d) P12:30_{PCL}-70_{PLLA}, where both blocks are crystalline.

A single broad peak was obtained in the scattering curves (Iq^2 vs. q). Two parameters can be calculated from the scattering curves: the relative invariant, Q' , as the integral below the curve Iq^2 vs. q , which is related to the extent of the phase separation; and the maximum on the scattering curve, $q_{\max}$, related to the size scale of the separated phases, calculated also from the curve Iq^2 vs. q . In **Fig. 7**, the evolution of the relative invariant and of the long period L with the temperature is shown for the same samples as for **Fig. 6**. When only PCL is crystalline (P25 in **Fig. 7**), the scattering increases steadily with temperature due to the thermal expansion of the material and finally disappears after melting as indicated by the drastic drop in Q' , leaving a homogeneous melt. The scattering maximum decreases (distance increases) during melting, approaching the beam stop until complete disappearance. For

copolymers where only the PLLA blocks are crystalline (P3 in **Fig. 7**), initially the peak is shallow and Q' increases steadily with temperature until approximately 85 °C, where a jump in Q' takes place due to the crystallization of the PLLA blocks, and the peak increases and becomes more defined. This is a marked difference respect to DSC experiments and is due to the differences in the sample preparation: DSC samples were taken from precipitated copolymers, where the PLLA blocks can crystallize, whereas SAXS samples were obtained by melting the precipitated polymers in DSC pans, and cooling of these samples was fast enough to produce a substantial amount of amorphous PLLA that could crystallize on heating when carrying out the SAXS experiment. After PLLA crystallization, Q' continues its steady growth until melting begins and Q' value starts to decrease until complete melting where scattering is lost and a homogeneous melt is produced.

Fig. 7. Relative invariant and long period from the SAXS experiments for: a) P25:85_{PCL}-15_{PLLA}, where only the PCL block is crystalline; b) P3:15_{PCL}-85_{PLLA}, where only the PLLA block is crystalline; c) P20:70_{PCL}-30_{PLLA}, where both blocks are crystalline; d) P12:30_{PCL}-70_{PLLA}, where both blocks are crystalline.

The value of the distance in the maximum of the scattering vector increases until crystallization of the PLLA blocks, then it reaches a small plateau and starts to grow again with temperature, and when melting of PLLA blocks initiate, the maximum approaches the beam stop faster (distance increases faster) until its disappearance. On cooling (not shown), at a certain temperature a maximum appears due

to PLLA crystallization. This maximum grows in intensity when temperature decreases and remains with an almost constant $q_{\max}$. For copolymers with both blocks crystalline, the drop in Q' at high temperature is due to the melting of the PLLA blocks, and at lower temperatures, we can find either an increase due to PLLA blocks crystallization when PCL crystallinity is low (P12 in **Fig. 7**), or a change in Q' due to PCL blocks melting when PCL crystallinity is high (P20 in **Fig. 7**).

Also for these copolymers, after PLLA blocks melt, the scattering disappears leaving a homogeneous melt. For the PCL-b-PLLA di-block copolymers described in literature, after PLLA block melting, a homogeneous melt is obtained in some cases [40, 46], and micro-separated phases in other cases [40, 41, 43, 47]. For the latter, in reference 41, no scattering was evident by SAXS after PLLA blocks melting, but by rheology measurements, it was concluded that 10K PLLA block produced a homogeneous melt whereas approximately 50K PLLA block gave micro-separated phases up to 220 °C. In reference 43, 6.9K PLLA blocks led to a homogeneous phase whereas for 11.1K and 12.4K PLLA blocks a maximum in SAXS was evidence of a phase separated morphology in the melt. Finally, in reference 47, a maximum in scattering corresponding to a distance of 20.4 nm, demonstrated the separated phases of a di-block copolymer with a 40K PLLA block. For our synthesized copolymers, with PLLA block lengths reaching up to 27K, we did not find evidence of scattering, and therefore, phase separation, after complete melting of the PLLA blocks. For other double-crystalline di-block copolymers, when the blocks are strongly incompatible such as PE-b-PCL, PE-b-PEO and PE-b-PLLA,[45, 48-50] blocks remain immiscible in the melt, whereas for more compatible blocks such as PEO-b-PCL and PEO-b-PLLA, melt is homogeneous [39, 40]. If we compare our PCL-b-PLLA copolymers with PE-b-PLLA and PEO-b-PLLA di-block copolymers, the compatibility of the blocks is intermediate. The melting point of the PLLA blocks is very close to the pure PLLA homopolymer melting point, as for PE-b-PLLA copolymers, showing that when the blocks crystallize, are very incompatible, but less than PE and PLLA blocks, that do not mix in the melt; PEO and PLLA blocks do mix in the melt as our PCL-b-PLLA copolymers, but when crystallized, the lower melting point of the PLLA block in the PEO-b-PLLA copolymers indicates the significant mixing of PEO blocks in the phase of the PLLA blocks.

Long period values L follows the same trend as the relative invariant: increases when PLLA blocks crystallize or also when PCL blocks melt followed by a small plateau, and increases until melting of the PLLA blocks. On cooling, a maximum due to PLLA crystallization appears (data not shown).

From SAXS experiments, the melting of the PLLA blocks can be measured at the last point of the Q' drop (end of melting), and the calculated values follow the same trend found by DSC although the

values are approximately 10 °C higher due to the thermal lag of the Linkam hot stage plus the difference due to the different criteria for assessing the melting temperature (maximum of the endothermic peak by DSC and end of melting by SAXS). When long period values L are analyzed, for copolymers with only crystalline PCL blocks, the only ones where L is solely related to PCL crystals, a parallelism is found between the increase in the melting point of the crystals and the increase on the distance in the maximum of the scattering curves at ambient temperature, as seen on **Fig. 8**, confirming that the melting point is related to the size of the crystals.

Fig. 8: Crystallinity and long period vs PCL molecular weight for copolymers with only crystalline PCL blocks. Lines are just a guide for the eye.

For copolymers with crystalline PLLA, L was calculated in the plateau after PLLA crystallization, at 100 °C, except for P14, P15, P16, P20 and P24, the copolymers with a melting point below 150 °C, where the plateau was found at lower temperatures and L was calculated at 70 °C. For copolymers with crystalline PLLA, no relationship between either melting point or percent of crystallinity (or block length) with L was found. However, it seems that there is a certain influence of the structure on L . As seen on **Fig. 9**, when PLLA block content is high, above 50%, L lies in between 10 and 14 nm.

Fig. 9: Long period vs molecular weight for PLLA block.

When PLLA block content is around 50%, for PCL length below 2000 g/mol, L values are still in between 10 and 14 nm, but when PCL length is above 2000, L increases to a value around 15 nm. When PLLA content is lower than 50%, L increases to more than 20 nm, with a higher value at lower PLLA content.

4. Conclusions

A series of 35 PCL-b-PLLA di-block copolymers were synthesized in bulk from four PCL monols and using $\text{Sn}(\text{Oct})_2$ as catalyst. Analysis by ^1H NMR confirmed the di-block structure of the copolymers and allowed for the calculation of the actual block lengths and relative amount of each block. Molecular weight of PCL block ranged from 560 to 6970 g/mol, and molecular weight of PLLA block ranged from 136 to 27100 g/mol, whereas the studied relative theoretical weight contents of PCL:PLLA blocks were 15:85, 20:80, 30:70, 50:50, 70:30, 80:20, and 85:15.

From calorimetric results, it was found that PLLA blocks need a minimum length in between 672 (9.3 LA repeat units) and 964 (13.4 LA repeat units) g/mol to crystallize, and above this value until 2000 g/mol, crystallinity rapidly increases. Finally the crystallinity reaches a plateau with a maximum crystallinity of around 65% and a melting point close to 170 °C. PCL blocks also need to reach a minimum value to crystallize, and crystallization is strongly affected by the PLLA crystallization. When PLLA is amorphous, PCL is highly crystalline for molecular weights above a value between 851 and 1133 g/mol reaching a maximum value at higher molecular weights of around 55%. When PLLA is crystalline, and except for some copolymers with 50:50 weight ratio, PCL minimum length to crystallize

is above 2000 g/mol and from this value PCL crystallinity increases following an exponential trend until a value of approximately 6500 g/mol where the maximum PCL crystallinity of around 55% is achieved. For the di-blocks with 50:50 weight ratio and short PLLA blocks (900 to 2700 g/mol), PLLA crystals do not strongly interfere on PCL crystallization and high values of PCL crystallinity are achieved.

For all the di-block copolymers, after melting of the blocks, the resulting material was amorphous with no evidence of phase separation, as found by SAXS. For copolymers with only PCL crystals, the higher the melting point of the crystals, the higher the long spacing. For copolymers with crystalline PLLA, no relationship was found between PLLA crystallinity or PLLA melting point and the long spacing of the phase separated morphology with only PLLA crystals present and the PCL blocks amorphous or molten. Only the relative content of PLLA in the di-blocks affected the long spacing values at low PLLA content. Thus, when PLLA block content is high, above 50%, L lies in between 10 and 14 nm. When PLLA block content is around 50%, and PCL length is above 2000, L increases to a value around 15 nm, and when PLLA content is lower than 50%, L increases to more than 20 nm, with a higher value at lower PLLA content.

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