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結合理論計算與實驗方法定量分析溶劑對生物醫學用共聚 化合物的選擇性及其微結構的影響(2/2)

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☐ 期中進度報告

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中文摘要

關鍵詞：塊狀共聚物溶液；自身聚集行為；溶劑選擇性

塊狀共聚物溶液，因著溶劑的選擇性以及溶劑的濃度，對其自身聚集的變化非常多元，可形成奈米級尺寸的各種微結構。目前雖然大部分的實驗結果均能定性地與理論計算結果一致，然而影響微結構形成的重要因素『溶劑選擇性』，隨著實際系統中共聚物與溶劑之間交互作用力的不同，有著非常大的差異；因此，為了要透徹地了解所欲研究的實際塊狀共聚物溶液系統如何應用於奈米級微結構中，首要的工作應為『溶劑選擇性』的定量分析，我們計畫比較理論計算與實驗方法所得到的結果，來定量分析出溶劑對於塊狀共聚物的選擇性。所研究的系統為目前非常熱門的生物醫學材料，例如 PEG-PLLA 共聚物。我們首先探討共聚物其結晶行為及微結構尺寸如何受到鏈段比例及結晶溫度的影響；接著運用理論計算定量分析各塊狀分子與溶劑之間的交互作用參數，探討 PEG-PLLA 共聚物在各種不同選擇性下的溶劑中，其微胞型態如何受到共聚物的組成、溶劑選擇性、以及溶劑的濃度之影響。

英文摘要

Keywords: block copolymer solutions; self-assembly behavior; solvent selectivity

Block copolymer solutions can self-assemble into various microstructures due to the effects of solvent selectivity. Though most of the related experimental and theoretical results are qualitatively consistent, there exists little quantitative comparison between both fields. This is due to the fact that it is difficult to determine the solvent selectivity in real systems. Therefore, we aim to quantitatively determine the interaction parameters between each block and the solvent by comparing both experimental and numerical results. We consider a poly(ethylene-glycol) -polylactide (PEG-PLLA) diblock copolymer. In particular, PLLA component can crystallize below the melting temperature. As PLLA is biodegradable and PEG is non-toxic to human bodies, these materials play a very important role on the biomedical research. We first focus on the crystallization as well as the structure analysis of PEG-PLLA copolymer melts. We then determine the interaction parameters between PEG-solvent and PLLA-solvent. We finally analyze the effects of solvent selectivity, copolymer volume fraction, and copolymer composition on the micelle formation of PEG-PLLA copolymers in the presence of a selective solvent.

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Introduction (前言、目的與文獻探討)

Biodegradable block copolymers continue to attract a lot of attention due to their numerous biomedical applications [1-3]. Much of the research has focused on the synthesis as well as the available applications. Although there have been recently a few studies in discussing the crystallization behavior and structure development [4-15], the crystallization kinetics in crystalline-crystalline biodegradable block copolymers has not been well understood. Thus, we adopt a type of self-synthesized crystalline-crystalline diblock biocopolymers to systematically analyze the crystallization kinetics during the isothermal crystallization processes. We then analyze their self-assembling behavior in the presence of a selective solvent.

We consider poly(L-lactide)-*block*-monomethoxy poly(ethylene glycol) (PLLA-*b*-MePEG) diblock copolymers. PLLA and PEG are known to be biocompatible. Their block copolymers have been proposed for a wide range of medical applications due to drug permeability and degradability [1,16]. So far, most studies related to PLLA-PEG block copolymers have focused on the synthesis and the thermal crystallization behavior, which are primarily based on the nuclear magnetic resonance (NMR), differential scanning calorimeter (DSC), and wide-angle X-ray diffraction (WAXD) measurements [4,6-7,11,13-14,17-20]. For example, Kim et al. [6] investigated the variation of the crystal structure and crystallization behavior with the block ratio in the PLLA-PEG diblock and triblock copolymers by WAXD and DSC experiments. They found that the crystal unit-cell structures of both PLLA and PEG are independent of the block length and the crystallization temperature. However, the crystallization of one block is deeply affected by the presence of the other block that is chemically connected to it. Sun et al. [13] synthesized a series of PLLA-MePEG block copolymers with identical MePEG molecular weight but various PLLA molecular weights. From the DSC experiments, they observed that with increasing the molecular weight of PLLA in the copolymers, the melting point of PLLA increases; but the melting point of MePEG tends to decrease. This phenomenon indicates that the crystallization of both components is greatly influenced by the presence of the other component. In addition, by comparing the PLLA-PEO-PLLA triblock copolymer with the same composition of binary PLLA and PEO blends, Shin et al. [14] observed that the chain connectivity in the triblock copolymer can reduce the chain mobility and the crystallization of each component thus decreases. Also they observed that the degree of optical retardation due to the crystallization of PEO was greater in the triblock copolymer.

In the first part of this project, we study the crystal unit-cell structures as well as the isothermal crystallization kinetics of PLLA in the PLLA-MePEG diblock copolymers by WAXD and DSC techniques. To our knowledge, there have been extensive studies that have focused on the crystallization kinetics of pure PLLA [21-28]. However, there exist few studies on the PLLA crystallization kinetics in the presence of MePEG that is chemically connected to it. We therefore synthesize a series of PLLA-MePEG block copolymers with the same MePEG molecular weight but various PLLA molecular weights. PLLA and PEG are reported to be miscible in the melt [22]. The melting temperatures of PLLA and MePEG are around 150-190 °C and 60 °C, respectively [29]. As such, when the samples are first melted at 200 °C and then quenched to $T_C \geq 70$ °C, we expect that only PLLA component can crystallize in the presence of amorphous MePEG block from a homogeneous melt state. We then examine the effects of the connected MePEG block on the formation of PLLA crystals and the crystallization behavior of PLLA at various temperatures T_C . From the WAXD analysis, the crystal unit-cell structure parameters as a function of T_C and the block ratio between PLLA and MePEG are studied. With the aid of DSC, we examine the melting temperature and the crystallinity of PLLA as a function of PLLA molecular weight and crystallization temperature. We also analyze the relative crystallinity of PLLA as a function of time via the Avrami equation, from which the crystallization growth rate of PLLA at a given T_C is obtained. Based on the Hoffman-Lauritzen analysis of the PLLA growth rate vs. crystallization temperature, we investigate the crystallization kinetics of regime transitions between II and III. In particular, the transition temperature from regime III to II, the activation energy for segment diffusion to the crystallization site U^* , and the fold surface free energy σ_e , are compared for each

diblock sample and pure PLLA.

In the second part of this project, we quantitatively determine the interaction parameters between each block and the solvent by comparing both experimental and numerical results. Once these interaction parameters are quantified, we analyze the effects of solvent selectivity, copolymer volume fraction, and copolymer composition on the micelle formation of PEG-PLLA copolymers in the presence of a selective solvent.

Experimental Section (實驗方法)

Materials and Synthesis

PLLA-MePEG diblock copolymers were synthesized by a ring-opening polymerization. L-lactide (Tokyo Kasei Kogyo Co.) was purified by recrystallization in ethyl acetate. The monomethoxy-terminated poly(ethylene glycol) (MePEG) with the number average molecular weight $M_n = 5000$ was purchased from Aldrich Co.. First, a certain weight ratio of L-lactide and MePEG (shown as Table 1) was put into a dry tri-stoppered flask with stirring. The flask was purged with nitrogen and placed in a silicone oil bath at 110 °C. After the mixture was fully melted, a few amount of the stannous octoate (1% of PLLA weight) was introduced to serve as the catalyst. The system was continuously purged with nitrogen and the temperature was raised up to 130 °C to proceed the ring-opening polymerization for 24 h. The reaction product was then dissolved in chloroform with stirring for 2-3 hours at room temperature, and precipitated by a mixture of *n*-hexane and methyl alcohol under an ice bath. The above purification process (dissolution/precipitation) was carried out for four times. The precipitate was finally filtered and dried under vacuum at 45 °C for 3 days. For the sake of comparison, PLLA homopolymer has been synthesized as well.

We employed $^1\text{H-NMR}$ (Jeol EX-400, deuterated chloroform) to determine the PLLA composition in the copolymer, f_{PLLA} , which is equal to $\frac{N_{\text{PLLA}}}{(N_{\text{PLLA}} + N_{\text{MePEG}})}$ with N_I = degree of polymerization of component *I*. The molecular weights of the samples were then obtained based on f_{PLLA} and the fixed number average molecular weight of MePEG ($M_n = 5000$). With the aid of gel permeation chromatography (GPC; PLgel, 5 μm Mixed-C, Polymer Laboratories Ltd.), we determined the polydispersities of the samples. Tetrahydrofuran (THF) was used as the evolving solvent at a rate of 0.8 mL/min. Polystyrene ($M_n = 580\sim 377400$) was used as a standard. Finally, we listed the characteristics of the PLLA-MePEG diblock copolymers and PLLA homopolymer used in this study in Table 1.

Wide-Angle X-ray Diffraction (WAXD)

The samples were first heated to 200 °C in the oven for 10 min annealing, and then rapidly transferred to another oven preheated to the desired crystallization temperature ($T_c \geq 70$ °C) for 24 h in order for PLLA to crystallize completely. Finally, the specimens were quenched to 25 °C so that MePEG can crystallize in the matrix of PLLA crystalline domains. WAXD experiments were carried out on the samples with a Rigaku Denki diffractometer with Cu K α radiation ($\lambda = 1.542$ Å) at a scanning rate 2° θ /min, where θ is the scattering angle (the angle between the incident X-ray beam and the scattered X-ray beam). The accelerating voltage was 40 kV, and the tube current was 100 mA. The X-rays were monochromated with a graphite. All measurements were performed at room temperature.

Differential Scanning Calorimeter (DSC)

The kinetics of isothermal crystallization of PLLA homopolymer and PLLA-MePEG diblock copolymers were analyzed by a PerkinElmer DSC-7 differential scanning calorimeter. First, the samples were heated from room temperature to 200 °C and annealed for 10 min in the melt in order to remove the thermal history. The samples were then quickly quenched to the desired crystallization temperature ($T_c \geq 70$ °C) at a cooling rate of 100 °C/min, and kept at this temperature to the end of the exothermic PLLA crystallization peak. The heat flow per gram of the sample evolved during the isothermal crystallization process was recorded as a function of time.

After complete crystallization of PLLA during the isothermal crystallization, the samples were heated from the crystallization temperature to a melt with a heating rate of 10 °C/min. The heat flow per gram of the sample evolved during the scanning process was measured as a function of temperature, from which the melting temperature of PLLA was determined from the maxima of the

melting peaks, and the apparent enthalpy of melting per gram (ΔH) contributed from the melting of PLLA crystals was obtained from the areas of the melting peaks. Hence, the crystallinity of PLLA component was determined.

Theoretical Section (理論方法)

我們運用蒙地卡羅(Monte Carlo)模擬方法得到單體與單體之間的交互作用能量參數。一但作用參數決定，我們運用分散粒子動力學(Dissipative particle dynamics, DPD)方法來模擬 PEG-PLLA 共聚合物的微胞形成如何受到共聚合物組成、共聚合物體積分率以及溶劑選擇性的影響。

Monte Carlo Simulation

若一般的多重積分不易得到數理值時，可用蒙地卡羅方法計算。如下列的多重積分

$$I = \int_0^1 \int_0^1 \cdots \int_0^1 f(x_1, x_2, \dots, x_N) dx_1 dx_2 \dots dx_N \quad (1)$$

計算此積分之蒙地卡羅方法為在 x 可能的區間內 ($0 \rightarrow 1$) 任意選取 n 組點 $\{x_1^{(i)}, x_2^{(i)}, \dots, x_N^{(i)}\}$ ，計算其對應的函數 f 的值。則此積分可近似為

$$I = \frac{1}{n} \sum_{i=1}^n f(x_1^{(i)}, x_2^{(i)}, \dots, x_N^{(i)}) \quad (2)$$

其計算的誤差約為 $n^{-1/2}$ ，接著考慮如下的積分式

$$F = \int_{x_1}^{x_2} f(x) dx \quad (3)$$

若將上式改寫為

$$F = \int_{x_1}^{x_2} \left(\frac{f(x)}{p(x)} \right) p(x) dx \quad (4)$$

式中 $p(x)$ 為一種機率的分配函數。設在範圍 $[x_1, x_2]$ 間依 $p(x)$ 選取 n 個亂數 ξ_x ，則此積分可寫成

$$F = \left\langle \frac{f(\xi_x)}{p(\xi_x)} \right\rangle \quad (5)$$

式中 $\langle \rangle$ 為選取 n 次的平均值。利用此方法可計算許多統計學中的積分，例如在 canonical 系集中，任一特殊坐標的函數 F 之平均值為

$$\bar{F} = \frac{\int \cdots \int F(\vec{r}_1, \vec{r}_2, \dots, \vec{r}_N) \exp\left(-\frac{U(\vec{r}_1, \vec{r}_2, \dots, \vec{r}_N)}{kT}\right) d\vec{r}_1 d\vec{r}_2 \dots d\vec{r}_N}{\int \cdots \int \exp\left(-\frac{U(\vec{r}_1, \vec{r}_2, \dots, \vec{r}_N)}{kT}\right) d\vec{r}_1 d\vec{r}_2 \dots d\vec{r}_N} \quad (6)$$

或將上式寫成

$$\bar{F} = \int \cdots \int F(\vec{r}_1, \vec{r}_2, \dots, \vec{r}_N) p(\vec{r}_1, \vec{r}_2, \dots, \vec{r}_N) d\vec{r}_1 d\vec{r}_2 \dots d\vec{r}_N \quad (7)$$

其中

$$p(\vec{r}_1, \vec{r}_2, \dots, \vec{r}_N) = \frac{\exp\left(-\frac{U(\vec{r}_1, \vec{r}_2, \dots, \vec{r}_N)}{kT}\right)}{\int \cdots \int \exp\left(-\frac{U(\vec{r}_1, \vec{r}_2, \dots, \vec{r}_N)}{kT}\right) d\vec{r}_1 d\vec{r}_2 \dots d\vec{r}_N} \quad (8)$$

為系統於 $(\vec{r}_1, \vec{r}_2, \dots, \vec{r}_N)$ 結構的機率。

由上式，當 $P(E_{ij})$ 代表交互作用能量的概率分布， E_{ij} 是各對DPD粒子各自平均的交互作用能量。則各對(pair)DPD粒子平均的交互作用能量 $\langle E_{ij} \rangle$ 對溫度的分佈平均 $\langle E_{ij}(T) \rangle$ 可由蒙地卡羅方法獲得：

$$\langle E_{ij}(T) \rangle = \frac{\int dE_{ij} P(E_{ij}) E_{ij} e^{-E_{ij}/kT}}{\int dE_{ij} P(E_{ij}) e^{-E_{ij}/kT}} \quad (9)$$

Dissipative Particle Dynamics

分散粒子動力學(Dissipative particle dynamics, DPD) 方法首先被 Hoogerbrugge 和 Koelman[30]提出，是一種適用於大尺度, 長時間的模擬技術。這個技術針對在更大且複雜的流體行為系統中，模擬時間可達到微秒範圍。DPD 模擬技術已成功地應用在如 block-copolymer microphase separation、特定區域的變化成長和流體相分離等領域上。

在 DPD 模擬中，一個分子內具有同樣化學性質的的片段以一系列的微粒"beads"來表示。DPD 粒子代表系統中的一個小區域範圍，並且粒子的行動依牛頓力學式決定。

作用在DPD粒子上的總力量是由粒子和其它DPD粒子之間的交互作用力 F^C 、分散力 F^D 以及隨機作用力 F^R 所組成。

$$F_i = \sum_{i \neq j} (F_{ij}^C + F_{ij}^D + F_{ij}^R) \quad (10)$$

F^C 是以各粒子為中心的交互作用力，其值跟 a_{ij} 相關。此處 r_{ij} 是DPD粒子間向量 $\hat{r}_{ij}$ 的值，

$$F_{ij}^C = \begin{cases} a_{ij}(1-r_{ij})\hat{r}_{ij} & r_{ij} < 1 \\ 0 & r_{ij} > 1 \end{cases} \quad (11a)$$

隨機作用力 F^R 作用在所有不同的粒子對之間。會對系統增加平均能量。而分散力 F^D 會以減慢微粒和抵消能量的形式存在。隨機作用力 F^R 與分散力 F^D 一起作用，類似作為系統的一個恆溫器，維持能量穩定。

$$F_{ij}^R = \begin{cases} \sigma \omega^R(r_{ij}) \zeta_{ij} \hat{r}_{ij} & r_{ij} < 1 \\ 0 & r_{ij} > 1 \end{cases} \quad (11b)$$

$$F_{ij}^D = \begin{cases} -\gamma \omega^D(r_{ij}) (\hat{r}_{ij} \cdot \mathbf{v}_{ij}) \hat{r}_{ij} & r_{ij} < 1 \\ 0 & r_{ij} > 1 \end{cases} \quad (11c)$$

ζ_{ij} 為 delta-correlated stochastic 因子，且 $\langle \zeta_{ij}(t) \rangle = 0$ ， $\langle \zeta_{ij}(t) \cdot \zeta_{kl}(t') \rangle = (\delta_{ik} \delta_{jl} + \delta_{il} \delta_{jk}) \cdot \delta(t-t')$ 。
 ω 為 weight function，且 $\omega^D(r) = [\omega^R(r)]^2$ 。

計算 F^C 要先求得交互作用的參數 a_{ij} 的值。這個參數與DPD粒子混合的能量有關。由這些粒子混滲能量的平均值 $\langle E_{ij}(T) \rangle$ ，計算可求得 Flory-Huggins 參數 $\chi_{ij}(T)$ ，

$$\chi_{ij}(T) = \frac{E_{mix}^{ij}(T)}{RT} = \frac{(Z_{ij} E_{ij}(T) + Z_{ji} E_{ji}(T) - Z_{ii} E_{ii}(T) - Z_{jj} E_{jj}(T))}{2RT} \quad (12)$$

接著依照由 Groot 和 Warren [31] 提出的式子可求得指定的DPD粒子密度 $\rho = N/V$ 下的參數 a_{ij}

$$a_{ij}(T) = a_{ii} + 3.497\chi_{ij}(T) \quad \text{for } \rho = 3 \quad (13a)$$

$$a_{ij}(T) = a_{ii} + 1.451\chi_{ij}(T) \quad \text{for } \rho = 5 \quad (13b)$$

所有長度、質量、時間均以約簡單位度量。m為DPD粒子質量， r_c 為DPD粒子間作用力截斷半徑(cut off radius)

$$\bar{r} = \frac{r}{r_c} \quad \bar{v} = \frac{v}{\sqrt{k_B \frac{T}{m}}} \quad \bar{t} = \frac{t}{\sqrt{\frac{mr_C^2}{k_B T}}} \quad (14)$$

在每個運動階段，由起初的粒子位置及所受作用力求出下一瞬間的粒子位置。再由新的位置求出新受的作用力以後，各個粒子經計算得到新的速度，再開始下一個時間步程(Time Step， Δt)。

$$r_i(t + \Delta t) = r_i(t) + \Delta t v_i(t) + \frac{1}{2}(\Delta t)^2 f_i(t) \quad (15a)$$

$$\tilde{v}_i(t + \Delta t) = v_i(t) + \lambda \Delta t f_i(t) \quad (15b)$$

$$f_i(t + \Delta t) = f_i(t + \Delta t, \tilde{v}_i(t + \Delta t)) \quad (15c)$$

$$v_i(t + \Delta t) = v_i(t) + \frac{1}{2} \Delta t (f_i(t) + f_i(t + \Delta t)) \quad (15d)$$

最後由在相空間中所有粒子依運動方程式所得出的軌跡求得 DPD 粒子的動態行為。

Results and Discussion (結果與討論)

(I) Crystallization Kinetics of PLLA-MePEG Diblock Copolymers

WAXD Analysis

We employ WAXD to examine the unit-cell structures of both crystalline PLLA and MePEG components for a series of PLLA-MePEG diblock copolymers (S1, S2, and S3), which are first isothermally crystallized at various T_C for PLLA crystallization and then quenched to 25 °C for MePEG crystallization. For a comparison, we also examine the WAXD results for pure PLLA crystallized at various T_C and for pure MePEG crystallized at 25 °C. Typical WAXD patterns in terms of Intensity (I) and scattering angle (θ) for each sample are shown in Figure 1, where PLLA component is crystallized at $T_C = 100$ °C and MePEG at 25 °C. As denoted in Figure 1, the main peak positions as well as their corresponding (hkl) reflection planes from pure PLLA and pure MePEG are consistent with those from ref. 32 and ref. 33, respectively, indicating that PLLA and MePEG belong to the orthorhombic and monoclinic crystal system, respectively. In further, the peak positions from the WAXD profiles for each diblock sample are a superposition of the peak positions from pure PLLA and pure MePEG, respectively, which reveals that each crystallizable component in the diblocks forms a similar unit-cell structure as the pure component. For pure components, the unit-cell structure parameters (a , b , c) are easily determined through the insertion of the values of λ (1.542Å) and the main peak positions of the reflection planes into the form of the interplanar spacing of the (hkl) reflection planes, which is given by

$$\left(\frac{1}{d_{hkl}}\right)^2 = \left(\frac{2 \sin(\theta_{hkl}/2)}{\lambda}\right)^2 = \frac{h^2}{a^2} + \frac{k^2}{b^2} + \frac{l^2}{c^2} \quad (16a)$$

for PLLA belonging to the orthorhombic system, and

$$\left(\frac{1}{d_{hkl}}\right)^2 = \left(\frac{2 \sin(\theta_{hkl}/2)}{\lambda}\right)^2 = \frac{1}{\sin^2 \beta} \left(\frac{h^2}{a^2} + \frac{k^2 \sin^2 \beta}{b^2} + \frac{l^2}{c^2} - \frac{2hl \cos \beta}{ac} \right) \quad (16b)$$

for MePEG belonging to the monoclinic system. In eq 1b, the value of β is equal to 125.4° [33]. For diblock samples, since the crystalline peaks contributed from both PLLA and MePEG are overlapped, the WAXD patterns need to be deconvoluted first in order to determine the unit-cell parameters of PLLA and MePEG crystals. Typical peak positions of the (hkl) planes, which contribute to the crystalline intensities, as well as the corresponding θ values used in the deconvolution procedures for pure PLLA and MePEG are listed in Table 2a and 2b, respectively. The WAXD deconvolution results to a combination of possible crystalline reflections as well as the amorphous phase with Gaussian curves for pure PLLA and MePEG, are then presented in Figure 2a and 2b, respectively. With the same crystalline reflection planes listed in Table 2, similar deconvolution procedure is also employed for each diblock sample, as shown in Figure 2c where we present the deconvolution results for diblock sample S3. It is clear that both the deconvolution results and the experimental results for each sample are in a very good agreement. Once the crystalline peaks contributed to each component for diblock samples are separated, we thus can calculate the unit-cell structural parameters of both PLLA and MePEG for the diblocks by eq 1. Table 3 lists the unit-cell parameters for both PLLA and MePEG crystals in the diblocks as well as for pure PLLA and pure MePEG. As can be seen clearly, the unit-cell parameters for each diblock sample are almost identical to those for pure components [6], which also agree well with the reported values [32,33]. That is, the formation of the PLLA and MePEG crystal unit-cell structures is not affected by the chain-connectivity as well as the block ratio between PLLA and MePEG. Also, these parameters do not change with the earlier crystallization temperature of PLLA T_C , indicating that not only the PLLA crystal unit-cell parameters remain the same with T_C ; but also there are no distortions of the MePEG crystal structures due to the presence of the earlier PLLA crystalline domains at T_C .

DSC Analysis

Melting Behavior

Figure 3a shows the DSC heating scans of pure PLLA that is crystallized at various values of crystallization temperature T_C and then heated at a rate of 10 °C/min. It is evident that two melting temperature peaks for the PLLA component, $T_{m1,PLLA}$ and $T_{m2,PLLA}$ as indicated by arrows in Figure 3a, are observed when $T_C \leq 105$ °C. Figure 3b presents the values of $T_{m1,PLLA}$ and $T_{m2,PLLA}$ versus T_C for pure PLLA component. As can be seen clearly, $T_{m2,PLLA}$ is independent of T_C , and even disappears when T_C increases to be greater than 105 °C. The value of $T_{m1,PLLA}$ is observed to be linearly increasing with T_C when $T_C \geq 95$ °C. By extrapolating the line of $T_{m1,PLLA}$ versus T_C to the line $T_m = T_C$ according to the linear Hoffman-Weeks analysis [34], we obtain the equilibrium melting temperature of PLLA, $T_{m,PLLA}^\circ$, which is around 196 °C. This result is in a good agreement with ref. 22, but comes with a difference of 10 °C compared with the value of $T_{m,PLLA}^\circ$ equal to 206 °C listed in the literature [21,23]. We believe this is due to the difference of the molecular weight of PLLA. As such, the equilibrium thickness of the crystallites as well as the equilibrium melting temperature varies. The existence of double melting peaks in the DSC heating profiles of pure PLLA may result from one of the following reasons: the presence of two different crystal structures, the presence of two different thicknesses of crystal lamellae with the same type of crystal structure formed at the isothermal crystallization conditions [35], and the simultaneous melting - reorganization/recrystallization - remelting of the lamellae originally formed during the crystallization process [36]. Our previous WAXD results have shown that only one unit-cell structure of PLLA crystals forms during the crystallization process. Thus, the occurrence of the double melting peaks is mainly caused by the existence of the single type of crystal unit-cell structure but with different crystal thicknesses. Are these two different thicknesses of crystals formed at the beginning of the isothermal crystallization process or at the later heating process due to the reorganization and/or the recrystallization of the original lamellar stacks? As illustrated in Figure 3a, the lower melting temperature peak $T_{m1,PLLA}$ becomes more obvious; while the higher peak $T_{m2,PLLA}$ gradually decreases as the crystallization temperature T_C increases. Furthermore, $T_{m2,PLLA}$ even disappears when T_C is above 105 °C. Moreover, $T_{m2,PLLA}$ also disappears with an increase in the heating rate. Recall that at higher crystallization temperatures, it is easier to form more perfect and larger crystal lamellae, and therefore it is more difficult for the reorganization and/or the recrystallization process to occur. Also, when the heating rate is too fast, there is no sufficient time for the original lamellar stacks to undergo the reorganization and/or the recrystallization process. As such, the fact that the higher melting peak $T_{m2,PLLA}$ does not exist in the DSC melting endotherms at high crystallization temperature and/or heating rate indicates that $T_{m2,PLLA}$ is a result of the melting of the crystallites recrystallized during the heating process. While the lower melting peak $T_{m1,PLLA}$ refers to the melting of the primary crystallites formed during the isothermal crystallization process.

Figures 4a, 5a, and 6a present the DSC melting endotherms of block copolymer samples S1, S2, and S3, respectively, first allowing PLLA component to isothermally crystallize at various T_C from a homogeneous melt state at 200 °C, and then heated at a rate of 10 °C/min. As illustrated in each figure, two melting temperature peaks for the PLLA component, $T_{m1,PLLA}$ and $T_{m2,PLLA}$, are also observed in the diblocks. In order to obtain the equilibrium melting temperature of PLLA for each block copolymer sample, we plot the values of both $T_{m1,PLLA}$ and $T_{m2,PLLA}$ versus T_C for S1, S2, and S3, in Figures 4b, 5b, and 6b, respectively. Similar to the analysis of pure PLLA, we obtain the value of $T_{m,PLLA}^\circ$, for S1, S2, and S3, equal to around 165 °C, 159 °C, and 147 °C, respectively.

This result shows a trend that the $T_{m,PLLA}^{\circ}$ value in the copolymers decreases with decreasing the PLLA molecular weight. However, compared with the value of $T_{m,PLLA}^{\circ}$ of pure PLLA, the equilibrium melting temperature of PLLA in the diblock samples decreases dramatically, suggesting that the presence of the miscible PEG blocks also has a great influence on the crystallization as well as the melting behavior of PLLA. That is, the dependence of the PLLA melting temperature in the diblock copolymers is not simply due to the variation of PLLA molecular weight. The dilution effect caused by the miscible MePEG also plays a very important role on this sharp $T_{m,PLLA}^{\circ}$ decreasing behavior.

The crystallinity of PLLA in the copolymer, $\omega_{mc,PLLA}$, is then calculated by

$$\omega_{mc,PLLA} = \Delta H / \Delta H_{PLLA}^{\circ} \quad (17a)$$

where ΔH is the apparent enthalpy of melting per gram contributed from the melting of PLLA crystals, and ΔH_{PLLA}° is the heat of melting per gram of 100% crystalline PLLA equal to 81 J/g [23]. To examine the effects of copolymer composition on the crystallinity of PLLA, we plot the normalized crystallinity of PLLA, defined as

$$X_{mc,PLLA} = \omega_{mc,PLLA} / wt_{PLLA} \quad (17b)$$

in Figure 7. As seen in Figure 7, the crystallinity of pure PLLA remains somewhat a constant when T_C is within 80 ~ 130 °C. As $T_C \leq 75^{\circ}\text{C}$, since it is close to the glass transition temperature of PLLA, the crystallinity of pure PLLA drops dramatically. However, within the same crystallization temperature regime, the normalized crystallinity of PLLA in the diblocks first shows an increasing trend with T_C and then a decreasing behavior with a further increase in T_C . The existence of such a maximum is quite reasonable for systems undergoing the crystallization with the process of nucleation and growth. That is, at low degrees of supercooling, the decreasing degree of the formed nuclei is much larger than the increasing degree of the PLLA mobility, and thus the crystallizability of PLLA decreases with T_C . However, at high degrees of supercooling, although the formed nucleation density increases, its increasing degree is still overcome by the great decrease of the mobility, and thus the crystallizability of PLLA decreases with T_C decreasing. Because decreasing PLLA molecular weight depresses the value of $T_{m,PLLA}^{\circ}$, the sample with lower PLLA molecular weight at the same T_C indeed undergoes a crystallization at a smaller degree of supercooling. As such, we observe that the crystallization temperature, at which the normalized PLLA crystallinity in the diblocks reaches a maximum, shifts to a lower value with decreasing the PLLA molecular weight. This showing a maximum behavior has also been observed in the crystallization growth rate of PLLA with respect to T_C , which will be discussed later.

Isothermal Crystallization Behavior

Figure 8a presents the DSC exothermic curves as a function of time t for each diblock sample and pure PLLA undergoing the isothermal crystallization of PLLA at $T_C = 90^{\circ}\text{C}$, from which the relative crystallinity of PLLA, $X(t)$, can be calculated by

$$X(t) = \frac{\int_0^t \frac{dH(t)}{dt} dt}{\int_0^{\infty} \frac{dH(t)}{dt} dt} \quad (18)$$

where $dH(t)/dt$ is the specific heat flow recorded at time t during the isothermal crystallization process. Once the values of $X(t)$ vs. t are obtained, as shown in Figure 8b, the isothermal crystallization kinetics where only the PLLA block crystallizes are interpreted in terms of the Avrami equation [37]

$$X(t) = 1 - \exp(-kt^n) \quad (19)$$

where n is known as the Avrami index, and k is the overall crystallization rate constant including

contributions from nucleation and crystal growth. It is clear that the fit of the Avrami equation to the experimental data of each sample crystallized at 90 °C is remarkably good in the whole conversion range, which is also true for the systems crystallized at other values of T_C . As listed in Table 4, where we present the fitting Avrami constants n and k , most of the n values for the temperature examined here are in the range between 3 and 4, indicating that the growing of spherulites due to the crystallization of PLLA is three-dimensional. However, for the sample S3 at $T_C \geq 100$ °C, n deviates to a higher value 5, which may result from the presence of major amorphous MePEG domains. Thus the growing of the ordinary PLLA spherulites is destroyed and the n value deviates from 3-4. Indeed, with increasing the amount of amorphous MePEG, the resulting spherulites vary from ordinary to banded to dendritic (tree-like) [13].

With the values of the overall crystallization rate k determined from the Avrami analysis, the time for half of the crystallization to develop, $t_{1/2}$, is calculated by

$$t_{1/2} = \left(\frac{\ln 2}{k} \right)^{1/n} \quad (20)$$

from which the crystallization growth rate, G , defined as $G = 1/t_{1/2}$, is obtained. Figures 9a-9d plot G vs. T_C for pure PLLA, S1, S2, and S3, respectively. We observe that with increasing the crystallization temperature of PLLA for each sample, the growth rate first shows an increasing and then a decreasing behavior. This behavior has been very common due to the balance between two well-known opposing effects on the crystallization rate. As T_C decreases and approaches the glass transition temperature, T_g , the crystallization growth rate is greatly retarded by the significant decrease of the chain mobility. While when T_C is high and approaches the equilibrium melting temperature T_m° , although the chain mobility increases, it is overcome by the great decrease of the formed nucleation density, and the crystallization rate decreases at low degrees of supercooling. However, it should be noted that the curve of the temperature dependence of the growth rate G for each sample in the temperature range studied here does not show a single bell shape. Instead, the experimental G data with T_C is best fitted with two bell-shape curves, which are mainly attributed to the different regimes that the spherulite growth is associated with and manifested below.

Based on the Hoffman-Lauritzen theory [38,39], the growth rate G of a linear polymer crystal with folded chains is given by

$$G = G_0 \exp\left(-\frac{U^*}{R(T_C - T_\infty)}\right) \exp\left(-\frac{K_g}{T_C \Delta T f}\right) \quad (21)$$

where G_0 is a pre-exponential term, U^* is the activation energy for segment diffusion to the crystallization site, R is the universal gas constant, T_∞ is the hypothetical temperature where all motion associated with the viscous flow ceases and defined as $T_g(K) - 30$, ΔT is the degree of supercooling $T_m^\circ - T_C$, the correction term $f = \frac{2T_C}{T_m^\circ + T_C}$ is introduced to account for the change in

heat of fusion with T_C . K_g is a parameter associated with the energy needed for the formation of nuclei of critical size and expressed as

$$K_g = \frac{\alpha b_0 \sigma \sigma_e T_m^\circ}{\Delta h_f k_B} \quad (22)$$

in which b_0 is the monomolecular thickness, σ and σ_e are the lateral and fold surface free energy, respectively, Δh_f is the enthalpy of fusion per unit volume of crystal, and k_B is the Boltzman constant. The value of α depends on the growth regime, and equal to 4 for regimes I (high T_C) and III (low T_C) and 2 for regime II (intermediate T_C). As a result, the value of K_g for regimes I and III should be twice of that for regime II.

In order to verify the crystallization regimes in the temperature examined here for each sample,

we plot $\ln G + \frac{U^*}{R(T_C - T_\infty)}$ vs. $\frac{1}{T_C \Delta T f}$ in terms of various values of U^* , in which the intercept and the slope correspond to $\ln G_0$ and K_g , respectively. For the analysis, the glass transition temperature T_g of PLLA for each sample is set around 48.3 °C, which is determined by DSC. The equilibrium melting temperature of PLLA, $T_{m, PLLA}^\circ$, for pure PLLA, S1, S2, and S3, is equal to 196 °C, 165 °C, 159 °C, and 147 °C, respectively, which has been obtained previously. Note that the values of $\ln G_0$ and K_g are strongly dependent of U^* . For example, Figure 10a plots

$\ln G + \frac{U^*}{R(T_C - T_\infty)}$ vs. $\frac{1}{T_C \Delta T f}$ with the value of $U^* = 1500$ cal/mol and 2900 cal/mol for pure

PLLA. It is clear that the experimental data for both values of U^* can be fitted with two straight lines having high and low slopes, which may correspond to regime III and regime II, respectively. In Table 5, we list the corresponding values of $\ln G_0$ and K_g for regimes II and III, and the transition temperature from regime III to regime II, $T_{C, III \rightarrow II}$. When U^* is set at 1500 cal/mol, which is frequently used for PLLA [21], the value of K_g for regime II, $K_{g(II)} = 3.0 \times 10^5 \text{ K}^2$ is between the reported values of $1.85\text{--}5 \times 10^5 \text{ K}^2$, and K_g for regime III, $K_{g(III)} = 5.02 \times 10^5 \text{ K}^2$ is also in the range between the reported values of $4\text{--}9 \times 10^5 \text{ K}^2$ [21,23,25-28]. However, the ratio $K_{g(III)}/K_{g(II)} = 1.67$ is not consistent the theoretical value 2 for our experimental data when $U^* = 1500$ cal/mol. Indeed, this discordance has also been observed in the literatures [23,25,27-28]. In order to obtain the best fit with $K_{g(III)}/K_{g(II)}$ equal to 2, we find the value of U^* for pure PLLA is around 2900 cal/mol. In this case, the value of K_g for regimes II and III is equal to $4.58 \times 10^5 \text{ K}^2$ and $9.10 \times 10^5 \text{ K}^2$, respectively. Though the ratio $K_{g(III)}/K_{g(II)}$ varies with the value of U^* , the transition temperature from regime III to regime II, $T_{C, III \rightarrow II}$, remains somewhat a constant 120 °C, which is in a good agreement with previous articles [23,25-28]. In order to examine the factor f effects, we also

analyze the plot of $\ln G + \frac{U^*}{R(T_C - T_\infty)}$ vs. $\frac{1}{T_C \Delta T f}$ with $f = 1$ (i.e., no correction term) at various values of U^* . In a comparison with the results obtained from including the correction term, we find that at a fixed value of U^* , though the intercept value $\ln G_0$ varies, the slope value K_g does not vary much with f term included or not. As such, when $f = 1$, the best fit with the ratio $K_{g(III)}/K_{g(II)}$ equal to 2 is also expected at $U^* = 2900$ cal/mol for pure PLLA, as shown in Figure 10a. Similar behavior has also been observed for diblock copolymer samples S1, S2, and S3. For the further discussion and simplicity, we only present the results based on the analysis without the correction

term ($f = 1$). As can be seen clearly in Figures 10b-10d, where we plot $\ln G + \frac{U^*}{R(T_C - T_\infty)}$ vs.

$\frac{1}{T_C \Delta T}$ with the insertion of appropriate values of U^* for S1, S2, and S3, respectively, we can also

find a transition from regime III to regime II with the ratio of $K_{g(III)}/K_{g(II)}$ close to 2 for each copolymer sample. All the parameters including U^* , $T_{C, III \rightarrow II}$, and $\ln G_0$ and K_g of regimes II and III for each copolymer sample are also listed in Table 5. As expected, we observe a similar trend of the transition temperature $T_{C, III \rightarrow II}$ as the behavior of $T_{m, PLLA}^\circ$. Also, the activation energy for the segment diffusing to the crystallization site U^* decreases with decreasing the PLLA molecular weight and/or composition. As reported in ref. 24, a decrease in the PLLA molecular weight may cause the decrease of U^* slightly. However, we believe that the presence of amorphous MePEG which is chemically connected to PLLA can also promote the ability for the PLLA chains to diffuse to the crystallization sites, which thus contributes to the decrease of U^* too.

Once all the parameters related to the crystallization kinetics of regimes II and III are

determined, the theoretical values of the crystallization growth rate G as a function of T_C for each sample are then calculated via eq 6 and plotted as the dashed bell-shaped curve and the solid bell-shaped curve, for regimes II and III, respectively, in each Figure 9a-9d. It is evident that the experimental G data with T_C are described very well by these two bell-shaped curves for each sample. From these two calculated G curves for each sample shown in Figures 9a-9d, we then obtain the temperature $T_{C,max}$, at which the crystallization growth rate reaches a maximum G_{max} , for each crystallization regime II and III, as also presented in Table 5. Similar to the transition temperature $T_{C,III \rightarrow II}$ behavior, the value of $T_{C,max}$ in both regimes also decreases with decreasing the PLLA composition. In more details, the ratios of $T_{C,max}/T_m^\circ$ in regimes II and III for pure PLLA are about 0.806 and 0.834, respectively, which are very close to the reported value of 0.83 for a wide variety of polymers [40]. For each diblock sample studied here, the ratios of $T_{C,max}/T_m^\circ$ in regimes II and III remain somewhat constants about 0.84 and 0.87, respectively, which are slightly higher than those for pure PLLA. In a comparison of the maximum growth rate G_{max} and $G_\circ$ between pure PLLA and diblock copolymers, we observe no specific trend of G_{max} as well as $G_\circ$ for both regimes with the PLLA molecular weight. While it has been found that both G_{max} and $G_\circ$ have the same molecular weight dependence, and therefore $G_{max}/G_\circ$ should be a constant for a given homopolymer with various molecular weights [41,42]. Nevertheless, it is obvious that all the values of G_{max} in both regimes for each diblock copolymer sample are larger than those for pure PLLA.

It is interesting to find that if we replot the calculated G vs. T_C results in terms of the reduced crystallization growth rate (G/G_{max}) and the reduced crystallization temperature ($T_C/T_{C,max}$), all the plots based on the calculated results for pure PLLA and each diblock sample in both regimes II and III, which are represented as the dashed and solid curves in Figure 11, almost overlap as a master curve. Since these calculated G results are obtained from the best fit of the experimental data, the plots of experimental data G/G_{max} vs. $T_C/T_{C,max}$ are expected to fall on this single master curve, too (see Figure 11). In fact, by equating the derivative of G in eq 6 with respect to temperature T_C to 0, i.e., $\partial G/\partial T_C = 0$, $T_{C,max}$ satisfies the following equation:

$$\frac{U^*}{R(T_{C,max} - T_\infty)^2} + \frac{K_g(T_m^\circ - 2T_{C,max})}{T_{C,max}^2(T_m^\circ - T_{C,max})^2} = 0 \quad (23)$$

and the maximum growth rate G_{max} has the form of

$$G_{max} = G_\circ \exp \left[\frac{-U^*}{RT_m^\circ \left(2 \frac{T_{C,max}}{T_m^\circ} - 1 \right)} \left(1 + \frac{\frac{T_\infty}{T_m^\circ} \left(1 - \frac{T_\infty}{T_m^\circ} \right)}{\left(\frac{T_{C,max}}{T_m^\circ} - \frac{T_\infty}{T_m^\circ} \right)^2} \right) \right] \quad (24a)$$

and

$$\ln \left(\frac{G}{G_{max}} \right) = \ln \left(\frac{G_{max}}{G_\circ} \right) \times \frac{(1-X)^2}{X(A-X)} \times \frac{X(1-2B+AB) - B(A-1)^2}{(X-B)(1-2B+AB)} \quad (24b)$$

where X is the reduced crystallization temperature, $T_C/T_{C,max}$, and A and B are equal to $T_m^\circ/T_{C,max}$ and $T_\infty/T_{C,max}$, respectively. With the above derived equations, Okui et al. [41,42] analyzed the temperature dependence of the data of the crystallization growth rate for various homopolymers in the literature, and reported that the experimental data of G/G_{max} vs. $T_C/T_{C,max}$ are well described by eq 9. Since it has been well known that both ratios of $T_{C,max}/T_m^\circ$ and T_∞/T_m° are nearly constants for a wide variety of homopolymers, the ratio of $G_{max}/G_\circ$ is only a function of U^* and

T_m° , as can be seen in eq 9a. As a result, $G_{\max}/G_0$ should be a constant, which depends on the polymer specimen. The plots of the reduced growth rate $G/G_{\max}$ in eq 9b vs. the reduced crystallization temperature $T_C/T_{C,\max}$ for a given homopolymer fall on a single master curve without molecular weight dependence. In further, a universal master curve is obtained when the ratio of $\ln(G/G_{\max})/\ln(G_{\max}/G_0)$ is plotted against $T_C/T_{C,\max}$ for various homopolymers. However, in a comparison of the PLLA crystallization in the diblock PLLA-MePEG samples with the pure PLLA, though the crystalline component is the same, our results show that the ratio of $G_{\max}/G_0$ varies with the PLLA composition. This is not surprising since the crystallization of PLLA is greatly influenced by the presence of the connected MePEG component, both U^* and T_m° are strongly dependent of the PLLA composition. In addition, the ratios of $T_{C,\max}/T_m^\circ$ in the diblocks are slightly higher than those in the PLLA homopolymer (as listed in Table 5). Also the T_g value of PLLA for each sample is fixed at 48.3 °C, and the ratio T_∞/T_m° is therefore not a constant. Although these factors have caused the variance of $G_{\max}/G_0$ with the PLLA composition, all the crystallization growth rate results in terms of $G/G_{\max}$ vs. $T_C/T_{C,\max}$ still show somewhat a master curve.

According to eq 7, the analysis of the lateral and fold surface free energy product ($\sigma\sigma_e$) requires the information of K_g , the monomolecular thickness b_0 , the enthalpy of fusion per unit volume of crystal Δh_f° , and T_m° . With the average unit-cell parameters obtained for PLLA crystals from the WAXD analysis in Table 3, the value of b_0 , which corresponds to the interplanar spacing of the growth front planes (110), is calculated and listed in Table 6. As expected, these b_0 values are almost the same and in an excellent agreement with the reported value 5.30 Å [43]. Δh_f° for PLLA crystals is equal to 111.083×10^6 J/m³ [44]. The values of $\sigma\sigma_e$ for each sample are then calculated by eq 7 and listed in Table 6. It is clear that $\sigma\sigma_e$ calculated by inserting the values of K_g for regime II and III, respectively, should be very close. We therefore employ the mean values $\sigma\sigma_e$ in the following further analysis.

In general, the lateral surface free energy (σ) can be estimated by the Thomas-Stavely equation [38],

$$\sigma = \beta b_0 \Delta h_f^\circ \quad (25)$$

where β is a constant, and ranges between 0.1 and 0.3 [38]. For high melting polyesters such as PLLA, $\beta = 0.25$ [25]. As expected, the lateral surface free energy σ listed in Table 6 is almost the same for pure PLLA as well as PLLA in the diblocks. However, σ_e calculated from the previously determined values of σ and $\sigma\sigma_e$ is quite different between pure PLLA and diblock copolymers, as presented in Table 6. For pure PLLA, σ_e is estimated as 8.84×10^{-2} J/m², which is slightly higher than the reported values in the literature [21,25]. This difference is primarily due to the fact that our K_g values from the previous kinetic analysis deviate slightly from those in the literature. In a comparison of σ_e between pure PLLA and diblock copolymers, we observe that σ_e is significantly reduced for PLLA crystallized in the diblock copolymers, suggesting that the presence of amorphous MePEG blocks may help the PLLA chains to fold along the surface, and thus increase the degree of chain folding in the diblocks. As a result, the formation of thinner lamellar crystals with lower melting temperatures is expected for PLLA crystallized in the presence of connected MePEG blocks. Although we have shown a significant reduction of the fold surface free energy σ_e for PLLA crystallized with MePEG block connected to it, the chain folding type of the PLLA in the diblock copolymers, however, needs to be clarified. Also, we are currently in the progress of examining the effects of the connectivity between PLLA and MePEG on the resulting PLLA lamellar crystal thickness by SAXS, which we believe may correlate with the sharp decreasing

behavior of $T_{m, PLLA}^{\circ}$ based on the DSC analysis.

(II) Quantification of the Flory Interaction Parameters

Monte Carlo Analysis

單體建構

系統建構包含了高分子系統中的 PLLA 和 MePEG 單體，與溶劑分子 acetone、tetrahydrofuran(THF)和 N,N-dimethylformamide(DMF)等共五種。藉由軟體建立出 PLLA 和 MePEG 單體結構，與溶劑分子 acetone、THF 和 DMF 的分子結構。利用 Molecular Mechanic，MM 的方式進行各單體的結構最佳化。

交互作用力計算

系統交互作用力計算以蒙地卡羅法(Monte Carlo, MC)，計算各單體/溶劑分子彼此間的交互作用力及配位數關係。起始結構分別為前項工作所建立 PLLA、MePEG 的單體 A、B 與溶劑 acetone、THF 和 DMF 的分子 C1、C2 及 C3。

A、B、C1、C2 和 C3 五者依次以自身為中心，將自身及另四種分子分別以 MC 的方式隨機在自身周圍放置多次，並統計每次放置時該位置兩者間的作用力及能量 E_{ij} 。所以對應 A 粒子會有 E_{AA} 、 E_{AB} 、 E_{AC1} 、 E_{AC2} 和 E_{AC3} 等五種作用關係的能量分佈。同理 B、C1、C2 和 C3 亦各自有對應的五種作用關係能量分佈。

接著同樣以 MC 的方式，依次以自身為中心，將自身及另四種分子隨機在自身周圍以最大可能數量包圍放置多次，可計算得到各粒子間的配位數 Z_{ij} 的分佈。如對應 A 粒子會有 Z_{AA} 、 Z_{AB} 、 Z_{AC1} 、 Z_{AC2} 和 Z_{AC3} 等五種配位數的分佈。再由 E_{ij} 與 Z_{ij} 可求出 $E_{mix}(T)$ ，由 $E_{mix}(T)$ 求出 Flory-Huggins 參數 χ_{ij} ，結果如 Table 7 所示，與文獻[45-47]中的交互作用力值相當吻合。

(III) Micelle Formation of PLLA-MePEG in the Presence of a Selective Solvent

Dissipative Particle Dynamics

高分子/溶劑組合系統介觀相行為模擬

高分子/溶劑分子的組成系統，導入分散粒子動力計算法(Dissipative Particle Dynamics, DPD)模擬系統的介觀相行為。首先將各個分子結構上各個不同功能表現的區域分別設定為不同的 DPD 粒子。在 PLLA-MePEG 系統中，分別將 PLLA 鏈段及 MePEG 鏈段分別以 A、B 兩種粒子的組成表示，而溶液部分的 acetone、THF 和 DMF 以 C1、C2 及 C3 表示。

接著設定各粒子間的交互作用參數 a_{ij} ，這裡所需的交互作用參數 a_{ij} 可由前部分系統交互作用力所求出的 Flory-Huggins 參數 χ_{ij} 求得。最後再設定粒子的分散力參數，即可進行模擬。依照由 Groot 和 Warren [31] 提出的式子可求得在指定的 DPD 粒子密度($\rho=N/V$)下的參數 a_{ij} 。

針對 PLLA-MePEG 系統，預定改變的變因包括(i)不同的 PLLA-MePEG 分子 $A_M B_N$ 的鏈長比例 M/N ；(ii)不同的溫度；(iii)不同濃度的溶劑，即 C1、C2 及 C3 與 $A_M B_N$ 的粒子數比例等三類。其中(iii)溫度效應會直接影響各粒子間的交互作用力 F^C ，所以對應不同的溫度 DPD 模擬會有不同的交互作用參數 a_{ij} 。由模擬結果可以看出在不同的分子鏈長比例，溫度及溶劑濃度作用下所形成的 PLLA-MePEG 結構變化。

首先我們利用分散粒子動力學來探討一般情況的塊狀共聚物添加溶劑之微結構衍變。當我們在探討塊狀共聚物的相行為時， χ_{AB} 與 N 的乘積才是兩鏈段間不相容性的決定因素。在本節研究中，為了節省計算時間，我們縮小 AB 對稱塊狀共聚物的聚合度至 $N=10$ ，也就是 $A_5 B_5$ ，設定 DPD 系統大小為 $15 \times 15 \times 15$ ，其粒子密度 $\rho=3$ ，利用式 13a 得到相同成分粒子間的作用力參數： $a_{AA}=a_{BB}=25$ ，而 A 與 B 成分間的作用力參數： $a_{AB}=51.3$ ，如 Table 8。由 Groot 所提出的 $(\chi_{AB} N)_{eff}$ 公式

$$(\chi^N)_{eff} = \frac{1}{1+3.9N^{2/3-2\nu}} \chi^N \approx \frac{\chi^N}{1+3.9N^{-0.51}} \quad (26)$$

得到在 $\rho=3$ ， $N=10$ 以及 $a_{AB}=51.3$ 時， $(\chi_{AB}N)_{eff} \sim 40$ 。我們在此系統中添加溶劑S，溶劑S對B成分的作用力參數 $a_{BS}=25$ 。

為了探討溶劑選擇性(a_{AS})與共聚合物體積分率的關係，我們所得到的相圖如Figure 12。當我們添加的是中性溶劑時($a_{AS}=a_{BS}=25$)，隨著共聚合物體積分率下降，微結構由層狀(L)→聚集狀態→無序狀態(D)，如Figure 13。當 $\phi>0.58$ 時，微結構為層狀。隨著共聚合物體積分率下降，我們發現當 $0.58>\phi>0.45$ ，雖然共聚合物有聚集的現象，但是並未形成微結構。此結果定性上與自洽平均場預測的結果相同，然而在定量上有很大的差異。因為分散粒子動力學的模擬計算中有包含熱擾動，Fredrickson 與Leibler[48]以及Olvera de la Cruz[49]考慮熱擾動的情況下，提出塊狀共聚合物在中性溶劑下：

$$(\phi^{1.59} \chi_{AB} N)_{ODT} = F(f) \quad (27)$$

經由此式得 $\phi_{ODT} \sim 0.43$ 。因此我們的結果較符合式 27。

當添加的是選擇性溶劑($a_{AS}>a_{BS}$)，我們發現隨著溶劑的添加(ϕ 下降)，微結構將由層狀→穿孔層板(PL_A)→六方柱狀堆積(C_A)→球狀微胞→無序狀態。球狀微胞中我們發現有面心立方堆積(FCC)以及無序微胞($D^{micelle}$)兩種狀態。首先我們觀察當溶劑選擇性相對較低($a_{AS}=35.0$)的系統，如Figure 14。此系統隨著共聚合物體積分率下降，微結構由層狀→穿孔層板→六方柱狀堆積→無序微胞。此系統中有趣的地方為層狀與六方柱狀堆積間穿孔層板結構的出現，此相在實驗[50,51,52]系統中已經出現。我們使用與Groot相同的聚合度 N ，利用分散粒子動力學來觀察熔融態塊狀共聚合物的相變化，他們在層狀與六方柱狀堆積間只有發現穿孔層板結構[53]。雖然自洽平均場計算中，預測了gyroid相為熔融態塊狀共聚合物層狀與六方柱狀堆積間最穩定的結構[54]。然而，若將我們的系統視為共聚合物添加一鏈段長度為其 1/10 的均聚物，自洽平均場在共聚合物添加均聚物的系統中，亦預測了隨著均聚物的添加gyroid相會衍變為穿孔層板結構[55]。目前理論計算在塊狀共聚合物添加溶劑的系統，對於層狀與六方柱狀堆積之間穩定結構的探討，尚付之闕如。在六方柱狀堆積與無序微胞間，自洽平均場計算預測體心立方球狀堆積的存在，然而在此系統中，我們的分散粒子動力學模擬並沒有發現球狀微胞排列的穩定結構，可能原因與接近無序狀態時的熱擾動效應有關。

另一方面，若將溶劑選擇性增大至 $a_{AS}=50.0$ ，我們發現球狀微結構出現面心立方堆積的相區，隨著共聚合物體積分率下降，微結構由層狀→穿孔層板→六方柱狀堆積→面心立方球狀堆積→無序微胞，如Figure 15。我們的自洽平均場計算中，預測共聚合物添加強選擇性溶劑，隨著共聚合物體積分率下降，微結構由層狀→穿孔層板→六方柱狀堆積→體心立方球狀堆積→面心立方球狀堆積→無序狀態。若將我們的系統視為共聚合物添加一鏈段長度為其 1/10 的均聚物，自洽平均場亦預測隨著均聚物體積分率增加，微結構由體心衍變為面心球狀堆積[55]。然而，我們的分散粒子動力學模擬中，並沒有發現體心立方球狀堆積。在上一節中，我們探討了共聚合物添加選擇性溶劑體心與面心立方球狀堆積穩定存在的因素，當共聚合物微胞外的B鏈段因為溶劑的膨潤效應而增長，導致微胞間的B鏈段重疊程度過高，因此微胞間的排列將會選擇較不緊密的體心立方堆積。我們選擇面心立方球狀堆積中，共聚合物體積分率最高的系統，改變溶劑選擇性大小，觀察球狀微胞間B鏈段的交錯情形，如Figure 16。在Figure 16 中，球狀微胞間B鏈段交錯比例不高，可將共聚合物微胞視為硬球，因此其排列將會採用面心立方球狀堆積。同時，隨著溶劑選擇性的降低，B鏈段交錯的比例逐漸增加，我們認為當溶劑選擇性低時，可能出現體心立方球狀堆積，然而因為接近無序區域，熱擾動的效應使得體心立方球狀堆積的區域減少。

Conclusions (結論)

We employ WAXD and DSC experiments to analyze the crystal unit-cell structures and the isothermal crystallization kinetics of PLLA in biodegradable PLLA-MePEG diblock copolymers. In particular, the effects due to the presence of MePEG that is chemically connected to PLLA as well as the PLLA crystallization temperature T_C are examined. We then use DPD to simulate the effects of solvent selectivity and copolymer volume fraction on the microstructural type of block copolymers.

From the WAXD analysis, both the PLLA and MePEG crystal unit-cell structures are not affected by the chain-connectivity as well as the block ratio between PLLA and MePEG. These structural parameters are also independent of T_C , as expected. With the aid of DSC, we observe that the equilibrium melting temperature of PLLA in the diblock samples decreases dramatically, indicating that this decreasing behavior is not simply due to the decrease in the PLLA molecular weight. The dilution effect caused by the miscible MePEG block connected to PLLA has a greater influence on the crystallization as well as the melting behavior of PLLA. As described by the crystallization process with a nucleation and growth, both the overall crystallinity and the isothermal crystallization growth rate G of PLLA first increase with T_C and then show a decreasing behavior with a further increase in T_C . In further, based on the Hoffman-Lauritzen theory, all the experimental data of G vs. T_C for PLLA crystallized in each sample are well described by two bell-shaped curves, which are mainly attributed to the crystallization regimes III and II, respectively. Similar to the behavior of $T_{m, PLLA}^\circ$, the transition temperature from regimes III to II, $T_{C, III \rightarrow II}$, as well as $T_{C, max}$, at which the growth rate reaches a maximum G_{max} , in both regimes show a decreasing behavior too. In further, all the plots of G / G_{max} vs. $T_C / T_{C, max}$ for each sample fall on a single master curve. This phenomenon is similar to that observed for a homopolymer with various molecular weights. However, unlike the behavior of $G_{max} / G_\circ$ remaining almost a constant for a homopolymer without molecular weight dependence, we observe that $G_{max} / G_\circ$ varies with each sample. This is not surprising since the crystallization of PLLA is greatly influenced by the presence of the connected MePEG component, two of the key factors which have a great influence on $G_{max} / G_\circ$, the activation energy for the segment diffusing to the crystallization site U^* and T_m° , are strongly dependent of the PLLA composition. Similar to T_m° , we find that U^* also decreases with the PLLA molecular weight and/or composition, suggesting that the presence of the amorphous connected MePEG indeed promotes the ability for the PLLA chains to diffuse to the crystallization sites. Also, it helps the PLLA chains to fold along the surface, and thus the fold surface free energy σ_e is significantly reduced for PLLA crystallized in the diblock copolymers.

我們同時利用分散粒子動力學模擬塊狀共聚物添加溶劑的系統。添加中性溶劑時，我們的結果與文獻的式子相符。當添加選擇性溶劑時，我們發現在層狀與六方柱狀堆積之間出現穿孔層板微結構。當溶劑添加量大時，出現球形微胞堆積成面心立方堆積的穩定區域，分析其 B 鏈段交錯比例不高，符合自洽平均場對於體心與面心立方球狀堆積的定性分析。

計畫成果自評

目前我們已具體完成了二年期計畫的第一部份，內容主要在分析生物醫學用高分子 PLLA-MePEG 共聚物，其結晶行為及結構尺寸如何受到鏈段組成以及溫度的影響；相關的實驗結果及分析已投稿在 Polymer 期刊，目前在 revised 的階段。

至於第二部份，我們首先探討一般共聚物其微結構型態如何受到共聚物的組成、溶劑選擇性、以及溶劑的濃度之影響，這一部份結合 DPD 與自洽平均場(SCMF)理論計算的結果，目前正在整理並計畫投稿至 Macromolecules 期刊。由於先前我們發現擴展 DPD 至稀薄溶液下之微胞形成結果，與文獻中實驗觀察結果有一些落差，目前我們還在努力探討其原因，

我們期許能定量分析稀薄的 PLLA-PEG 共聚合物在不同溶劑下的微胞形成，並與目前文獻中的實驗結果比較，最後我們希望將藥物引進至微胞中，再分析其藥物釋放的機制。

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Table 1. Sample Characteristics of PLLA Homopolymer and PLLA–MePEG Diblock Copolymers

Samples	L-Lactide		N_{PLLA}	N_{MePEG}	f_{PLLA}	M_n	Polydispersity ^a
	Content in the Feed(wt%)						
PLLA	100		283		1	20400 ^a	1.6
S1	74		181	114	0.61	18000	1.1
S2	58		86	114	0.43	11200	1.1
S3	48		56	114	0.33	9000	1.2

^aDetermined by GPC.

Table 2. Typical Reflection Planes and Corresponding θ Values Used in the Deconvolution Procedures for (a) PLLA and (b) MePEG, respectively

(a)

θ	h	k	l
16.7	2	0	0
16.8	1	1	0
19.1	2	0	3
25.0	1	1	6
29.0	2	1	6

(b)

θ	h	k	l
14.4	0	2	1
18.7	1	2	0
22.5	1	1	2
22.9	0	3	2
25.8	0	2	4
26.3	1	3	1
27.3	1	1	3
30.0	2	0	1
32.4	1	1	4
34.6	1	4	2
35.7	1	2	-7
38.8	2	0	3

Table 3. Unit-Cell Structure Parameters of both PLLA and MePEG Crystals Obtained from the Diblock Copolymer Samples and Pure PLLA and MePEG after Complete Crystallization of PLLA at T_c and MePEG at 25 °C

Samples	T_c (°C)	PLLA			MePEG		
		a(Å)	b(Å)	c(Å)	a(Å)	b(Å)	c(Å)
PLLA	70	10.58	6.08	28.64			
	80	10.57	6.06	28.99			
	90	10.56	6.04	28.92			
	100	10.61	6.07	28.96			
	110	10.54	6.04	28.94			
	120	10.60	6.07	28.93			
	130	10.43	5.98	28.89			
	average	10.56	6.05	28.90			
S1	90	10.58	6.06	28.61	8.20	12.81	20.56
	100	10.84	6.20	28.68	8.44	13.04	20.83
	110	10.92	6.22	28.30	8.37	13.10	21.08
	120	10.85	6.24	28.59	8.45	12.99	21.12
	130	10.34	5.90	28.88	8.01	12.64	20.22
	average	10.71	6.12	28.61	8.29	12.92	20.76
S2	80	10.84	6.25	28.51	8.39	13.10	20.84
	90	10.36	6.08	28.67	8.26	12.84	20.87
	100	10.69	6.12	27.60	8.20	12.94	20.67
	110	10.68	6.13	28.67	8.32	12.91	21.05
	120	11.10	6.36	26.20	8.57	13.39	21.29
	average	10.73	6.19	27.93	8.35	13.04	20.94
S3	70	10.66	6.10	28.65	8.27	12.86	20.84
	80	10.71	6.15	28.61	8.34	12.89	21.07
	90	11.05	6.38	27.60	8.60	13.21	21.40
	100	10.76	6.15	28.46	8.18	13.14	19.75
	110	10.75	6.16	28.46	8.37	12.88	21.11
	average	10.79	6.18	28.36	8.35	13.00	20.83
MePEG	25				8.24	13.36	18.59

Table 4. Avrami Parameters for Various Samples at T_C

Samples	$T_C(^{\circ}C)$	n	k	Samples	$T_C(^{\circ}C)$	n	k
PLLA	80	3.55	7.45E-10	S2	80	2.60	4.29E-05
	85	3.51	6.41E-09		85	2.75	2.61E-05
	90	3.06	6.49E-07		90	3.07	7.06E-06
	95	3.26	1.49E-06		95	3.23	4.72E-06
	100	3.46	1.42E-06		100	3.31	1.07E-06
	105	3.47	1.38E-06		105	3.36	5.23E-07
	110	3.45	1.67E-06		110	3.24	1.01E-07
	115	2.82	8.86E-06		115	3.08	3.85E-08
	120	3.51	4.90E-08		120	3.11	1.33E-09
	125	3.59	1.45E-08				
	130	3.51	7.70E-09				
S1	90	2.90	5.02E-05	S3	70	3.01	9.14E-06
	95	2.74	1.91E-04		75	2.51	1.10E-05
	100	3.32	1.35E-05		80	3.16	4.56E-06
	105	3.35	6.69E-06		85	3.57	4.24E-07
	110	3.57	1.32E-06		90	3.72	9.46E-08
	115	3.17	1.51E-06		95	3.97	1.45E-08
	120	2.87	1.22E-06		100	5.04	1.22E-11
	125	3.20	2.64E-08		105	5.67	1.49E-14
	130	3.15	2.09E-09		110	5.13	6.29E-15

Table 5. Crystal Growth Kinetic Data of PLLA in Various Samples

Sample	U* (cal/mol)	T _m ^o (°C)	T _{C, III→II} (°C)	Regime III					Regime II				
				K _g (10 ⁵ K ²)	ln G _o (sec ⁻¹)	G _{max} (sec ⁻¹)	T _{C,max} (°C)	$\frac{T_{C,max}}{T_m^o}$	K _g (10 ⁵ K ²)	ln G _o (sec ⁻¹)	G _{max} (sec ⁻¹)	T _{C,max} (°C)	$\frac{T_{C,max}}{T_m^o}$
PLLA ^a	1500	196	120	5.02	21.16				3.0	13.74			
PLLA ^a	2900	196	119	9.1	42.81	0.027	105	0.806	4.58	26.47	0.009	118	0.834
PLLA	2900	196	120	8.25	37.15	0.026	104	0.803	4.15	23.45	0.008	119	0.836
S1	2100	165	109	3.79	25.50	0.051	96	0.843	1.93	16.76	0.020	108	0.870
S2	1900	159	97	4.57	27.95	0.028	88	0.836	2.30	18.05	0.017	99	0.861
S3	1530	147	86	4.71	28.85	0.031	77	0.833	2.34	17.92	0.016	88	0.860

^aDetermined from the plot of $\ln G + \frac{U^*}{R(T_C - T_\infty)}$ vs. $\frac{1}{T_C \Delta T f}$ with the correction term $f = \frac{2T_C}{T_m^o + T_C}$.

Table 6. Values of b_o , $\sigma\sigma_e$, σ , and σ_e for PLLA Crystallized in Various Samples

Sample	b_o (Å)	$\sigma\sigma_e$ (10^{-4} J ² /m ⁴) From regime III	$\sigma\sigma_e$ (10^{-4} J ² /m ⁴) From regime II	mean $\sigma\sigma_e$ (10^{-4} J ² /m ⁴)	σ (10^{-2} J/m ²)	σ_e (10^{-2} J/m ²)
PLLA	5.25	12.86	12.94	12.90	1.46	8.84
S1	5.31	6.22	6.32	6.27	1.47	4.27
S2	5.36	7.56	7.60	7.58	1.49	5.09
S3	5.36	8.05	7.99	8.02	1.49	5.38

Table 7. PLLA 和 MePEG 對各種溶劑的作用力參數 χ (at 370K)

Solvents	PLLA	MePEG
acetone	0.66	0.45
THF	1.37	0.35
DMF	0.24	-0.08

* $\chi_{\text{PLLA-MePEG}} = 1.08(300\text{K}); 0.44(370\text{K})$

Table 8. DPD 模擬的各成份間作用力參數 a

	A	B	S
A	25.0	51.3	25.0-50.0
B	51.3	25.0	25.0
S	25.0-50.0	25.0	25.0

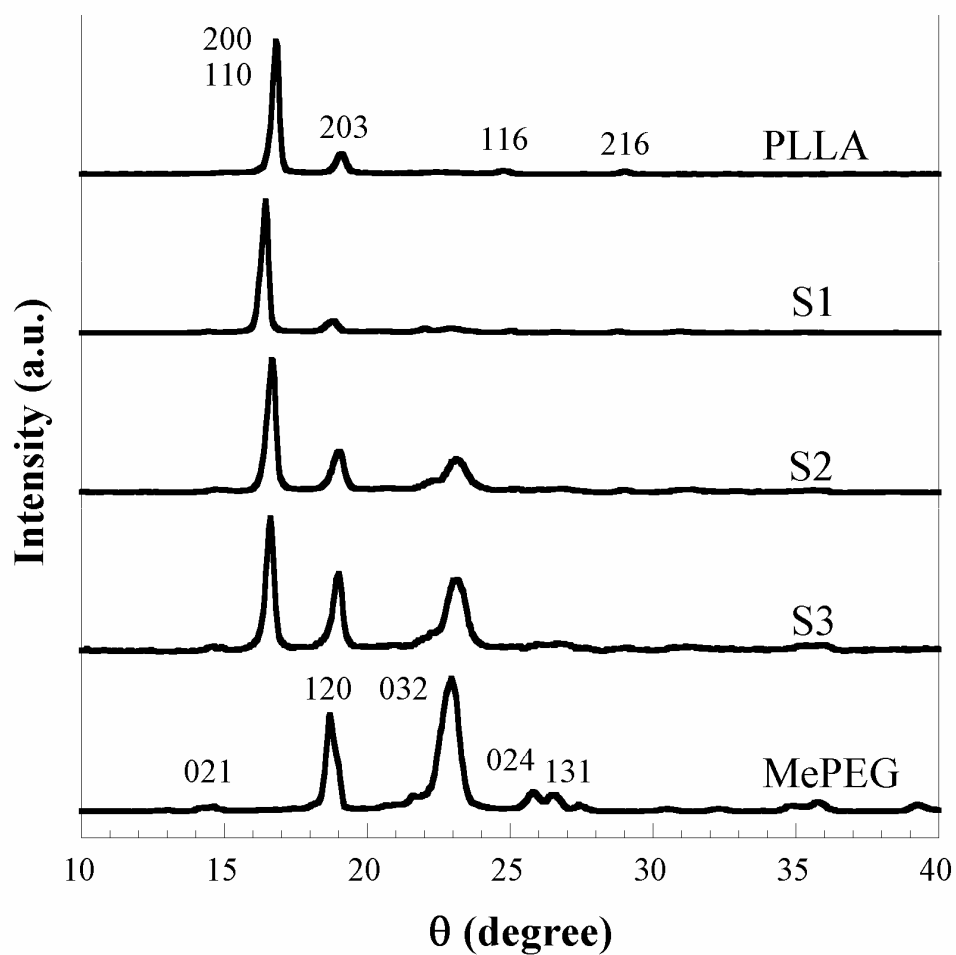


Figure 1. WAXD patterns in terms of the intensity and θ for each diblock sample first isothermally crystallized at $T_C = 100\text{ }^\circ\text{C}$ for PLLA crystallization and then quenched to $25\text{ }^\circ\text{C}$ for MePEG crystallization. In a comparison, the WAXD patterns for pure PLLA and MePEG crystallized at $100\text{ }^\circ\text{C}$ and $25\text{ }^\circ\text{C}$, respectively, are also shown.

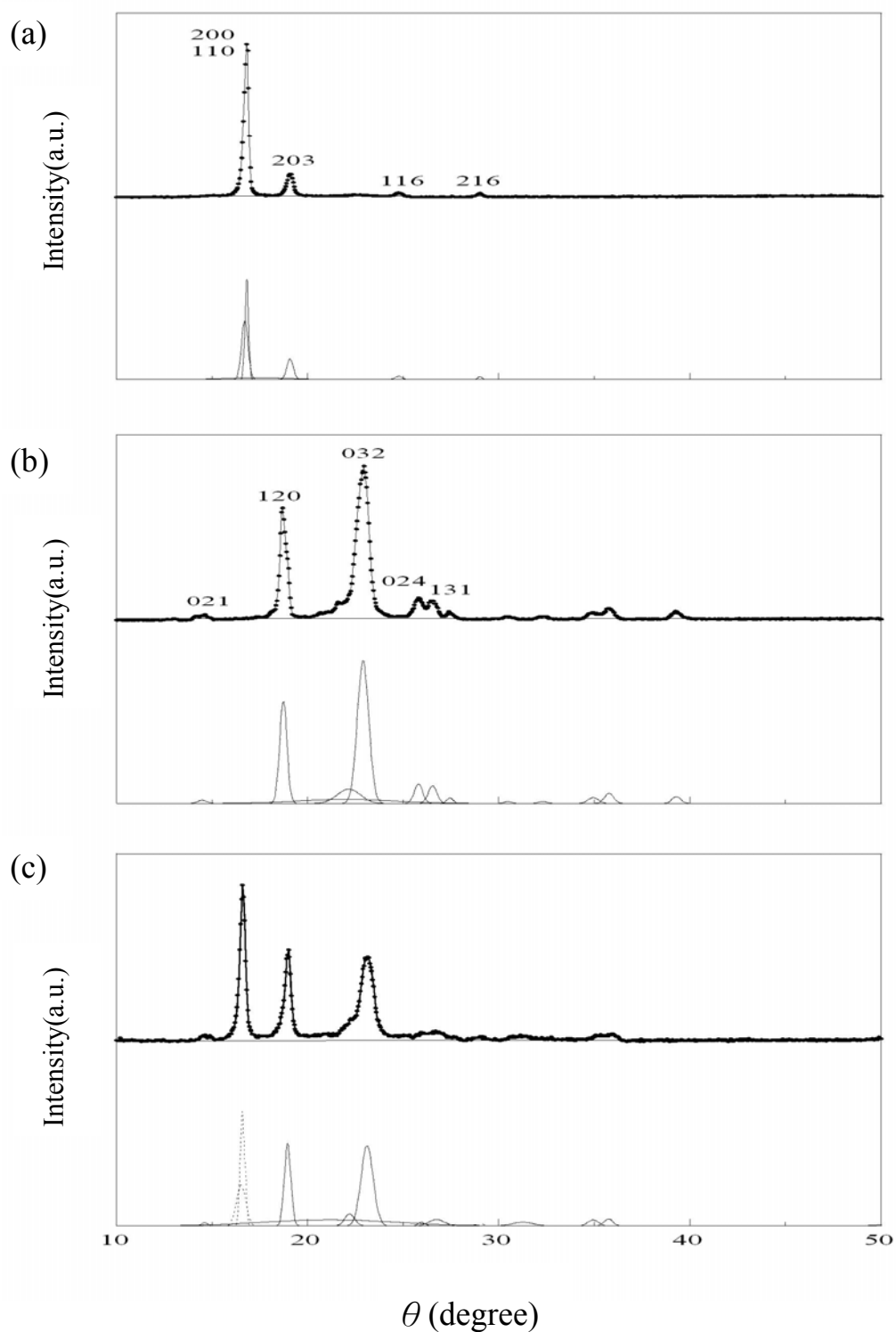
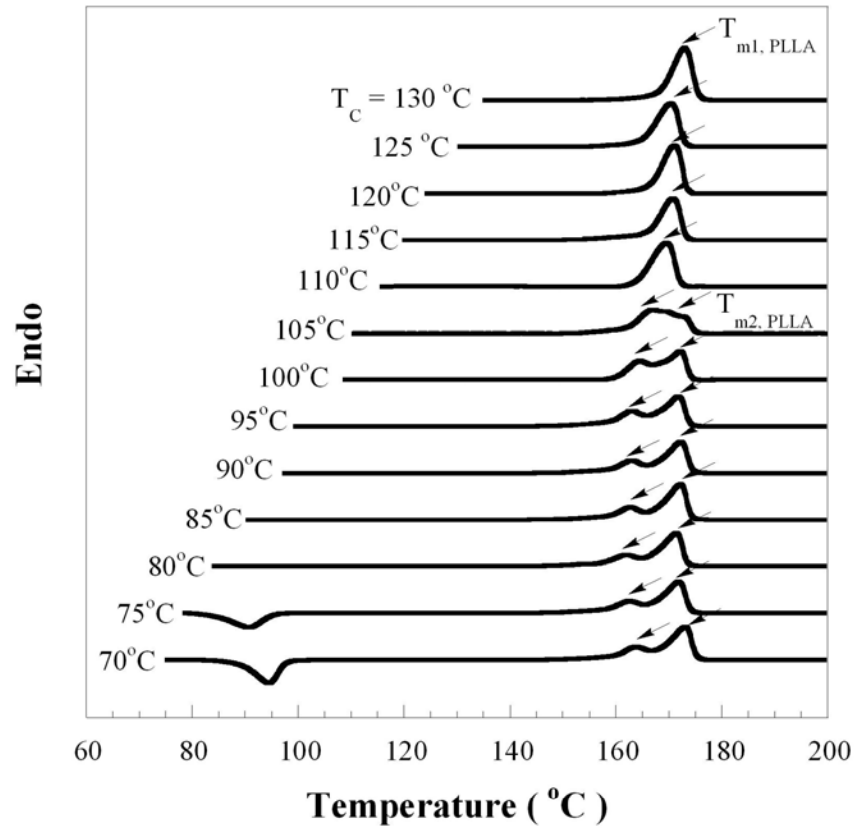


Figure 2. Peak deconvolution of the WAXD profiles for (a) pure PLLA, (b) pure MePEG, and (c) S3 diblock copolymer, in which PLLA and MePEG are crystallized at $T_C = 100^\circ\text{C}$ and 25°C , respectively. In part (c), the peaks presented by the dashed and solid curves correspond to the crystalline reflection planes contributed from PLLA and MePEG, respectively.

(a)



(b)

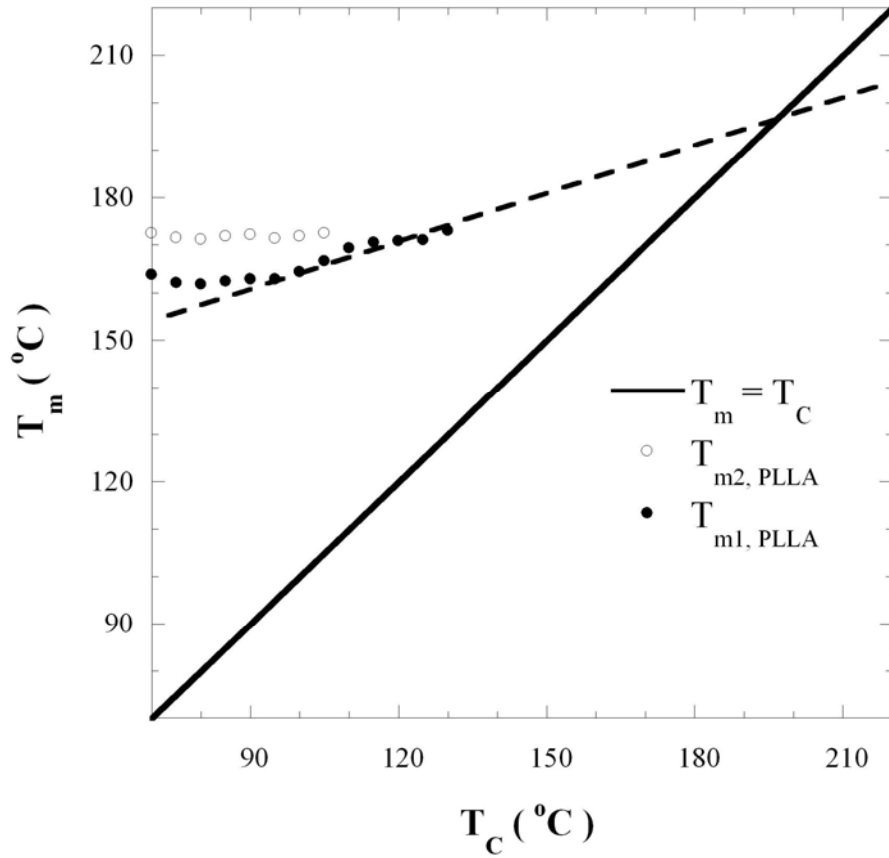


Figure 3. (a) DSC melting endotherms of pure PLLA with a heating rate of $10^\circ\text{C}/\text{min}$ after isothermal crystallization at various temperatures T_c . (b) Variation of $T_{m1, PLLA}$ and $T_{m2, PLLA}$ with T_c for pure PLLA.

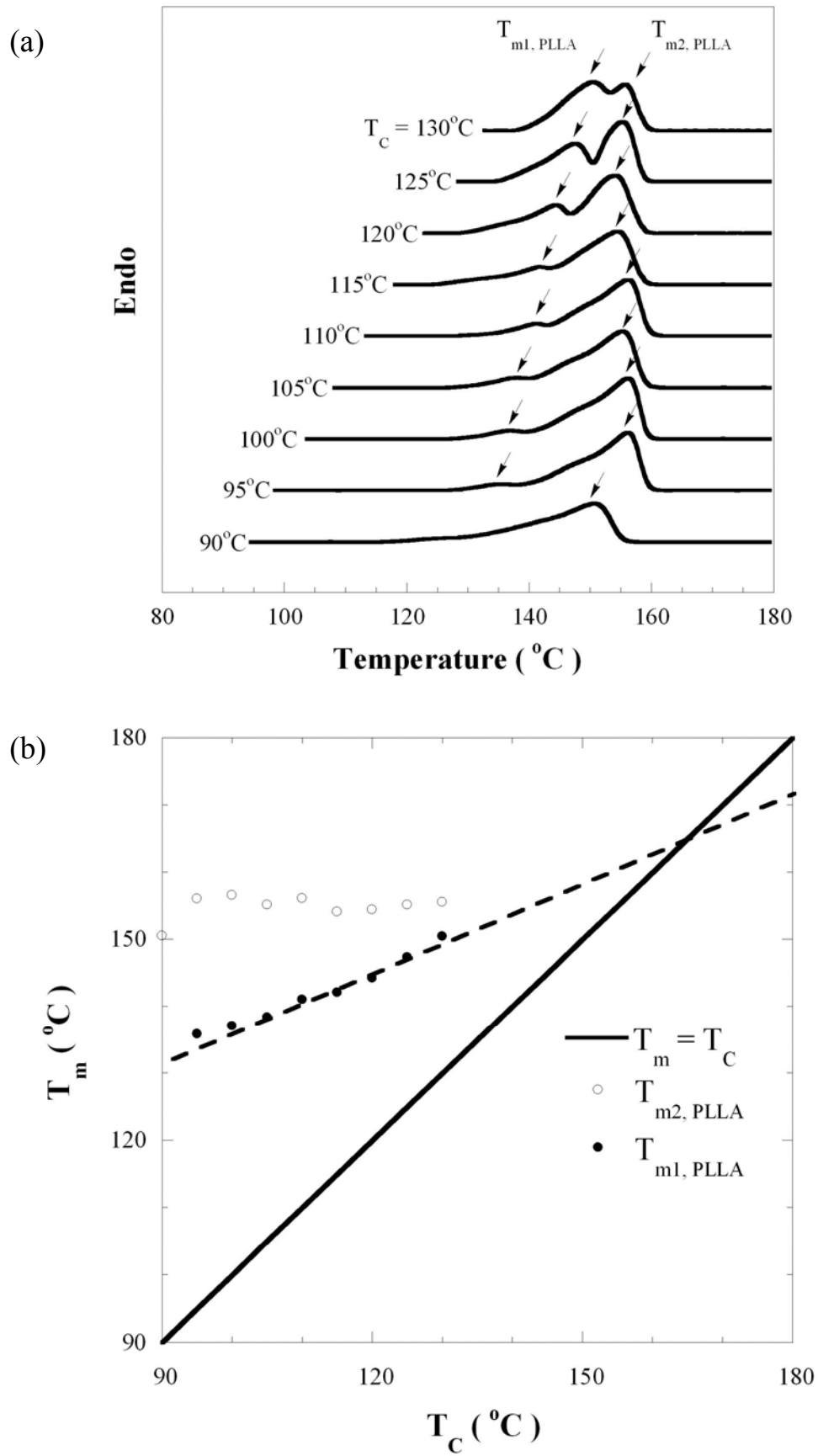
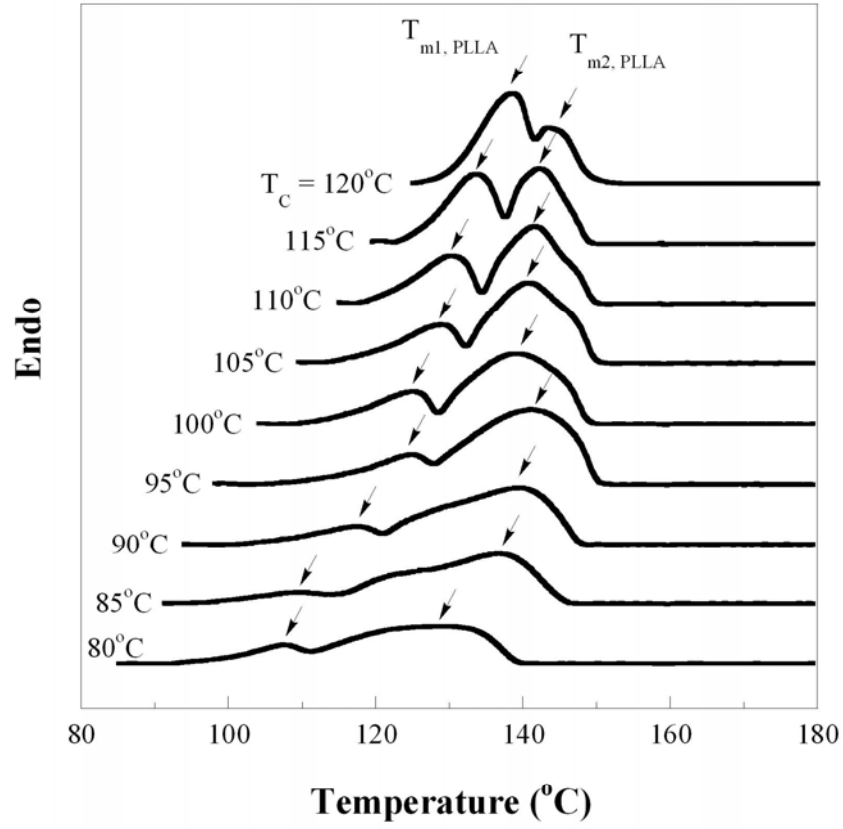


Figure 4. (a) DSC melting endotherms of sample S1 with a heating rate of 10 °C/min after isothermal crystallization at various temperatures T_C . (b) Variation of $T_{m1, PLLA}$ and $T_{m2, PLLA}$ with T_C for sample S1.

(a)



(b)

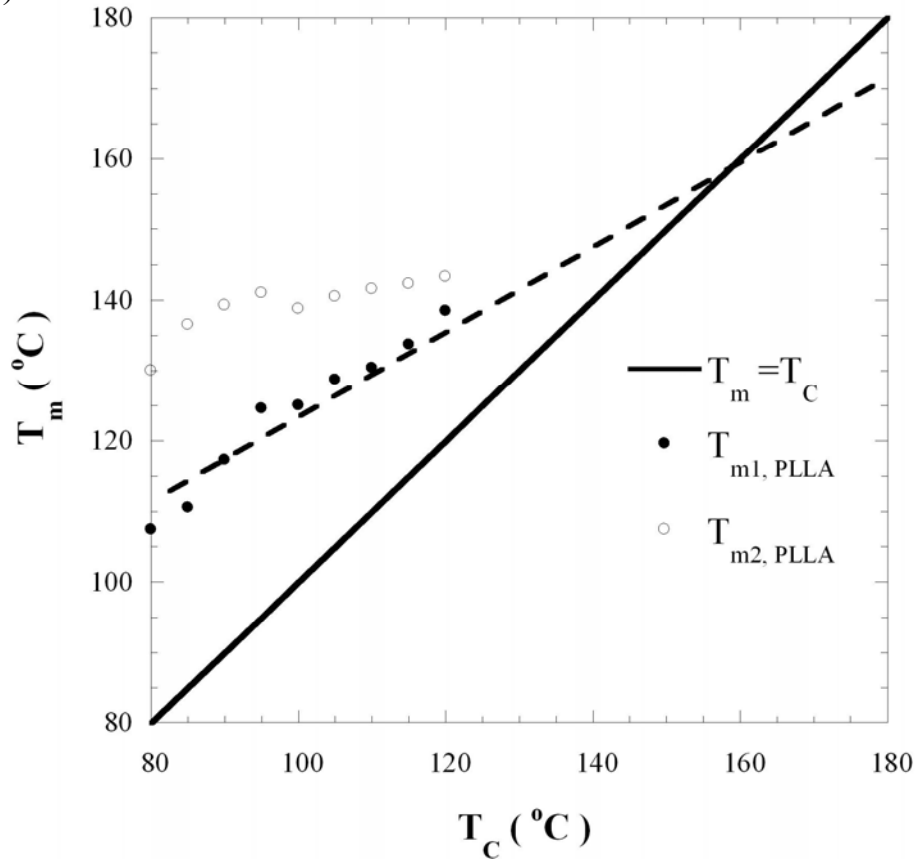


Figure 5. (a) DSC melting endotherms of sample S2 with a heating rate of 10 °C/min after isothermal crystallization at various temperatures T_c . (b) Variation of $T_{m1, PLLA}$ and $T_{m2, PLLA}$ with T_c for sample S2.

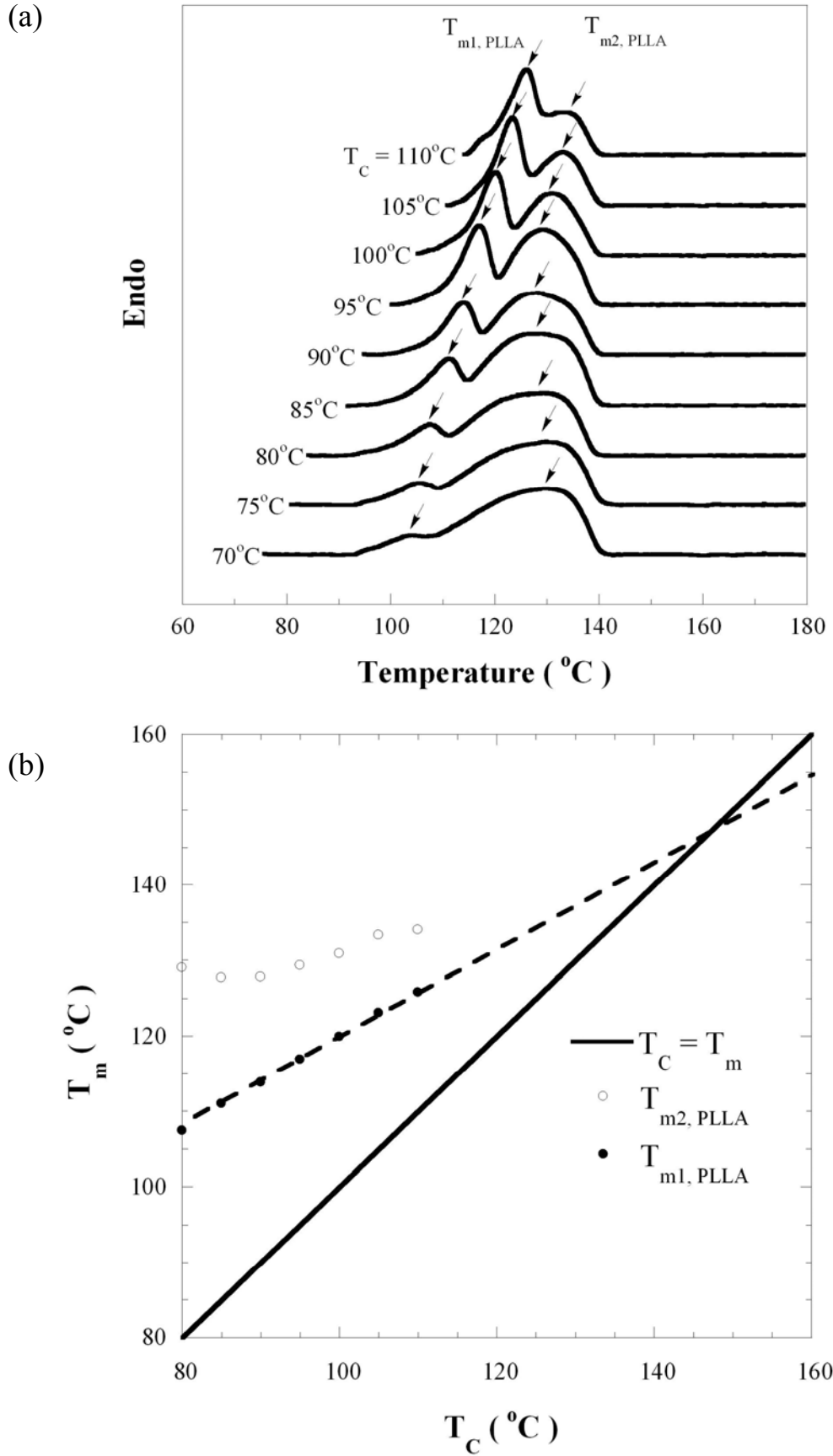


Figure 6. (a) DSC melting endotherms of sample S3 with a heating rate of 10 °C/min after isothermal crystallization at various temperatures T_C . (b) Variation of $T_{m1, PLLA}$ and $T_{m2, PLLA}$ with T_C for sample S3.

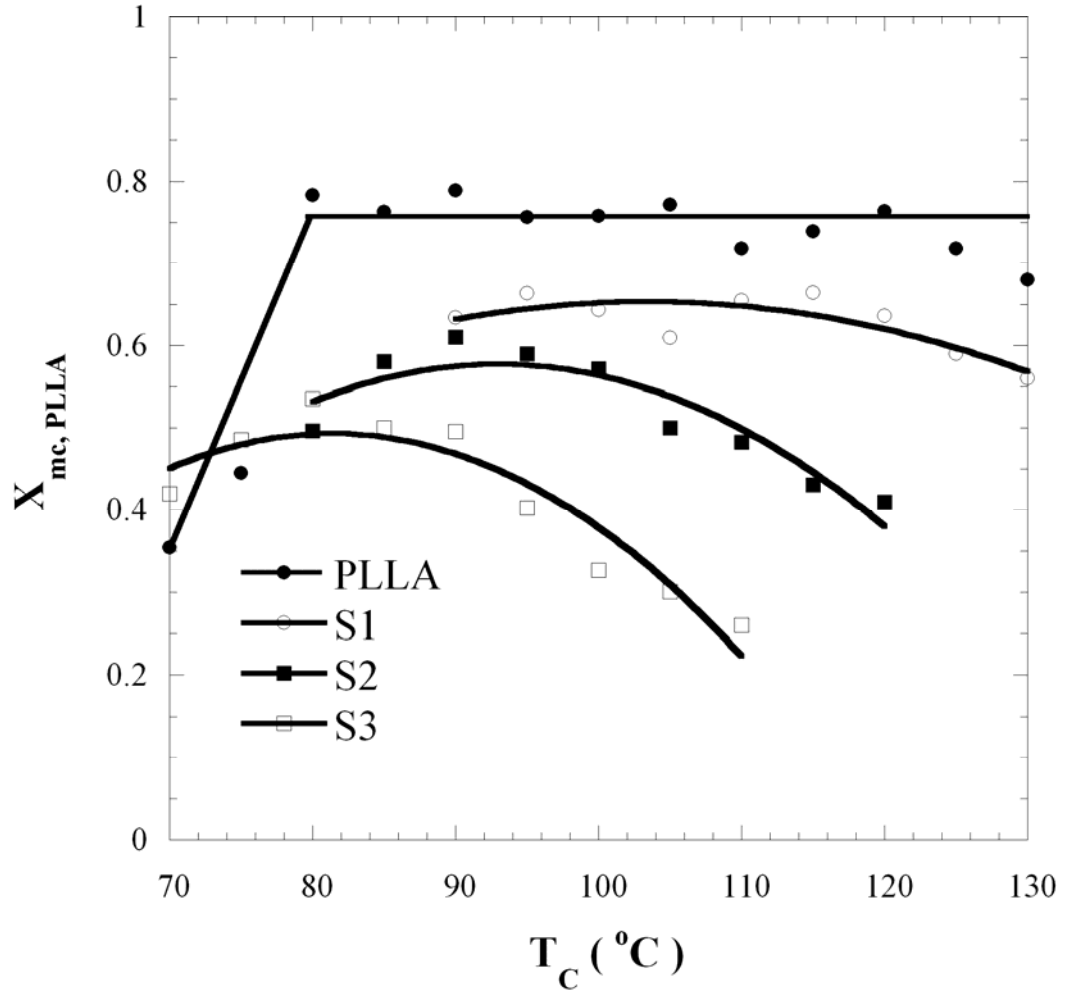


Figure 7. Plot of the normalized crystallinity of PLLA component, $X_{mc, PLLA}$, as a function of T_C for various systems isothermally crystallized at T_C from a homogeneous melt state at 200 °C.

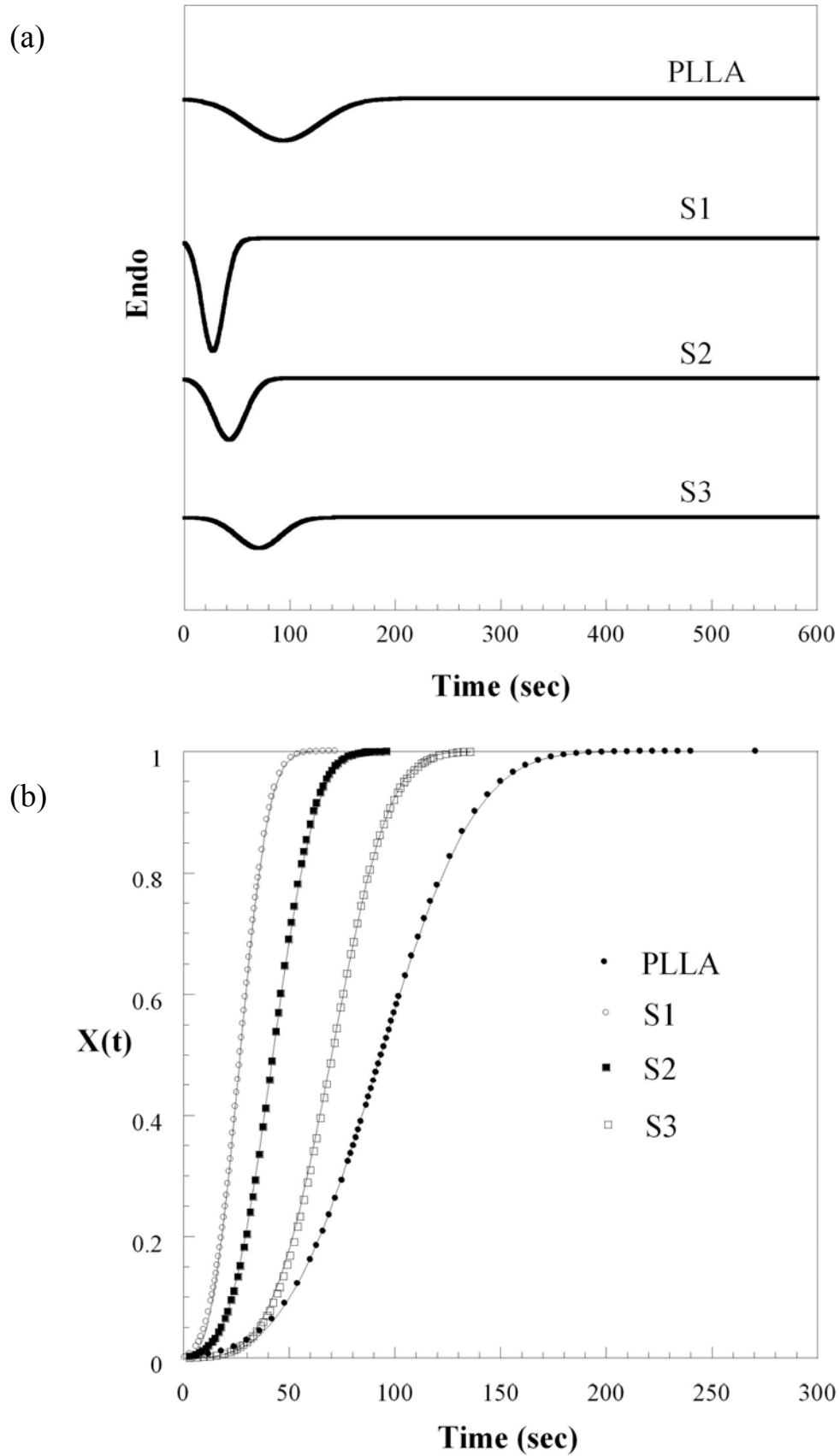
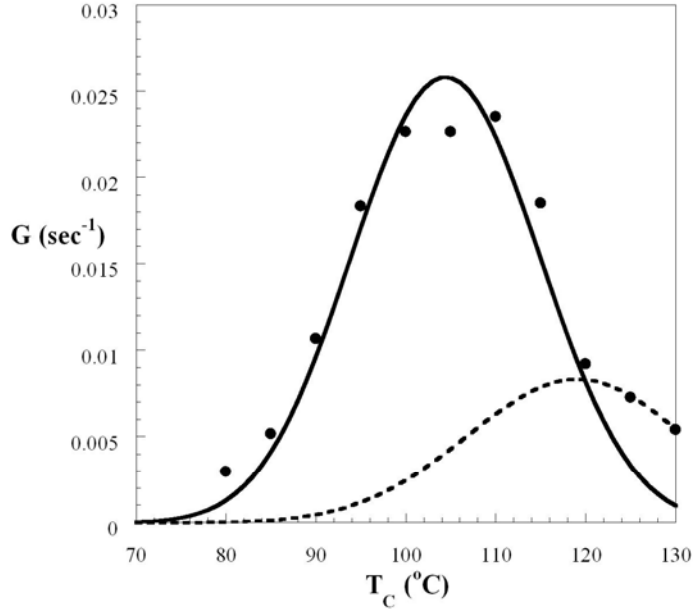
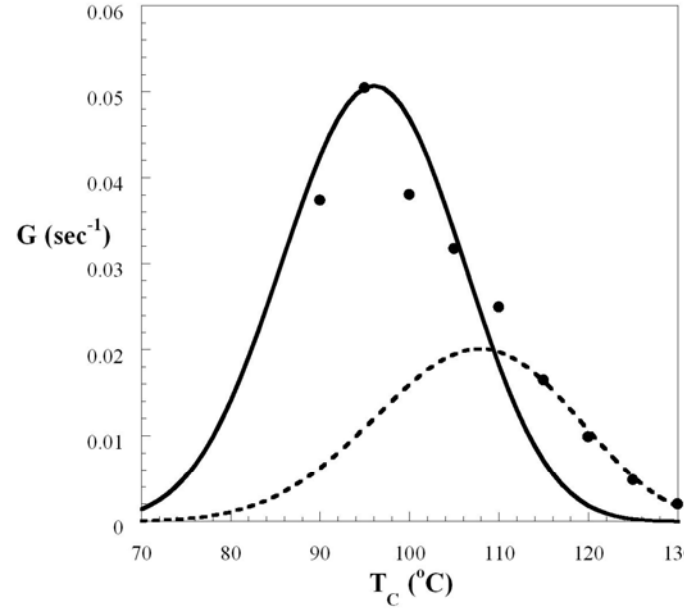


Figure 8. (a) DSC exothermic curves and (b) PLLA relative crystallinity vs crystallization time for various systems undergoing an isothermal crystallization process at $T_C = 90\text{ }^{\circ}\text{C}$. The solid curves in (b) represent the best fitting results from the Avrami equation.

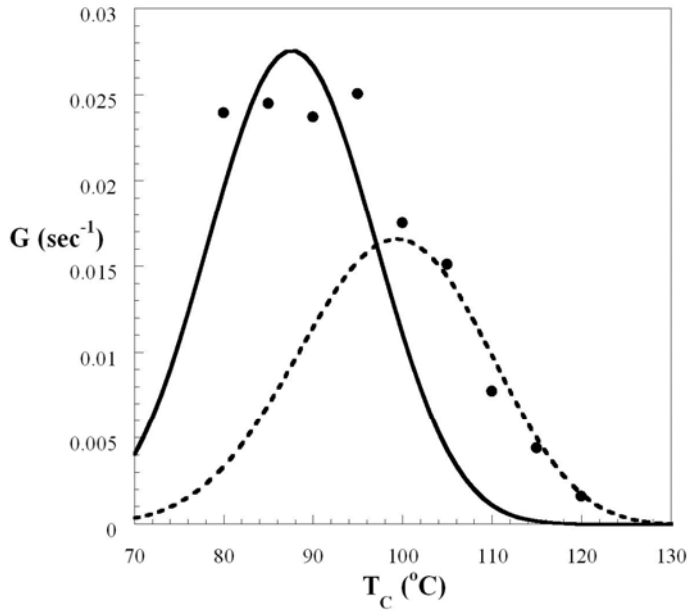
(a)



(b)



(c)



(d)

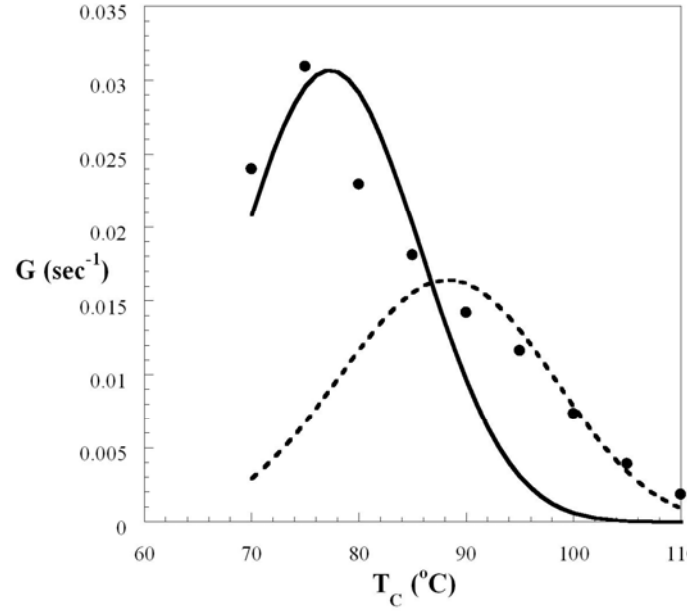


Figure 9. Plot of the PLLA crystallization growth rate G in (a) PLLA, (b) S1, (c) S2, and (d) S3, as a function of crystallization temperature T_c . The solid and dashed curves correspond to the theoretical G values calculated by inserting the best fitting parameters in Table 5 into eq 6 for a given T_c in regimes III and II, respectively.

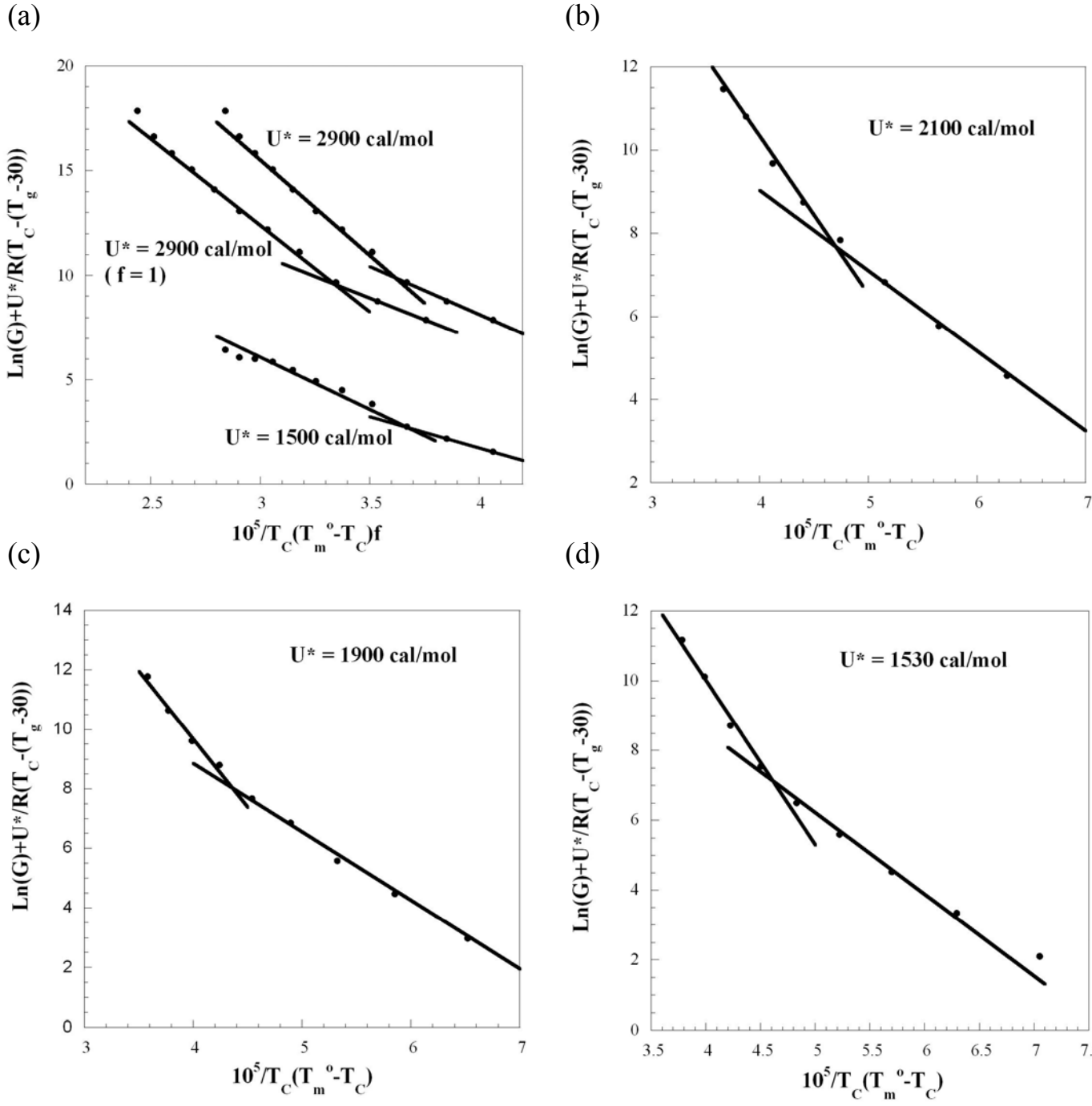


Figure 10. Kinetic Analysis of the crystallization growth rate data for (a) PLLA, (b) S1, (c) S2, and (d) S3, using different values of U^* .

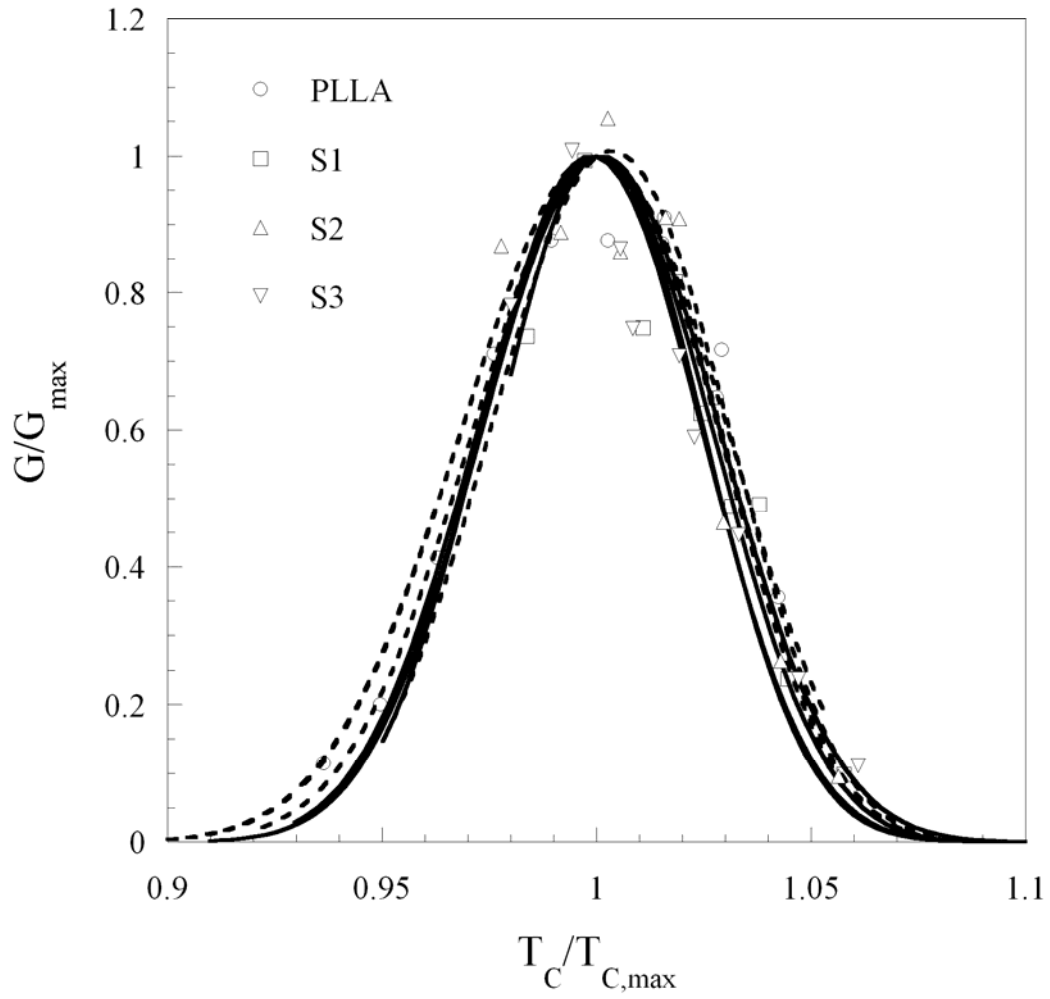


Figure 11. Plot of the reduced crystallization growth rate ($G / G_{\max}$) vs. the reduced crystallization temperature ($T_C / T_{C,\max}$) for various samples. The solid and dashed curves correspond to the calculated results for various samples in regimes III and II, respectively.

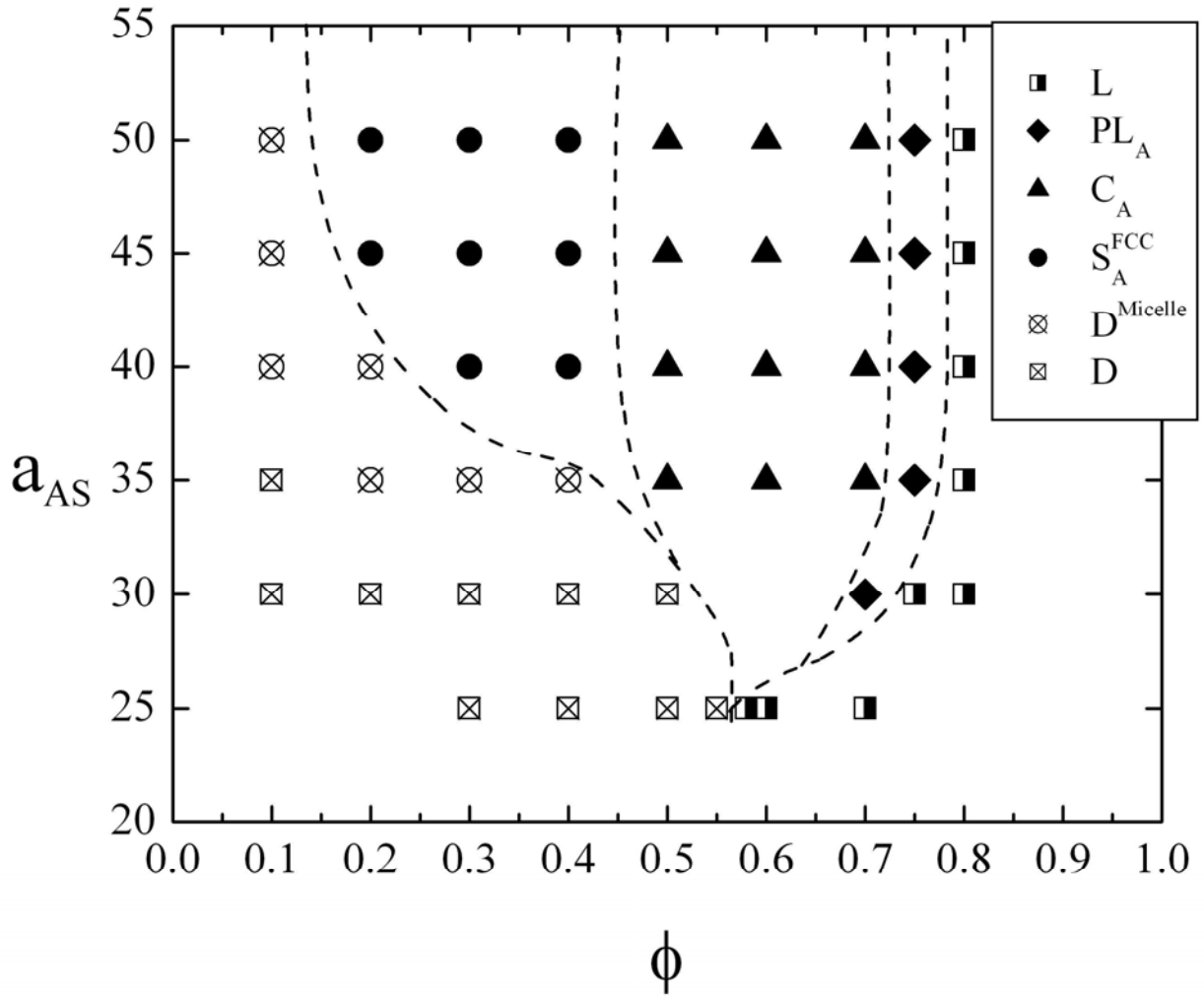


Figure 12. 溶劑選擇性(a_{AS})對AB對稱塊狀共聚物($f=0.5$)體積分率(ϕ)的相圖，固定 $N=10$ 、 $a_{AA}=a_{BB}=a_{BS}=25$ 以及 $a_{AB}=51.3$ 。其中L為層狀，PL為穿孔層板，C為六方柱狀堆積，S為球狀堆積，上標FCC代表面心立方堆積， D^{micelle} 為無序微胞，D為無序狀態。圖中下標A代表微結構是由A成分構成。

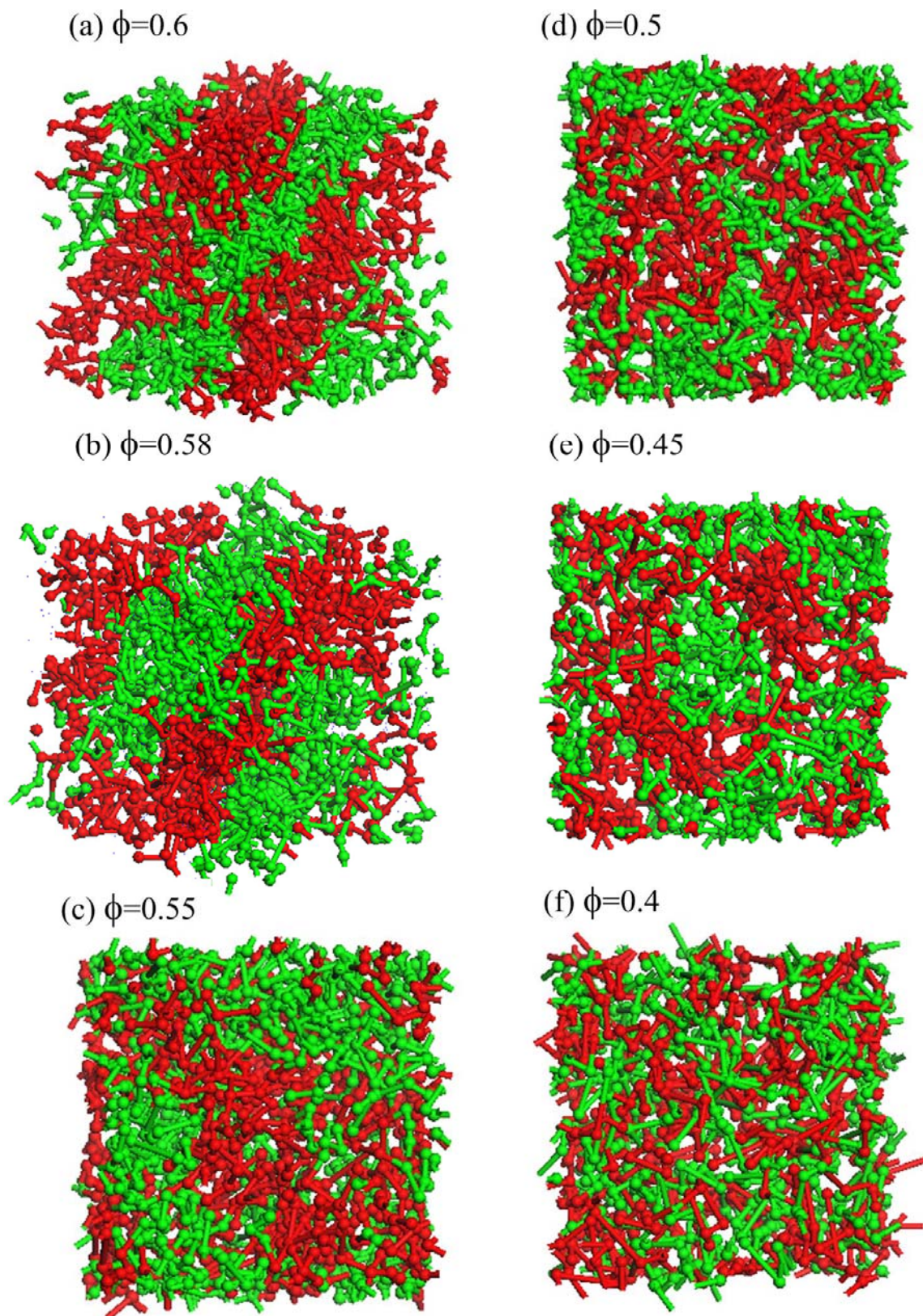


Figure 13. 對稱塊狀共聚合物 $N=10(A_5B_5)$ ，固定 $a_{AA}=a_{BB}=25$ 、 $a_{AB}=51.3$ 添加中性溶劑 $S(a_{AS}=a_{BS}=25)$ ，隨著共聚合物體積分率(ϕ)變化的結構圖。其中，紅球、綠球分別代表A成分以及B成分。

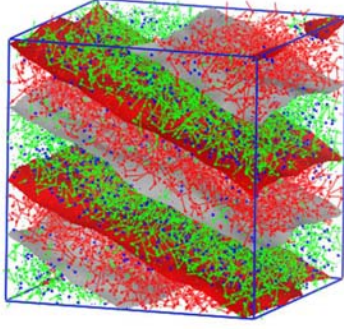
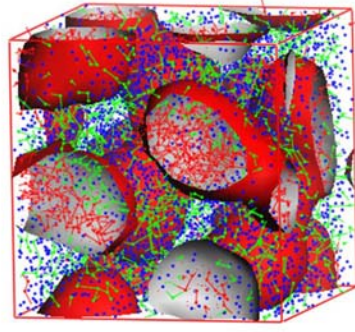
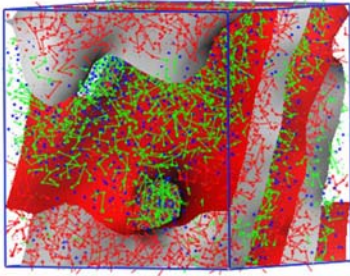
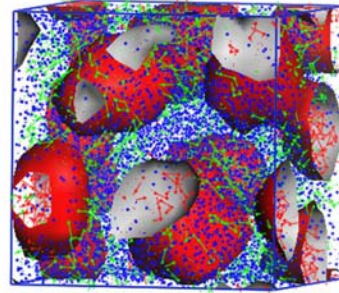
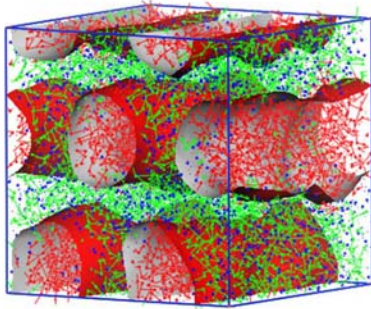
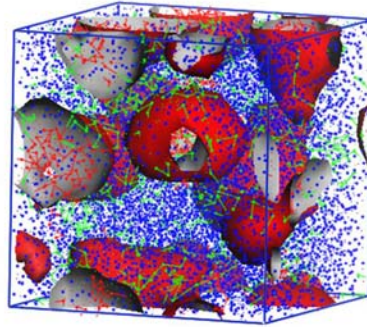
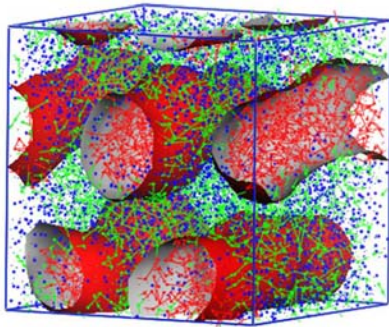
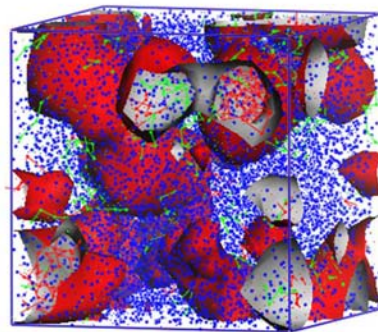
(a) $\phi=0.8$ (c) $\phi=0.4$ (b) $\phi=0.75$ (f) $\phi=0.3$ (e) $\phi=0.7$ (g) $\phi=0.2$ (d) $\phi=0.5$ (h) $\phi=0.1$ 

Figure 14. 對稱塊狀共聚物 $N=10(A_5B_5)$ 添加溶劑 S ，固定 $a_{AA}=a_{BB}=a_{BS}=25$ 、 $a_{AB}=51.3$ 以及 $a_{AS}=35.0$ 系統中，隨著共聚物體積分率(ϕ)變化的結構圖。其中，紅點、綠點以及藍點分別代表A成分、B成分以及溶劑 S ；紅色曲面為A成分之等位面(iso-surface)。

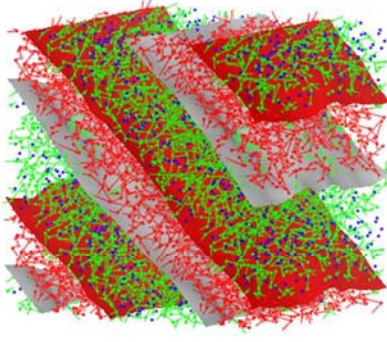
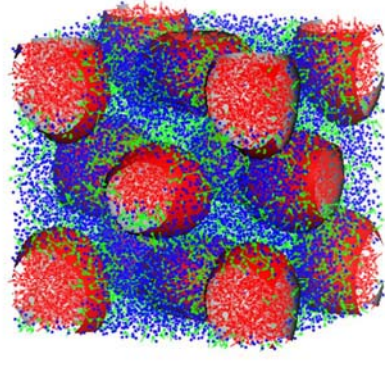
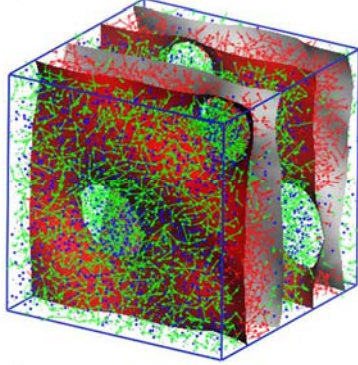
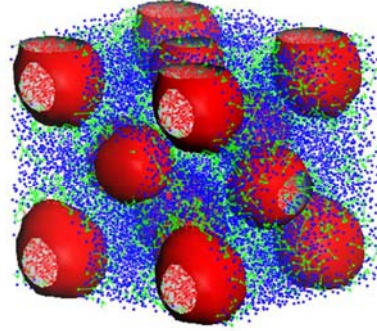
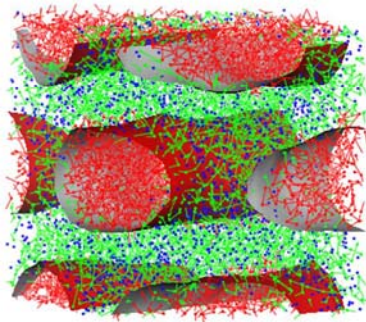
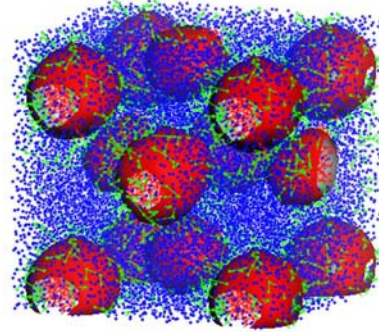
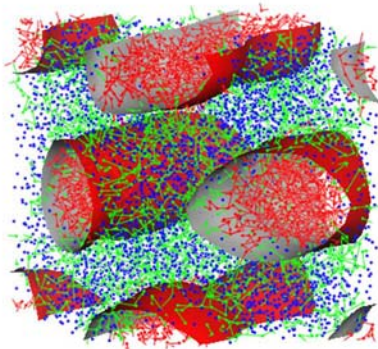
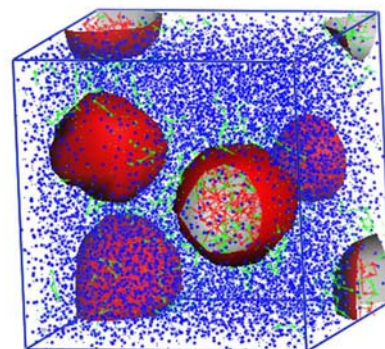
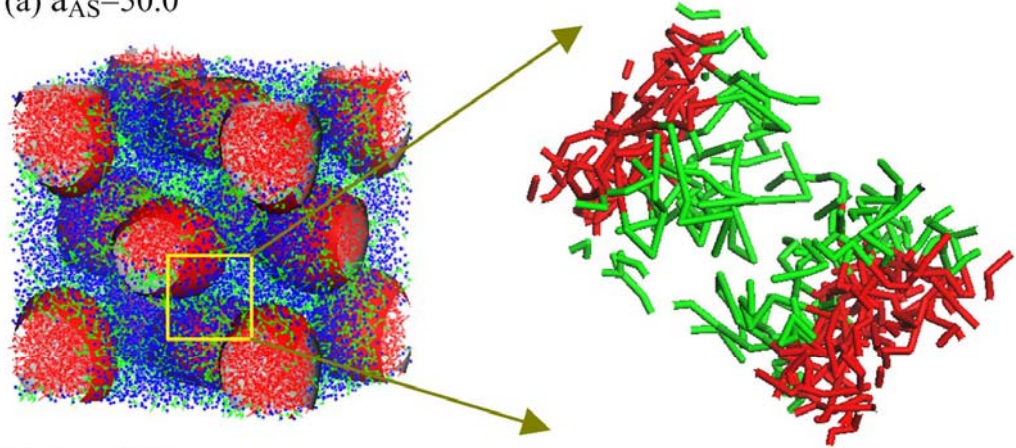
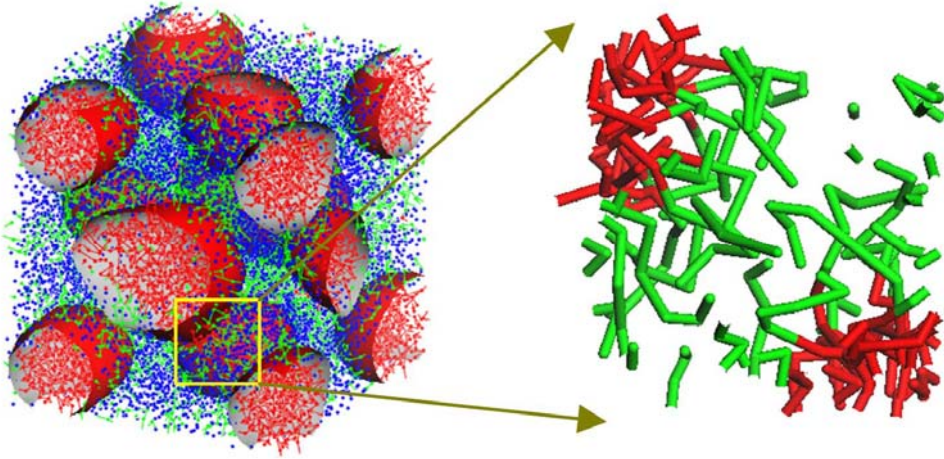
(a) $\phi=0.8$ (c) $\phi=0.4$ (b) $\phi=0.75$ (f) $\phi=0.3$ (e) $\phi=0.7$ (g) $\phi=0.2$ (d) $\phi=0.5$ (h) $\phi=0.1$ 

Figure 15. 對稱塊狀共聚合物 $N=10(A_5B_5)$ 添加溶劑S，固定 $a_{AA}=a_{BB}=a_{BS}=25$ 、 $a_{AB}=51.3$ 以及 $a_{AS}=50.0$ 系統中，隨著共聚合物體積分率變化的結構圖。

(a) $a_{AS}=50.0$



(b) $a_{AS}=45.0$



(c) $a_{AS}=40.0$

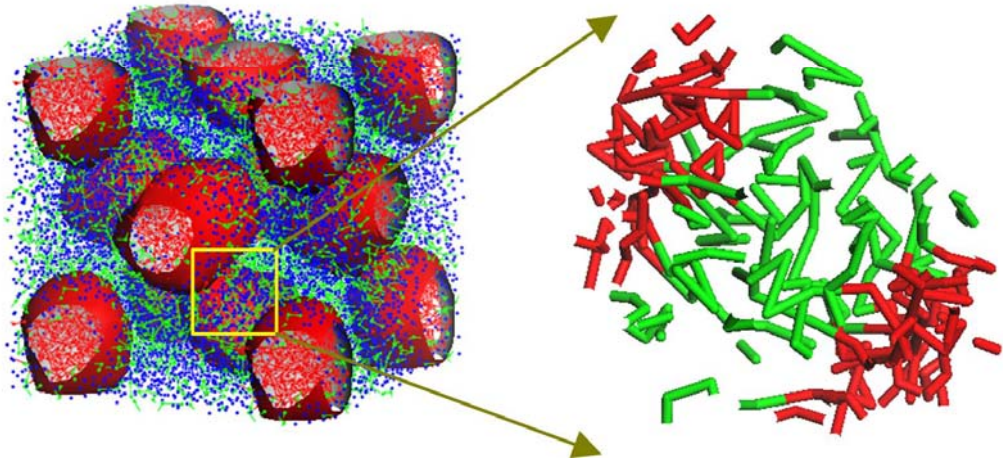


Figure 16. 對稱塊狀共聚合物 $N=10(A_5B_5)$ 添加溶劑S，固定 $a_{AA}=a_{BB}=a_{BS}=25$ 、 $a_{AB}=51.3$ 以及共聚合物體積分率 $\phi=0.4$ ：(a) $a_{AS}=50.0$ (b) $a_{AS}=45.0$ (c) $a_{AS}=40.0$ 的面心立方球狀堆積結構圖以及球狀微胞間B鏈段(綠線)的交錯情形。在右邊的圖形中，我們忽略了溶劑S。

行政院國家科學委員會補助國內專家學者出席國際學術會議報告

94 年 10 月 25 日

附件
一

報告人姓名	黃慶怡	服務機構 及職稱	國立台灣大學 高分子科學與工程學研究所
時間	2005/05/22~2005/05/27	本會核定 補助文號	NSC 93-2216-E-002-027
地點	Maastricht, The Netherlands (荷蘭)		
會議 名稱	(中文) 第三十四屆 CALPHAD 材料熱力學與相圖計算會議 (英文) CALPHAD XXXIV: International Conference on Phase Diagram Calculations of Materials		
發表 論文 題目	Phase Behavior of a Diblock Copolymer in the Presence of Two Solvents		

報告內容包括下列各項：

一、參加會議經過

05/21 抵達

05/22~05/27 參與會議

05/25 論文以口頭形式發表

05/27 離開荷蘭

二、與會心得

“CALPHAD”計算軟體應用在金屬以及氧化物的相圖計算已有三四十年的歷史，晚輩在美國西北大學修讀博士學位的時候，即曾經碰觸過此軟體應用在金屬方面的相變化研究。當時就有一個想法，想要把此軟體擴展至高分子的系統中。隨著回國研究的這七年中，深深覺得探討高分子相行為，為學子而言雖是一個基礎但也非常重要的一個課題，如果能夠藉由軟體的模擬演練方式，想必能夠提高學生對於基礎結構科學的興趣。

此行最大收穫即是認識了荷蘭 University Utrecht 的 Prof. H.A.J. Oonk 及其研究團隊，從他們的介紹以及交談中，瞭解到 CALPHAD Thermodynamics 在 University Utrecht 的應用與推廣情形。另外，Prof. H.A.J. Oonk 也談到目前 CALPHAD Thermodynamics 最遺憾的就是未能有效分析高分子的相行為，因此，如何建立高分子相行為的 database、以及修改 CALPHAD 軟體使其能描述高分子的熱力學與相圖計算，將是未來的研究重心！所以，後學能被邀請至會議中發表高分子的相變化，實肩負著此一很重要的使命。雖然，要達成此一目標還需走很長的路，但是經由與這些冶金熱力學的專家學者們的初步交談，再加上未來的持續合作，我們的目標應是指日可待的！

三、攜回資料名稱及內容

會議論文摘要紙本

部分參與研究人員名片或電子郵寄地址