Quantification of Thermal Material Degradation During the Processing of Biomedical Thermoplastics

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ABSTRACT: Three-dimensional plotting, a melt-based technique that involves the guided deposition of extruded filaments, is a promising method for the manufacture of thermoplastic scaffolds for tissue engineering. However, the polymer degrades during processing because of a prolonged residence time at an elevated temperature in the dispensing head before actual extrusion. This thermal degradation has been attributed to random chain scission; in this research, it has been quantified as a function of the residence time for a selection of polylactide-based polymers. The utilized techniques include thermogravimetric analysis (overall mass loss), differential scanning calorimetry (glass-transition, melting, and recrystallization temperatures), gel permeation chromatography (molecular mass), and inherent viscosity measurements. Experiments have shown that a static heating interval greater than 6 h is sufficient to destroy the molecular chain integrity of poly-(l-lactide) and poly(d,l-lactide). Copolymers with poly(ε-caprolactone) and poly(glycolic acid) are more resilient to thermal loading but are still severely affected. Overall, the differential scanning calorimetry results have been judged to be a good source of complementary data for the quantification of the degradation phenomenon, whereas inherent viscosity measurements have been confirmed to be related to the gel permeation chromatography results, which indicate shortening of the polymer chain. © 2011 Wiley Periodicals, Inc. J Appl Polym Sci 120: 2872–2880, 2011

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INTRODUCTION

Tissue engineering is an advanced biomedical research field with the aim of meeting the growing demand for donor organs and tissues.[1–3] According to the principles of tissue engineering, biodegradable scaffolds are used as support structures for culturing a patient's harvested cells in an *in vitro* environment to (re)create healthy tissues meant to be replacements for diseased ones. Although this neotissue grows, the scaffold slowly degrades into nontoxic components and eventually leaves only the new, functional, and healthy tissue behind. This final construct can be implanted into the patient and will not solicit any rejection because the cells are the patient's own. A popular production method for tissue engineering scaffolds in thermoplastic biodegradable polymers is layer additive manufacturing, which is often called three-dimensional (3D) plotting.[4–10] Such techniques operate batchwise; this means that a quantity of the material is heated to melting and is then gradually extruded to build a scaffold. The polymer often resides for longer times at temperatures above its melting point before it is actually processed. As described in a previous article,[11] the batch material will degrade because of this elongated heating time. The material flow changes, the scaffolds become more brittle, and the geometrical results differ from one scaffold to the next. In brief, an otherwise well-controlled production process such as 3D plotting loses reproducibility with respect to geometrical and mechanical properties because of material degradation during processing. The aim of the research at hand was to quantify this thermally induced material degradation to assess the gravity of the complications caused by batchwise 3D plotting of scaffolds with thermally sensitive polymers. This article focuses on medical-grade poly(lactic acid) and some of its copolymers. Although this research stems from polymer processing by 3D plotting, the results can be extrapolated to any production technique requiring a thermoplastic to be processed as a melt and especially to those techniques in which the molten material spends a certain time in waiting. Within this context, we define the residence time (t_r) as the time during which a material is kept

at an elevated processing temperature (T_p) before actual processing. Although the operation is continuous, microextrusion in itself is not a strongly degradative production process.[12] The polymer supply is kept in a cold hopper and fed into a heated barrel on a progressive basis. The material passes through a heated zone just before its immediate extrusion into a fiber; thermal loading of the polymer is thus limited. 3D plotting, however, combines microextrusion and programmed deposition of the extruded filament in such a fashion that a batch of the polymer supply is incorporated into the heated dispensing head of the machine.[11] This is true for both well-known commercially available systems: the Bioplotter from Envisiontec (Gladbeck, Germany) and the BioScaffolder from SysEng (Salzgitter, Germany). Hence, t_r will be prolonged to several hours as a function of the batch size. Smaller material batches have the advantage of lower t_r values, but they require closer monitoring by the operator, who must ensure that the material supply is not depleted halfway through a production run. With larger batch sizes, the machine may operate independently for longer times once the parameters are set, but the extension of t_r inevitably occurs. Moreover, and perhaps even more importantly, for a continuously heated batchwise system, t_r will increase for each subsequent scaffold produced.

Much research has been conducted on the degradation mechanisms of biomedical thermoplastic polymers; transesterification,[13–18] chain scission,[13,16,17,19–25] and unzipping[13,25] have been described. Most of these studies have investigated the depolymerization of these polymers at elevated decomposition temperatures.[14,19,22,25,26] Equally, *in vitro* and *in vivo* studies have described physical degradation by the hydrolysis of polymer-based implants under physiological circumstances,[27–30] typically around 37°C with a controlled pH. Of interest to the engineer processing materials from a granulate form into a scaffold, however, is the degradation behavior of the polymer at T_p. Pillin et al.[23] investigated the thermomechanical degradation of a polylactide (PLA) material induced by multiple injection cycles. They described a strong decrease in the weight-average molecular mass (M_w) and confirmed a 50% loss in M_w as soon as the completion of the third injection cycle. Signori et al.[16] processed PLA at different temperatures and noted an increase in the M_w loss with T_p rising. Analyzing the drop in M_w between virgin material and processed material, Carrasco et al.[20] recently reported decreases in the mechanical properties and molecular weight of PLA due to processing via injection and extrusion. All three of these studies showed that polymer degradation at T_p is caused by random chain scission. They described the degradation effect induced by the injection molding of PLA materials (preceded by extrusion

in the case of Carrasco et al.): the thermomechanical loading of the polymer differs from 3D plotting for scaffolds. With injection molding, the material enters the barrel as a solid granulate, and it is molten as it travels along the screw, which is subjected to an ascending heating profile. Moments after the highest temperature is reached in the nozzle area, the polymer is injected into the mold. The reservoir in which the material awaits entry into the barrel is at room temperature or a slightly elevated drying temperature far below the melting temperature (T_m). Both thermal and mechanical loadings are limited to the period of transport throughout the screw; no real t_r interval occurs. This is possible because of the larger scale of injection molding versus 3D plotting for scaffolds. Even for microinjection, the screw is still typically 10 to 15 mm in diameter and will take multiple granules into the channel depth of its lead, whereas the extrusion screw of the 3D plotting technique is only 6 mm in diameter and cannot take even a single granule (typically 2–4 mm in diameter) into its lead; thus, the material must be molten or at least sufficiently viscous already upon screw entry. As such, the batch material in the hopper reservoir is kept above T_m with a certain value of t_r.[11] This elongated heating period causes the polymer to degrade while it is waiting to be processed. The aim of this research was to quantify this thermal degradation of the thermoplastic during the extension of t_r before its actual processing through microextrusion and 3D deposition.

Thermogravimetric analysis (TGA) is often used to determine the weight loss of a material over a certain time or temperature interval.[14,16,17,22,27,31] This provides a good indication of the polymer's bulk degradation rate for a certain value of t_r at T_p, yet it offers little insight into changes in processing-related properties, which are not always coupled to material loss. These include the material viscosity, the melt flow, important temperature values such as the glass-transition temperature (T_g) and T_m, and the crystalline fraction upon cooling. Therefore, additional characterization tests have been introduced: inherent viscosity (η_{inh}) measurements, differential scanning calorimetry (DSC), and gel permeation chromatography (GPC).

As the material degrades, the polymer chain length, which is directly tied to the molecular mass, will decrease because of the scission of polymer molecule chains. As such, we decided to quantify the material degradation by the polymer's molecular mass evolution. The shorter chains are less entangled and enjoy greater mobility; this facilitates their transformation.[32] Both T_g[26–28,33] and η_{inh}[15,21,26,27,30,33] are considered to be representative material properties for the number-average molecular mass (M_n), with the effect on η_{inh} being significantly larger than that on T_g.[26,30] DSC analysis can yield a determination of

TABLE I
Selected Polymers

Material	Trade name	T_p (°C)
PDLLA (homopolymer)	Purasorb PL-18	225
PDLLA (50 : 50 D,L-copolymer)	Purasorb PDL-20	225
PLLA–PCL (70 : 30 copolymer)	Purasorb PLC-7015	140
PDLLA–PGA (50 : 50 copolymer)	Purasorb PDLG-5010	140

T_g and T_m. In addition to the information retrieved from T_g, the decrease in or even complete disappearance of T_m indicates the reduction (up to disappearance) of any remaining crystalline fraction in semicrystalline polymers.[32] Finally, GPC is a size exclusion technique used to determine the molecular mass.

EXPERIMENTAL

Materials

PLAs and their copolymers with poly(ε-caprolactone) (PCL) and poly(glycolic acid) (PGA) have a wide range of biomedical applications, such as scaffolds for tissue engineering or drug delivery[33–37] and resorbable parts (e.g., plates, screws, cages, and sutures).[26,28,35,38] Because of their environmentally friendly degradation capacity, PLAs are also strongly on the rise in the packaging industry, in which they are used in pure forms, filled forms, or blended forms with other polymers such as polyhydroxybutanoate.[12,39–41]

A representative selection of biodegradable thermoplastic materials, provided by Purac (Gorinchem, The Netherlands), was tested: poly(l-lactide) (PLLA), poly(d,l-lactide) (PDLLA), and two PLA-based copolymers (one with PCL and one with PGA; Table I). PLLA and PLLA–PCL are semicrystalline polymers, whereas PDLLA and PDLLA–PGA are amorphous in structure.

t_r

t_r in a heated batchwise processing system was mimicked by an isothermal heat treatment in a furnace at an equivalent value of T_p; intervals of 0, 1, 6, 12, and 24 h (i.e., t_r) were used. For the PLA materials, T_p was set at 225°C, whereas for the PGA and PCL copolymers, this was set at 140°C.

TGA

TGA was performed with a Netzsch (Selb, Germany) 449 F3 STA machine under a protective nitrogen atmosphere. Samples were heated to T_p (at a heating rate of 10°C/min) and then kept isothermally at T_p for 6 h, during which the mass loss was determined with Netzsch Proteus TA software.

DSC

DSC was performed with a TA Instruments (Zellik, Belgium) DSC 2910 under a nitrogen atmosphere. The sample material (15 mg) in an open aluminum pan was referenced against a empty open aluminum pan. The samples were heated to 30°C above T_m of the fresh material at a heating rate of 10°C/min. Measurements were started at 35°C for PLLA and PDLLA; for PDLLA–PGA and PLLA–PCL, they were started at −20°C. Real-time changes in enthalpy were determined with TA Instruments software, and T_m and T_g were thus ascertained.

η_{inh}

η_{inh} measurements were performed with a Schott Geräte (Hofheim, Germany) AVS 350 viscosity meter according to the ISO 1628 standard (25°C, a 0.25 g/dL solution in chloroform, and an Ubbelohde 0C column with a diameter of 0.47 mm) referenced against the pure solvent (chloroform UN 1888, Fiers, Kuurne, Belgium). Each sample was measured five times, and the results for the effluent time were averaged.

η_{inh} was calculated as follows:

$$\eta_{inh}(dL/g) = (\ln t_p/t_s)/c \tag{1}$$

where c is the concentration of the solution (g/dL) and t_p and t_s are the effluent times of the polymer solution and the pure solvent (s), respectively.[42]

GPC

GPC was performed at 30°C with a Waters (Zellik, Belgium) 410 differential refractometer, a Melz (Zellik, Belgium) LCD 212 refractive-index detector, and a reference cell filled with pure tetrahydrofuran. A mixed C.5-µm column (range = 200–2,000,000 Da) was used. Each sample (5–10 mg) was dissolved in 1 mL of tetrahydrofuran (3486 Chromasolv, Sigma–Aldrich, Bornem, Belgium). A Waters 510 high-performance liquid chromatography pump provided a constant flow of 1 mL/min. The detector was a Waters 410 differential refractometer. Detection signals were monitored in real time with Empower software from Waters. The setup was calibrated with polystyrene standards (1200, 7200, 21,000, 70,950, and 419,000 Da).

RESULTS

A general overview of the results is given in Table II. They are also depicted graphically per experiment in Figures 1–5.

TGA

The results of the TGA tests are listed in Table III and are illustrated in Figure 1. The heating of the material

TABLE II
Overview of the Characterization Results for Selected Polymers

Material	t_r (h)	t_p (s) Mean	t_p (s) Standard deviation	η_{inh} (dL/g)	T_g (°C)	T_m (°C)	M_n by GPC (Da)
Chloroform (t_s)	NA	125.84	0.250	1.000	—	NA	—
PLLA	0	172.52	0.273	1.262	70.36	196.19	139,948
PLLA	1	152.61	0.123	0.771	59.41	172.21	93,717
PLLA	6	134.61	0.339	0.269	52.85	151.97	3,600
PLLA	12	129.54	0.154	0.116	51.62	—	10,909
PLLA	24	126.88	0.110	0.033	50.65	—	5,329
PDLLA	0	193.55	0.092	1.722	55.80	NA	142,949
PDLLA	1	151.30	0.123	0.737	51.77	NA	20,341
PDLLA	6	129.62	0.217	0.118	49.00	NA	3,813
PDLLA	12	128.19	0.164	0.074	47.82	NA	6,104
PDLLA	24	127.33	0.266	0.047	48.17	NA	6,792
PDLLA–PGA	0	153.58	0.311	0.797	49.79	NA	178,271
PDLLA–PGA	1	150.38	0.545	0.712	47.42	NA	138,150
PDLLA–PGA	6	148.10	0.245	0.651	47.92	NA	116,990
PDLLA–PGA	12	147.14	0.211	0.625	47.92	NA	90,044
PDLLA–PGA	24	145.71	0.247	0.586	47.16	NA	90,156
PLLA–PCL	0	172.79	0.151	1.268	15.34	109.26	126,489
PLLA–PCL	1	172.11	0.227	1.252	21.47	—	123,297
PLLA–PCL	6	171.38	0.254	1.235	21.55	—	104,548
PLLA–PCL	12	167.54	0.332	1.145	21.23	—	86,823
PLLA–PCL	24	149.16	0.223	0.680	19.76	—	74,957

NA = Not applicable.

from room temperature to T_p is represented as the step from the virgin material to the 0-h point. Some mass loss has already occurred and can most likely be attributed to a drying effect of the raised temperature. For all materials, a discernible mass loss occurs when t_r is extended up to 6 h. This conforms to the expectation that semicrystalline polymers are more resilient against high temperatures than amorphous polymers. We ascribe this global mass loss under isothermal heating below depolymerization temperatures to continuous chain-end depolymerization and subsequent evaporation of the released oligomers. This mechanism has been identified as a low-tempe-

rature step in the process of nonisothermal degradation of PLA-based materials[13,19,43] and PCL.[24]

DSC

An overview of the results of the DSC analysis is listed in Table II and is further illustrated in Figure 2.

Changes in T_g are more pronounced for the pure PLA materials than for the copolymers. The evolution of T_g exhibits a clear decrease when t_r is extended, whereas PDLLA–PGA seems to stabilize after the first hour, and PLLA–PCL even demonstrates an initial increase in T_g. Among the selected
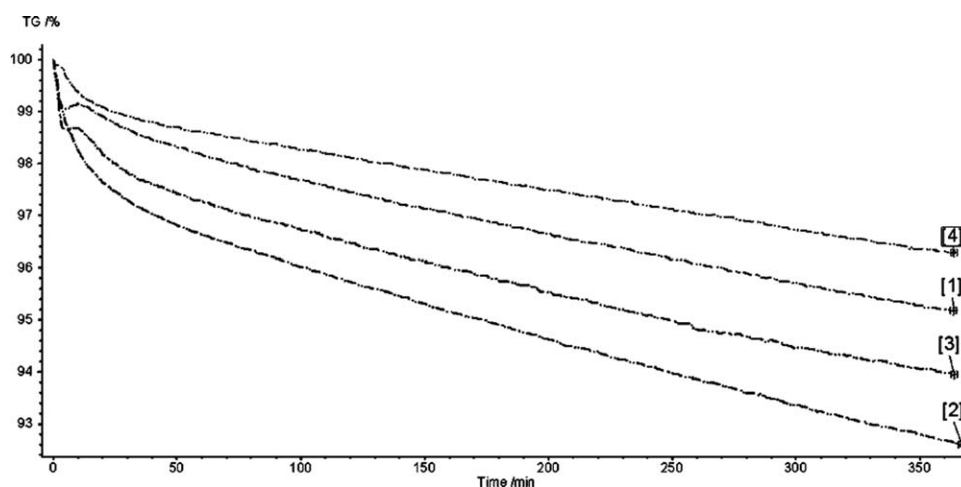

Figure 1 Mass loss as a result of extended heating at T_p: (1) PLLA–PCL, (2) PDLLA, (3) PDLLA–PGA, and (4) PLLA.
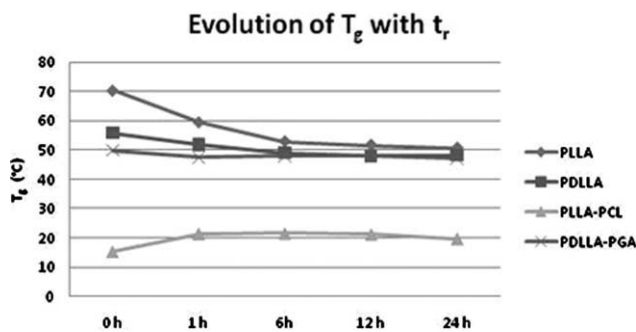
Figure 2 Decrease in T_g with the extension of t_r.

materials, the semicrystalline homopolymer PLLA is most affected. An overlay of the DSC curves for PLLA is shown in Figure 3. Over the first 6 h of the t_r extension, T_m falls from 196 to 152°C, and this implies a serious reduction in the crystalline fraction after cooling. After the 12-h interval, T_m is completely absent. The material is now completely amorphous. Furthermore, the large recrystallization peak that is evident for the 1-h-treated polymer is quickly reduced at the 6-h interval and is equally gone from 12 h onward. The matched T_m and recrystallization results confirm that the potential for crystallization has been eliminated, the shorter chains being unable to form such a highly organized, regular crystalline structure. For the virgin material, there is no recrys-

tallization peaking during the DSC heating cycle, and this indicates that the virgin material possesses it full potential crystalline fraction.

When we consider PLLA–PCL, we observe that the energy peak for T_m is evident only for the virgin material. Even though both homopolymers are semicrystalline in nature, only the PLA fraction crystallizes in the copolymer.[44] Once it is heated to melting, there is no more crystalline fraction upon cooling; all the material is now a single amorphous phase. This is corroborated by the rise in T_g after this first hour of thermal loading, which indicates a now much larger amorphous fraction to transform. DSC tests showed no peaks of recrystallization during heating, whereas a heating rate of 10 K/min does facilitate crystallization within PCL.[45] The crystallization potential of the PLLA–PCL material is effectively gone after $t_r = 1$ h. T_g remains stable around 21.5°C; after that, there is another 2°C decline at $t_r = 24$ h.

Both amorphous polymers show only an initial drop in T_g after the first heating interval, and this quickly levels off to a stable value of approximately 48°C for both PDLLA and PDLLA–PGA.

η_{inh}

The results of the η_{inh} measurements are listed in Table II and are illustrated in Figure 4. For all
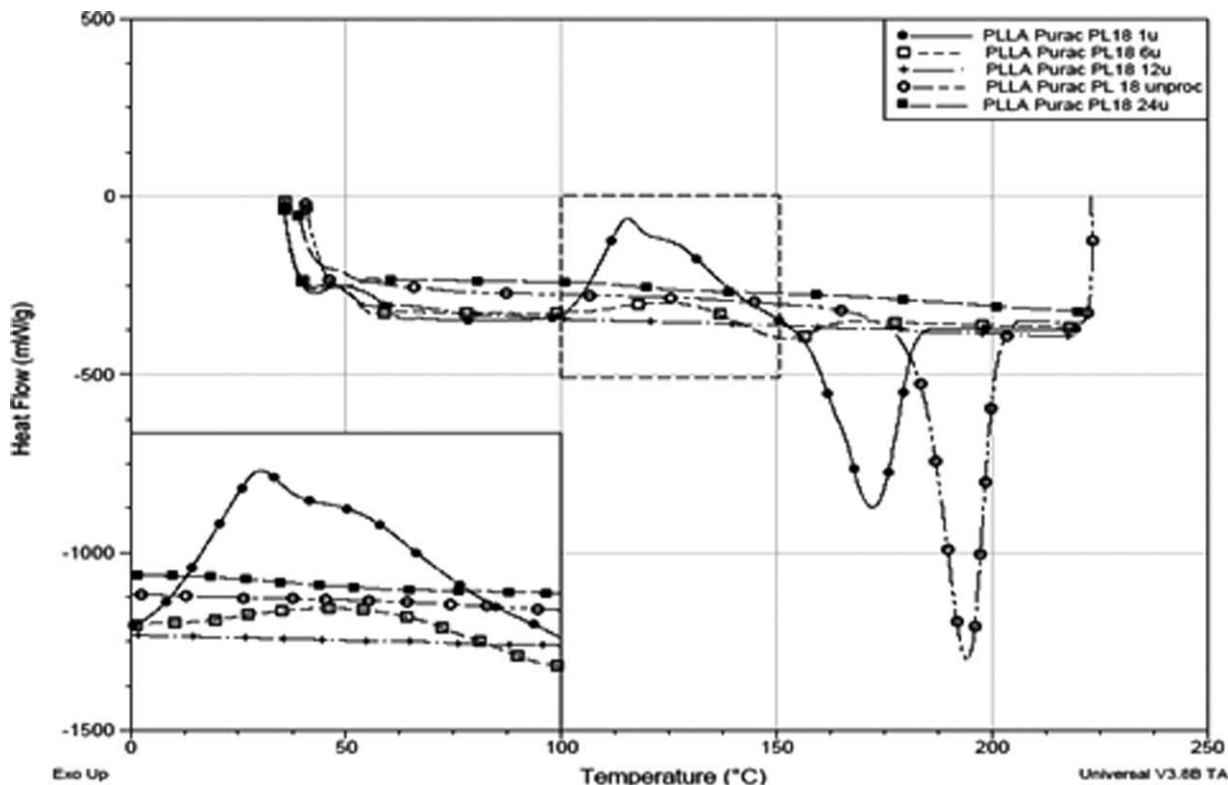

Figure 3 Overlay of the DSC results for the PLLA material. The inset is a magnification of the region of recrystallization during heating.
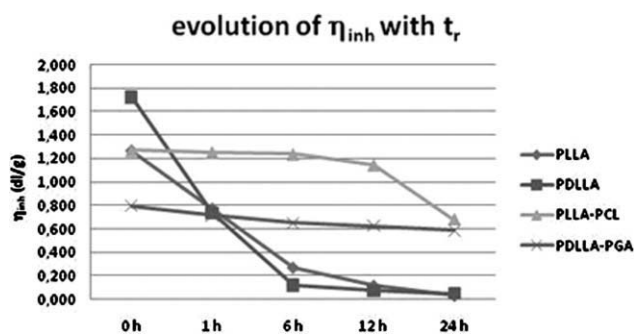
Figure 4 Decrease in η_{inh} with the extension of t_r.

materials, a downward trend in η_{inh} is evident with the prolongation of t_r. There is a very strong drop in η_{inh} for the pure PLA materials PLLA and PDLLA. Their solutions approach the same viscous flow rate as the solvent for the 12- and 24-h intervals, little effect of the remaining dissolved polymer is demonstrated. For both pure PLA polymers, the fall in η_{inh} is very steep over the first 6 h of t_r: it decreases from 1.26 to 0.27 dL/g and from 1.72 to 0.12 dL/g for PLLA and PDLLA, respectively.

Like T_g, η_{inh} of the PDLLA–PGA and PLLA–PCL copolymers appears to be less affected by the extended heating time, the interaction between the two monomers keeping the polymer stable and its solution viscous. Thermally more stable than PLA, the PCL fraction in PLLA–PCL contributes to a largely unmodified viscosity with t_r growing; only with the final 24-h interval does η_{inh} decrease significantly. η_{inh} of the PDLLA–PGA copolymer decreases moderately at a steady rate and drops by approximately 25% at $t_r = 24$ h.

GPC

The results of the GPC analysis are listed in Table II and are illustrated in Figure 5. The M_n trends largely follow those of η_{inh}. For the pure PLA materials, M_n drops spectacularly over the first 6 h of t_r. Semicrystalline PLLA maintains over half its chain length after 1 h of exposure to T_p but crashes in the 1–6-h interval. For both PLA materials, M_n values less
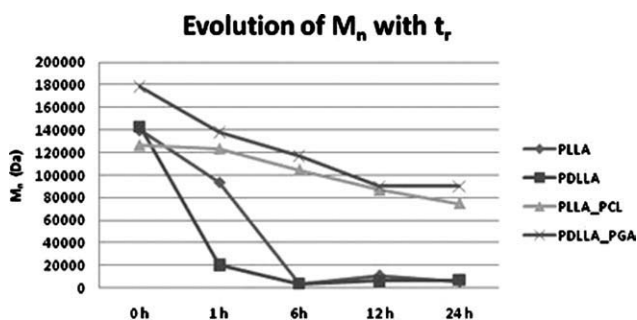
than 10,000 Da were measured for t_r values of 6 h or higher. Fluctuations at these low levels are attributed to the use of the large-range column.

Not having a crystalline backbone, amorphous PDLLA sees its M_n fall enormously from 143,000 to just over 20,000 Da after the first hour. The molecular mass of the copolymers with PGA and PCL is also significantly affected by the extended heating time, but in a less detrimental manner; both materials have their M_n values roughly halved with $t_r = 24$ h. The downward evolution of M_n is more pronounced than that of η_{inh}, and this suggests that although the polymer chains are noticeably shorter, the interaction between the two polymer fractions will maintain a viscous aspect until the 6-h interval. Thereafter, the remaining chains are extraordinarily short, and M_n drops below 10,000 Da.

Signori et al.[16] defined a degradation parameter (k) based on the M_n values of the unprocessed material and the processed material; on the basis of these results, it can be applied as follows:

$$k = \frac{M_{n(t_r=0)}}{M_{n_{(t_r)}}} \quad (2)$$

They reported a progressively increasing value of k with a rise in T_p for the PLA material up to $k = 1.7$. Carrasco et al.[20] reported k values of 1.44 and 1.62 for injected and extruded/injected PLA materials, respectively. The k parameters calculated for the PLLA and PDLLA polymers of our experiments are at least 10 times greater for the heating periods of 6 h and more. k rises to 1.98 and 1.69 for the PDLLA–PGA and PLLA–PCL copolymers, respectively, with $t_r = 24$ h. These results confirm a much stronger influence of the heating time versus an elevated T_p for the shorter thermomechanical loading period of conventional injection molding.

DISCUSSION

TGA experiments provide a relatively simple way of determining the occurrence of degradation. The shorter segments created by chain-end depolymerization will evaporate and result in some overall

TABLE III
TGA Results for Selected Polymers

Material	T_p (°C)	Virgin	Mass after t_r (%)		
			0 h	1 h	6 h
PLLA	225	100	99.84	98.61	96.31
PDLLA	225	100	98.58	96.63	92.69
PLLA–PCL	140	100	99.05	98.18	95.2
PDLLA–PGA	140	100	98.67	97.27	93.97

Figure 5 Decrease in M_n with the extension of t_r.

mass loss. The lead mechanism of thermal degradation at T_p is believed to be random chain scission, however,[16,20,21,23] and TGA results offer no insight into the measurement of material degradation at the level of the polymer chain length and the related molecular mass. To quantify the thermal degradation phenomenon, additional experiments more closely related to the polymer chain length (DSC, η_{inh} measurements, and GPC) have provided more relevant data. Overall, the results from these experiments indicate significant shortening of polymer chains during the material's isothermal heating at T_p, and this supports the occurrence of material degradation by random chain scission.

Information obtained by DSC reflects the appearance of shorter chains in the polymer as a decreasing value of T_g, which, however, appears to stabilize at a lower level after the 6-h interval. Interestingly, the semicrystalline (co)polymers PLLA and PLLA–PCL do not retain their crystalline fraction over the t_r extension. This destruction of the crystalline potential may actually be considered a positive influence on *in vitro/in vivo* degradation by hydrolysis of the thermoplastic scaffolds because the crystalline lattice is much harder to disrupt. Likewise and much less desirably, it will strongly affect the reproducibility of mechanical scaffold properties.

The results for η_{inh} and M_n appear to be related for the different materials; curves for both pure PLA materials show a sharp drop to very low values over the first 6 h of t_r, whereas the copolymers undergo a more steady decline with the prolongation of t_r, with the curve of M_n being somewhat steeper than that of η_{inh}. When they are compared on a logarithmic scale (ln η_{inh} versus ln M_n), a statistical analysis yields a correlation of $r = 0.87$ (n = 20, $P < 0.01$). This relationship between η_{inh} and the molecular weight is described by the Mark–Houwink–Sakurada equation:[46]

$$\eta_{inh} = KM_v{}^a \qquad (3)$$

where M_v is the viscosity-average molecular mass and K and a are constants for a particular polymer–solvent pair at a particular temperature. Because it is difficult to determine directly, M_v is often replaced by M_w for the purpose of determining K and a. For a solution in chloroform, $K = 5.45$ dL/g and $a = 0.73$ and $K = 2.21$ dL/g and $a = 0.77$ have been reported for PLLA and PDLLA, respectively.[39] These have been employed in the literature via eq. (3) to determine M_v from η_{inh}, and they have validated η_{inh} experiments as an indirect means of measuring the molecular mass.[47,48]

For the PLLA and PDLLA materials, it is remarkable that the first 6 h of t_r is sufficient to reduce M_n below 10,000 Da, with k values higher than 20 noted. This is reflected by matching η_{inh} values that drop below 0.05 dL/g, with the dissolved polymer chains barely adding to the viscosity of the solvent. Even in the DSC results this trend is manifested, with T_g dropping more sharply during the 6 h of t_r before stabilizing at a lower value afterwards. Copolymerization has a strong stabilizing influence on the PLA material when it is exposed to T_p. With k values below 2, the impact of prolonged heating at T_p is still noteworthy but no longer as detrimental as it is for the pure PLA materials. The degradation effect is closer to that of thermomechanical loading during a single injection-molding cycle.[20] T_g of both copolymers remains stable with the t_r extension. Such improved resistance to thermal degradation has equally been confirmed for copolymers and for blends of PLA with other degradable polymers.[16,49] It would be interesting for the sake of reference to evaluate similar degradation behaviors of pure PGA and PCL materials with which PLA is copolymerized. For PLLA–PCL with $t_r = 24$ h, for example, for which η_{inh} and M_n both suddenly drop, it can be speculated that thermal degradation has set in for the PCL component of the polymer as well; previously, the overall polymer chain length was largely maintained.

Notwithstanding a possible favorable effect on the *in vivo* degradation time, thermal degradation of the scaffold material during the production process is devastating to scaffold reproducibility with respect to the geometry and mechanical properties. This conflicts with an important strategic goal in the field of tissue engineering: the translation of research-level technology to clinical applications on a larger scale.[50] For this industrial level to be accomplished, a reliable scaffold production technique with high reproducibility is required.[4] The same set of processing parameters and materials should yield the same scaffold geometry and properties every time. As such, the thermal loading of the polymer during processing either must be limited as much as possible or, when it cannot be avoided, should be controlled in such a fashion that the virgin polymer is subjected to an equal t_r for every scaffold produced.

CONCLUSIONS

The thermal degradation due to a static t_r value at T_p has been quantified for PLLA, PDLLA, PDLLA–PGA, and PLLA–PCL materials. Both the semicrystalline PLLA and the amorphous PDLLA lose nearly all molecular chain integrity after 6 h of exposure to T_p, and although the copolymers are significantly more resistant, they also undergo thermal degradation by chain scission. Hence, for the melt-based processing of these polymers, we conclude that a prolonged heating time before the actual processing will strongly affect the physical properties and result in a severe loss of final part reproducibility.

We strongly recommend that the melt-processing techniques of thermally sensitive PLA-based polymers be revised to eliminate as much as possible any t_r, and we suggest the heating of the virgin material just before the actual processing instead of using a heated batch.

Analyzing the polymer degradation mechanism itself was not considered within the scope of this study because this has been reported at length in existing articles. We subscribe to the well-documented scientific opinion that degradation at T_p is caused mainly by random chain scission.[16,20,21,23] We have endeavored to quantify this degradation as a function of t_r to document the practical problem of changes in the material flow and properties during 3D plotting. This study does not take into account the possible influence of oxygen from the surroundings on the polymer degradation process during isothermal heating. It has been reported that PLA material degrades faster in air than under a nitrogen atmosphere because of a thermooxidation step.[16,21]

NOMENCLATURE

3D	three-dimensional
η_{inh}	inherent viscosity (dL/g)
a	constant
c	concentration of a polymer solution (g/dL)
DSC	differential scanning calorimetry
GPC	gel permeation chromatography
k	degradation parameter
K	constant
M_n	number-average molecular mass (Da)
M_v	viscosity-average molecular mass (Da)
M_w	weight-average molecular mass (Da)
PCL	poly(ε-caprolactone)
PDLLA	poly(d,l-lactide)
PGA	poly(glycolic acid)
PLA	polylactide
PLLA	poly(l-lactide)
T_g	glass-transition temperature (°C)
TGA	thermogravimetric analysis
T_m	melting temperature (°C)
t_p	effluent time for a polymer solution through an Ubbelohde capillary (s)
T_p	processing temperature (°C)
t_r	residence time (s)
t_s	effluent time for a pure solvent through an Ubbelohde capillary (s)

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