Degradable polyurethane networks based on D,L-lactide, glycolide, ε-caprolactone, and trimethylene carbonate homopolyester and copolyester triols

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Several trifunctional, hydroxy-telechelic polyester and poly(ester–carbonate) homopolymers and copolymers were synthesized by the triol-initiated, ring-opening bulk polymerization of D,L-lactide, glycolide, ε-caprolactone, and/or trimethylene carbonate. The molar compositions of the copolymers were determined from ^{13}C n.m.r. spectra; hydroxy equivalent weights were determined by acetylation titration, and the glass transition temperatures (T_g) and melting points were determined by differential scanning calorimetry (d.s.c.). Crosslinked polyurethane networks were prepared by reacting the hydroxy-telechelic prepolymers with tolylene diisocyanate. Network characterization included determination of sol content by solvent extraction, T_g by d.s.c., tensile properties by stress–strain measurements, and degradation properties by hydrolytic degradation studies. Equilibrium swelling results indicated that the per cent weight gains correlated well with the prepolymer hydroxy equivalent weights, with the exception of the poly(glycolide-*co*-trimethylene carbonate) network. The poly(D,L-lactide) and poly(D,L-lactide-*co*-trimethylene carbonate) (PLTMC) networks had the highest tensile strengths of 49.60 and 41.27 MPa, respectively, and glass transition temperatures of 51.3 and 21.3°C, respectively. All other networks were highly flexible with tensile strengths of 12 MPa or less. Hydrolytic degradation studies were conducted by placing network samples in a solution of phosphate-buffered saline (pH = 7.4) at 37°C. Weight uptake of buffer and weight loss of the networks were monitored gravimetrically over time. Tensile properties, monitored as a function of degradation time, indicated that the poly(ε-caprolactone-*co*-D,L-lactide) and PLTMC networks displayed a linear loss of strength with respect to weight during the first 30 days of degradation; the other networks degraded either too slowly or too quickly to establish such a linear relationship.

(Keywords: polyurethane networks; homopolyesters; copolyesters)

INTRODUCTION

In recent years, bioabsorbable polymers have been developed for medical applications such as surgical sutures[1], temporary scaffolds for hard and soft tissue[2], and implantable drug delivery devices[3]. Most of these devices are made from high-molecular-weight linear aliphatic polyesters, such as poly(glycolide) (PG), poly(D,L-lactide) (PL), polydioxanon (PDS), poly(ε-caprolactone) (PC), poly(trimethylene carbonate) (PTMC), and their copolymers and terpolymers. These thermoplastics are well suited for use as fibres owing to their semi-crystalline morphology. Aliphatic polyesters do, however, present obvious strength limitations, and it would appear that only composite structures, composed of directionally reinforcing fibres and a suitable matrix, are capable of providing the initial strength and rigidity required of general non-suture applications such as tissue scaffolds and fixation devices. We have reported the synthesis of bioabsorbable epoxy resins for use as composite matrices[4], as well as the fabrication of composites using the absorbable surgical mesh Dexon® (PG) as the reinforcing fibre, and matrices formed by crosslinking liquid polyester prepolymers either through free radical polymerization[5] or through the formation of urethane linkages[6].

Crosslinkable liquid prepolymers offer advantages for composite fabrication using continuous fibre reinforcement, and they provide amorphous polyester networks which represent an alternative to the common linear, semi-crystalline polymers. Owing to their morphology, the linear semi-crystalline materials often display heterogeneous degradation patterns, i.e. the amorphous regions degrade faster than the crystalline regions. This can be a disadvantage in that their degradation profiles typically show a non-linear loss of strength with respect to loss of mass over time, and quite often total mechanical failure occurs while a large fraction of the mass of the polymer remains. It has been suggested that amorphous, crosslinked polymer systems may offer a more linear decrease in strength with respect to mass during degradation[7–9].

In the course of our investigations of amorphous, crosslinked polyester systems, we have concentrated our efforts in two areas: one is the crosslinking chemistry itself, e.g. free radical polymerization, reaction of epoxide

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groups, etc.; the other is the composition of the primary polyester chains. Using random copolymerization, there are many possible combinations of bioabsorbable monomers that can lead to liquid, reactive prepolymers, and through proper design they can be tailored to yield amorphous networks which meet specific material requirements, such as glass transition temperature (T_g) and degradation rate.

The monomers used in this study were chosen from the well-accepted family of lactones and cyclic carbonates used to produce the linear, semi-crystalline, bioabsorbable polymers discussed above. By varying the combinations of these monomers, a variety of amorphous, hydroxy-telechelic prepolymers were produced, which were either polyesters or poly(ester–carbonate)s. Networks were prepared directly from each prepolymer using tolylene-2,6-diisocyanate (TDI) as the crosslinking agent. Of course, TDI yields non-bioabsorbable crosslinks; however, its use allowed quick preparation of the networks for screening purposes, therefore avoiding more elaborate crosslinking systems involving modification of the prepolymer end functionality, e.g. to epoxide, or use of a biocompatible polyisocyanate such as L-lysine diisocyanate. This paper discusses the preparation of these amorphous, hydroxy-telechelic prepolymers and urethane networks, and their thermal, tensile, and degradation properties.

EXPERIMENTAL

Materials

All reagents were used from the supplier without further purification, unless specified otherwise. Acetic anhydride (99+%), chloroform-d (99.8 at%), 1,2-dichloroethane (DCE, 99%, distilled from calcium hydride), diethyl carbonate (99%), trimethylolpropane (99+%), magnesium sulfate (99%), 1-methylimidazole (99+%), 1,3-propanediol (98%), δ-valerolactone (99%), tolylene-2,6-diisocyanate (97%, distilled from calcium hydride prior to use), and 1,1,2,2-tetrachloroethane (99%) were purchased from the Aldrich Chemical Co. Chloforom, methylene chloride, benzene, hexanes, methanol, and tetrahydrofuran (THF, distilled from calcium hydride) were all ACS reagent grade and purchased from Fisher Scientific. D,L-lactide (99+%) and glycolide (99+%) were purchased from Henley Chemicals Inc. Tin(II) octoate and phosphate-buffered saline (dry powder blend, pH = 7.4) were purchased from the Sigma Chemical Co. ε-Caprolactone, distilled from calcium hydride prior to use, was donated by the Union Carbide Co.

Instrumentation

Gel permeation chromatography (g.p.c.) was performed on all prepolymer samples to determine molecular weights (M_n) and molecular weight distributions (M_w/M_n) with respect to polystyrene standards (Polysciences Corporation). The Waters Associates system used was equipped with a Rheodyne injector, a Waters 510 high performance liquid chromatography (h.p.l.c.) solvent pump, two linear Ultrastyragel® columns connected in series, and a Waters 410 differential refractometer. The columns and refractometer were maintained at 37°C. THF, freshly distilled from calcium hydride, was used as the eluent, and was delivered at a flow rate of 1.0 ml min^{-1}. Sample concentrations were approximately 0.5% (w/v) in THF with an injection volume of 50 μl.

^{1}H n.m.r. (200 MHz) spectra were recorded on a Bruker AC-200 spectrometer using 5 mm tubes. Sample concentrations were 1–2% (w/v) in deuterated solvents using tetramethylsilane (TMS) as an internal reference. ^{13}C n.m.r. spectra were recorded on a Bruker AC-200 or Bruker AC-300 spectrometer using 5 mm tubes. The molar compositions of the monomers in the copolymers were determined from their respective 50.7 MHz ^{13}C n.m.r. spectra. Sample concentrations were 25% (w/v) in $CDCl_3$ with TMS as the internal reference. In order to integrate the peaks in the ^{13}C n.m.r. spectra, a gated decoupling program was used with the following acquisition parameters: 30 μs pre-scan delay, two dummy scans, 5 μs pulse width, 10H decoupling power, and 200–3000 transients.

Tensile tests were conducted on an MTS 810 universal test machine at room temperature using microdogbone test samples. The microdogbone was a scaled-down version of the ASTM standard dogbone. The samples were tested using either a 1000 lb (1 lb = 0.45 kg) load cell at 10% range or a 5 lb load cell at full range, and the maximum stroke length was 10 in (1 in = 2.54 cm). Modulus, tensile strength and strain were calculated with the MTS TestWorks software.

Trimethylene carbonate

1,3-Propanediol (132.8 g, 1.745 mol), diethyl carbonate (240.3 g, 2.034 mol, 16% molar excess), and sodium methoxide (0.0087 mol, 0.5 mol%) were combined in a 500 ml distilling flask equipped with a thermometer, distillation apparatus and magnetic stirrer. The flask was placed in a 130°C oil bath, and the temperature (internal flask temperature) was raised to 170°C over a period of 3 h as ethanol was distilled off. A vacuum was applied for 15 min at the end of the reaction. The cooled residue was dissolved in benzene (500 ml), washed with water (2 × 200 ml), dried over $MgSO_4$ and filtered, and the solvent was removed under reduced pressure. The washed product was vacuum distilled. The first fraction was unreacted 1,3-propanediol (70–90°C, 0.25 mmHg) and the second fraction was trimethylene carbonate. The monomer, recrystallized from anhydrous ether prior to use, was obtained in a yield of 48% (85.8 g, 0.84 mol): boiling point 110–115°C at 0.25–0.80 mmHg; ^{1}H n.m.r. (200 MHz, $CDCl_3$) δ 4.46 (t, 4H), 2.16 (quintet, 2H); ^{13}C n.m.r. (50.7 MHz, $CDCl_3$)δ 148.29, 67.75, 21.24.

Hydroxy-telechelic oligomers

All polymerizations were conducted in the bulk under nitrogen at 140°C. Glassware was dried at 145–155°C for 24 h, fitted with rubber septa and cooled under a flow of dry nitrogen. Polymerizations (50–100 g) were run in 250 ml Erlenmeyer flasks with 24/40 ground glass joints sealed with evacuated glass stoppers wrapped with Teflon tape. Trimethylolpropane served as the trifunctional initiator and tin(II) octoate was used as catalyst at a concentration of 1.4×10^{-4} mol per mole of monomer. A typical experimental procedure was as follows. To a flask (250 ml) containing a magnetic stirrer bar were added D,L-lactide (72.06 g,0.500 mol), ε-caprolactone (57.07 g, 0.500 mol), trimethylolpropane (8.66 g, 0.0646 mol), and tin(II) octoate. The flask was purged with nitrogen and heated in a 140°C constant temperature bath for 16 h with continuous stirring. The polymerization was quenched in an ice–water bath, and the polymer was dissolved in methylene chloride, followed by precipitation in a 10-fold excess of stirred hexanes. The hexanes layer was decanted, and the polymer was rinsed with fresh aliquots of hexanes (3 × 100 ml). The isolated polymer was then redissolved in methylene chloride, the solution was transferred to a specimen jar, and the pure polymer was freed of solvents by heating for 24 h in an 80°C oven, and then for 24–48 h at 80°C *in vacuo*. Feed compositions and [M]/[OH] ratios for all polymerizations are given in *Table 1*.

Acetylation titration of hydroxy-telechelic polyesters

This procedure has been described elsewhere[10].

Urethane networks

An illustrative procedure was as follows. To an 8 oz specimen bottle containing a stirrer bar were added PCL (25.00 g, 0.036 eq hydroxy groups) and 1,1,2,2-tetrachloroethane (TCE) (100 g). The mixture was stirred until it became homogeneous. TDI (3.14 g, 0.036 eq isocyanate groups) was then added, and the mixture was stirred until it became homogeneous, after which tin(II) octoate (three drops, 0.1 wt% based on total solids content) was added with further stirring. Approximately 12 g (~2.4 g of polymer per pan) of the solution were added to smooth-sided, aluminium weighing pans that had been pretreated by coating with a fluorocarbon mould release agent and heating at 80°C for 2 h. The pans containing the solution were placed in an 80°C oven for 72 h and were then heated at 80°C under vacuum (30 inHg; 1 in = 25.4 mm) for 24 h. The networks were allowed to cool to room temperature slowly under vacuum (30 inHg).

Extraction and equilibrium swelling

Extraction and swelling studies were carried out using anhydrous 1,2-dichloroethane. The extraction procedure was as follows. A disc ($d = 0.5$ in) was stamped from the polymer network, weighed to the nearest 0.1 mg, and placed in a 2 oz specimen jar; 1,2-dichloroethane (30 ml) was added to the jar. After 24 h the solvent was discarded, and a fresh aliquot was added; this was repeated for a minimum of three changes. The swollen network was then dried to constant weight *in vacuo*. Sol contents reported are the average of three individual extraction samples, with the exception of the degradation samples. The following equilibrium swelling procedure was typical. The physical dimensions of the previously extracted disc were measured; the disc was weighed, and its density was calculated. The sample was placed in a 2 oz specimen jar containing 1,2-dichloroethane (30 ml). The sample was removed, patted dry, and weighed at frequent intervals initially, until the uptake of solvent slowed, and then in increments of several hours for a period of several days.

Degradation analyses

Network degradation was studied by placing tared discs ($d = 0.5$ in) into 11 dram (1 dram = 39 ml) vials containing 35 ml of phosphate-buffered saline (pH = 7.4) in an incubator at 37°C. Both weight gain due to water uptake and weight loss due to degradation were measured as follows. Periodically, two samples were removed, patted dry, and weighed to determine the average weight gain due to hydration. The samples were then placed in tared scintillation vials (20 ml) and dried under high vacuum at room temperature for 7 days or until a constant weight was reached. These samples were then dried under vacuum (30 inHg) at room temperature until they reached a constant weight, from which the weight loss was determined. These samples were then used for extraction studies. Some time later two more samples were removed and the procedure was repeated, and this process was continued until the discs became too tacky to weigh accurately while wet, after which only the weight loss was determined.

Loss of tensile strength over time was determined by placing a number of tared dogbones, in individual 2 oz specimen jars containing buffer solution (30 ml), in the incubator and removing them immediately prior to tensile testing. The samples were analysed to determine weight gain and loss and were tensile tested while saturated with buffer.

RESULTS AND DISCUSSION

The properties of amorphous, bioabsorbable networks may be quickly explored through the creation of polyester-based polyurethanes via the reaction of hydroxy-terminated polyester oligomers and polyfunctional isocyanates. Such oligomers are conveniently prepared by the ring-opening homopolymerization and/or copolymerization of lactones and carbonates, initiated by an alcohol and catalysed by tin(II) octoate. The propagation reaction consists of the stepwise addition of monomer to the growing chain ends, which have been suggested to be terminal hydroxy groups[11,12] or stannyl ether bonds[13]. The use of difunctional alcohol initiators results in linear polymers, whereas multifunctional alcohol initiators form star-type polymers, all of which are hydroxy-telechelic. The number average molecular weight (M_n) at any time has been shown to be proportional to conversion and to the initial monomer/alcohol ratio[10,12].

In this work, three-arm star homopolymer and copolymer oligomers were prepared using trimethylolpropane as initiator at an [M]/[OH] ratio designed to produce molecular weights of 2000 g mol^{-1}. Since the monomers had different formula weights, this ratio varied for each polymer system. *Figure 1* shows the repeat units for the various monomers used, and *Table 1* lists the six copolymers and two homopolymers that were prepared, and the acronyms which are used to refer to them throughout this paper. The feed composition for each copolymer was 50/50 (mol%) with the exception of PCV, for which the feed composition was chosen to be 60/40 since that has been shown to yield the lowest melting

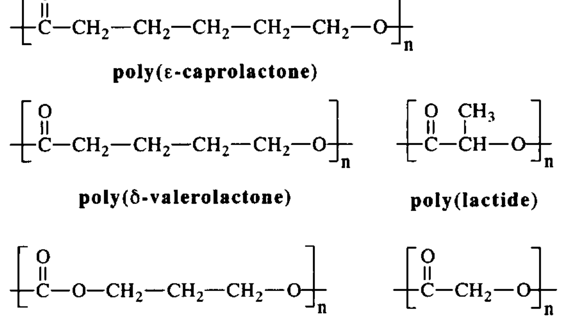

Figure 1 Repeat units for the individual monomers used to prepare the polyester and poly(ester–carbonate) prepolymers

Table 1 Monomer feed compositions and [M]/[OH] ratios for triol polymerizations

Sample	M_1	Weight (g)	Mole fraction	M_2	Weight (g)	Mole fraction	[M]/[OH]
PCL	C[a]	57.07	0.50	L	72.06	0.50	5.16
PCV	C	68.48	0.60	V	40.05	0.40	6.14
PLTMC	L[b]	25.01	0.50	TMC[e]	17.71	0.50	5.42
PCTMC	C	25.22	0.50	TMC	22.56	0.50	6.12
PVTMC	V[c]	30.58	0.50	TMC	31.18	0.50	6.59
PGTMC	G[d]	25.06	0.50	TMC	22.04	0.50	6.11
PC	C	122.45	1.00	–	–	–	5.84
PL	L	75.49	1.00	–	–	–	4.63

[a] ε-Caprolactone
[b] D,L-Lactide
[c] δ-Valerolactone
[d] Glycolide
[e] Trimethylene carbonate

temperature of all possible compositions for that particular copolymer[10].

Table 2 lists the experimental molecular weights (acetylation titration and g.p.c.) and comonomer compositions (^{13}C n.m.r.) of the various oligomers. The molecular weights obtained by titration ($M_{n(OH)}$) were significantly lower than the theoretical value of 2000 g mol^{-1}, and this was most likely due to incomplete monomer conversion since all polymerizations were conducted for 16 h regardless of composition. Molecular weights determined by g.p.c. using a polystyrene calibration varied widely, as would be expected for polymers of such diverse compositions, and thus the g.p.c. results were used only as a qualitative tool to check for residual monomer, peak shape, size distribution, etc. Typically in the literature, copolymer compositions of copolyesters and poly(ester–carbonates) have been determined by ^{1}H n.m.r. However, such studies have involved comonomers in which the signals for the polymer methylene units did not overlap, and the copolymers were of high molecular weight, thereby minimizing the effects of the initiator fragments and ultimate monomer units. In the case of the alcohol-initiated lactone and carbonate polymerizations, the signal for the initiator fragment, in many cases, was overlapped by the signal for the methylene linkages of one of the monomer repeat units, resulting in inaccurate integrations for these signals, especially since there was a high concentration of trifunctional initiator in these oligoesters. Therefore, copolymer compositions were determined by ^{13}C decgate n.m.r., which is a gated decoupling program allowing the complete relaxation of the carbon nuclei in order that their signals may be integrated. Different sets of signals were used to determine the compositions of the copolymers, depending on their chemical shifts and resolutions. The copolymer compositions determined were within 0.3–7.7% of the initial feed ratios (*Table 1*).

The oligomers all had T_gs below room temperature (*Table 3*), producing liquid resins in all cases except for the PC and PL homopolymers. The latter, with a T_g of 14°C, was very viscous, and nearly glass-like. PC, PCV, and PGTMC were all semi-crystalline materials. However, at room temperature PCV was above its T_m of 10.1°C and was non-crystalline, and PGTMC remained a liquid in spite of two very weak melting endotherms at 42.1 and 126.0°C, attributed to short runs of glycolide units in the copolymer. The PC homopolymer was highly crystalline as expected, but it was also anticipated to produce a network that was amorphous since it was of such low molecular weight.

Table 2 Compositions and molecular weight results for trifunctional hydroxy-telechelic polymers

Sample	Composition[a] (mol%)	$M_{n(OH)}$	$M_{n(g.p.c.)}$	M_w/M_n[b]
PCL	55.3 C 44.7 L	2110	1490	2.28
PVC	62.6 C 37.4 V	1410	2620	1.74
PLTMC	45.3 L 54.7 TMC	1240	1360	2.58
PCTMC	54.5 C 45.5 TMC	1300	1060	3.71
PVTMC	50.3 V 49.7 TMC	1220	1290	3.37
PGTMC	42.3 G 57.7 TMC	1590	1224	2.51
PC	100.0	1351	2322	1.75
PL	100.0	1540	1762	1.40

[a] Determined from ^{13}C decgate n.m.r.
[b] Determined from g.p.c. chromatograms

Table 3 Effect of copolymer composition on the thermal transitions of triol polymers and networks

Sample	Copolymer T_g[a] (°C)	Copolymer T_m[b] (°C) (ΔH)	Network T_g[a] (°C)
PCL	−26.2	–	5.9
PCV	−64.8	10.1 (66.6 J g^{-1})	−44.7
PLTMC	−5.2	–	21.3
PCTMC	−49.6	–	−20.6
PVTMC	−46.4	–	−17.1
PGTMC	−12.0	42.1 (3.0 J g^{-1}) 126.0 (0.8 J g^{-1})	4.1
PC	−59.0	21.5 (8.5 J g^{-1}) 47.3 (62.8 J g^{-1})	−40.9
PL	14.0	–	51.3

[a] Determined from d.s.c. thermograms at 20°C min^{-1}
[b] Determined from d.s.c. thermograms at 10°C min^{-1}

Although biocompatible curing chemistries were our eventual goal, in this study the triols were cured with

TDI to provide a quick screening of properties and to avoid more elaborate crosslinking systems such as derivatization of the prepolymer end functionality, e.g. to epoxide[4,14], or the use of L-lysine diisocyanate[15], which is not presently commercially available. Networks were formed by heating films of oligomer and TDI (OH/NCO = 1/1), cast from 20 wt% solutions in DCE, at 80°C for 72 h, and then at 80°C under vacuum (30 inHg) for 24 h. Sol contents, discussed later, indicated that this procedure provided nearly complete network formation. Generally, gel times were between 0.5 and 2 h at a catalyst concentration of 0.1 wt% based on total solids content. Thermal analysis results for the networks, also in *Table 3*, showed that all were amorphous and had a wide range of glass transition temperatures, varying from −44.7 to 51.3°C. The high degree of crosslinking increased the glass transition temperatures by an average of 25.6°C compared to the parent polyester oligomer. Only two of the networks, PL and PLTMC, were rigid with T_gs at or above room temperature. The rest were above their T_gs at room temperature and were highly flexible.

In spite of the differing compositions of the networks, equilibrium swelling results (*Table 4* and *Figure 2*) showed that the equilibrium degree of swelling, measured in terms of volume fraction of solvent or per cent weight gained at equilibrium, correlated simply with the $M_{n(OH)}$ of the prepolymer, with the exception of the PGTMC network. The low degree of swelling exhibited by the latter network may be due to a less favourable interaction parameter between it and DCE.

Tensile tests (*Figure 3* and *Table 5*) showed that the PL homopolymer and the PLTMC copolymer networks have the highest tensile strengths of 49.60 and 41.27 MPa, respectively, and that all others display tensile strengths of 12.46 MPa or less. The values discussed above and those listed in *Table 5* are average values of no less than three individual runs, whereas the curves shown in *Figure 3* represent an individual run for each of the networks. Compared to the PL homopolymer network, which was a brittle glass, the PLTMC copolymer network (54.7% TMC) was a ductile plastic exhibiting increased flexibility, a much higher per cent strain at break, a distinct yield point, and only a moderate decrease in tensile strength. It is interesting to compare the PLTMC and PCL copolymers, which are both composed of approximately 45% D,L-lactide yet show significant differences in their tensile properties. The rigid poly(ester–carbonate) structure resulted in a network with a much higher modulus and tensile strength than the PCL network. The latter, essentially an elastomer, displayed a strain at break of 861.0% compared to 215.9% for PLTMC. The arrows pointing from the PCV and PC labels in *Figure 3* indicate the break points for the samples. The break point for PCV is indicated to be at the centre of the PC curve owing to the coincidence of the stress–strain curves for these samples.

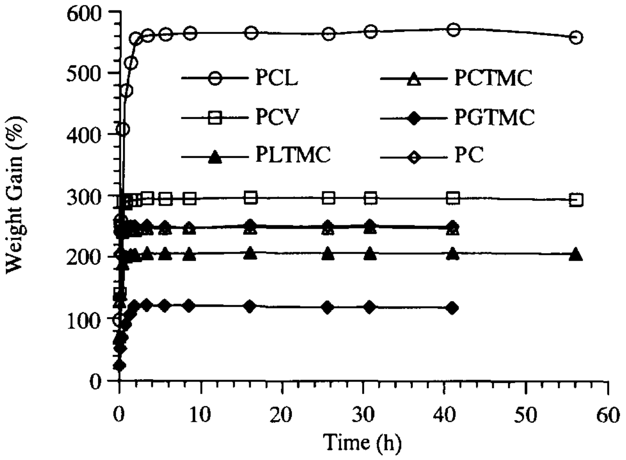

Figure 2 Equilibrium swelling curves for triol–TDI poly(ester–urethane) networks

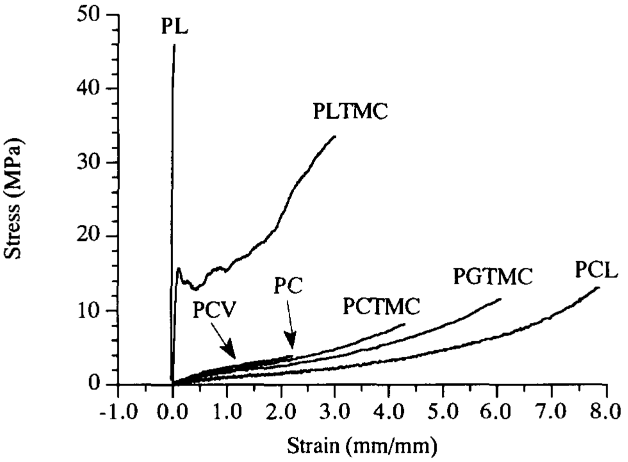

Figure 3 Stress *versus* strain curves for polyester and poly(ester–carbonate) triol–TDI urethane networks

Table 4 Equilibrium swelling results for triol–TDI networks

Sample	Density (g cm^{-3})	Weight gain (wt%)	V_1[a] (cm^3)	V_2[b] (cm^3)
PCL	1.21	563.5	0.84	0.16
PCV	1.17	294.5	0.73	0.27
PLTMC	1.24	206.3	0.67	0.33
PCTMC	1.20	246.7	0.70	0.30
PGTMC	1.36	119.7	0.56	0.44
PC	1.15	249.6	0.70	0.30

[a] Volume fraction of 1,2-dichloroethane
[b] Volume fraction of network

Table 5 Effect of composition and sol content on the tensile properties of triol–TDI networks

Sample	Sol content (%)	Modulus (MPa)	Strain at break (%)	Tensile strength (MPa)
PCL	1.50	1.36	861.0	12.46
PCV	1.74	3.13	197.5	3.22
PLTMC	2.44	766.71	215.9	41.27
PCTMC	1.83	3.01	339.6	6.22
PVTMC	–	–	–	–
PGTMC	4.81	2.14	606.0	9.22
PC	1.99	3.21	249.0	4.40
PL	–	417.49	6.6	49.60

Comparison of the networks containing ε-caprolactone (i.e. PCL, PCV, PCTMC and PC) showed that their rankings in terms of tensile strength and strain at break were the same, while the modulus was uniformly 3 MPa with the exception of PCL, which showed a modulus of 1.36 MPa. The significantly lower modulus of the PCL network may reflect the fact that the PCL polyol prepolymer was of significantly higher molecular weight than the other prepolymers, all of which were in a relatively narrow range of molecular weights centred at about 1350 g mol^{-1}. When these results were considered in view of the T_gs of the networks, it was observed that the tensile strength and per cent strain at break decreased with decreasing T_g, even though all T_gs were below room temperature.

Degradation of 0.5 in diameter discs in phosphate-buffered saline (pH = 7.4) indicated that the PGTMC and PCL networks display the most significant weight uptake of buffer solution and network weight loss, as determined from the hydrated and vacuum-dried discs, respectively (*Figures 4* and *5*). It was expected that the PGTMC network would degrade most rapidly, since glycolide is known to have a very fast degradation rate[16,17]. In this case, the observed degradation rate is certainly enhanced owing to the amorphous nature of the network, whereas degradation studies on glycolide-containing polymers have traditionally been performed on semi-crystalline materials. It was also expected that the PGTMC network would degrade faster than the PCL network, owing primarily to the greater hydrophobicity of lactide units compared to glycolide units.

The PCL network degraded faster than the PLTMC network; the lower T_g of the former network caused it to swell more easily, allowing a greater rate of buffer uptake, and therefore a higher rate of hydrolysis and subsequent weight loss. The other networks containing ε-caprolactone showed essentially no weight gain or loss during a period of 58 days in buffer solution. PC, poly(δ-valerolactone), and PTMC homopolymers have very slow degradation rates; therefore, copolymers prepared from 50/50 feed compositions of these monomers would be expected to yield networks with similar degradation profiles.

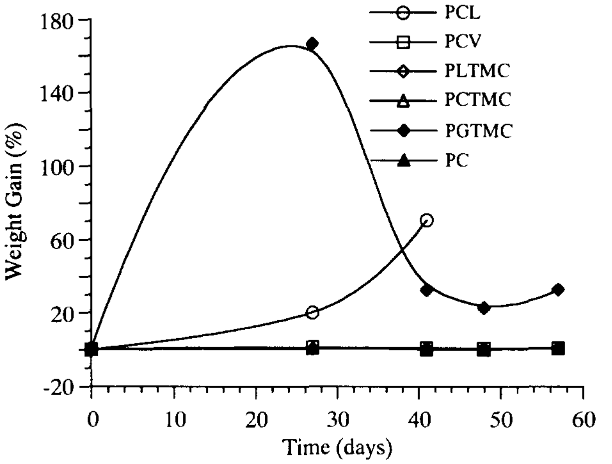

Figure 4 Weight uptake of buffer solution for network discs degraded in phosphate-buffered saline (pH = 7.4)

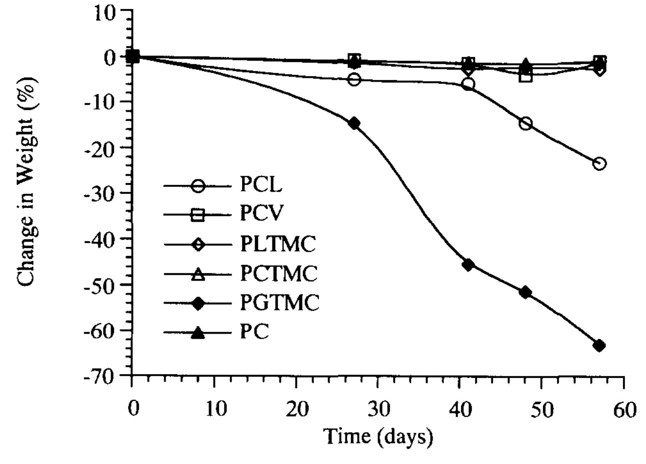

Figure 5 Change in weight of network discs degraded in phosphate-buffered saline (pH = 7.4)

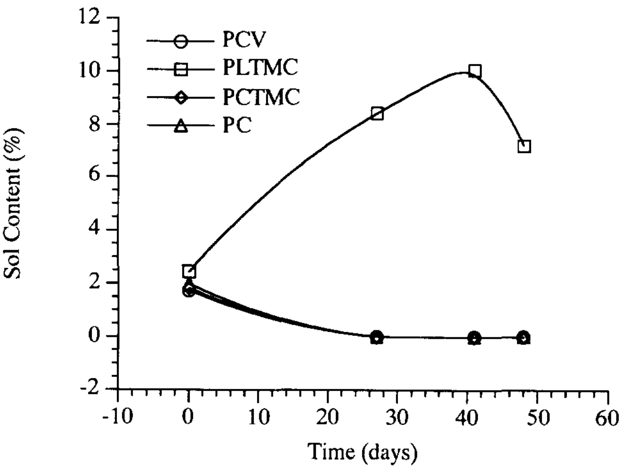

Figure 6 Degraded network disc sol contents *versus* time

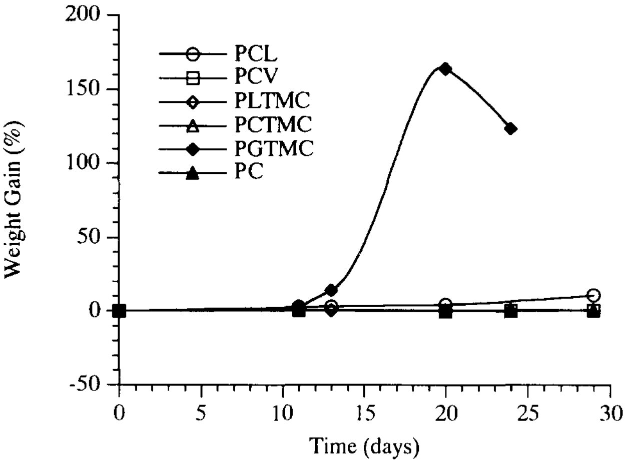

Figure 7 Weight uptake of buffer solution for network microdogbones degraded in phosphate-buffered saline (pH = 7.4)

Since degradation was being measured solely by weight loss, it was possible that degradation processes which produce water insoluble products could have been occurring undetected. Thus, the discs being subjected to degradation for the four slowest-degrading networks, PCV, PLTMC, PCTMC, and PC, were extracted with DCE to monitor the change in sol content during degradation. *Figure 6* shows that the PCV, PCTMC and PC networks lost only their initial sol fractions, and therefore showed no production of extractables during degradation. However, the PLTMC network showed an initial increase in sol content during the first 41 days, followed by a slight decrease thereafter, consistent with *Figure 5*, which indicated a slight weight loss during the latter part of the 58 day degradation period. These results indicate that for the PLTMC network, degradation via hydrolytic chain scission did indeed occur within the network without apparent weight loss, presumably because the degradation products were only marginally soluble in the buffer solution.

The evolution of the tensile properties of the PC homopolymer and the copolymer networks with time of degradation was studied using microdogbone tensile bars immersed in phosphate-buffered saline at 37°C. Samples were removed at intervals, weighed to determine water uptake, tensile tested in the hydrated state, and thereafter vacuum desiccated and reweighed to determine weight loss due to degradation. As shown in *Figure 7*, the weight gain due to water absorption for the slowly degrading PCV, PLTMC, PCTMC, and PC networks was similar to that of the discs described earlier, i.e. the weight gain was negligible within a period of 25–30 days. Although slight differences were noted, the general behaviour of the most hydrophilic network, PGTMC, was essentially the same for disc and dognone geometries. However, the PCL microdogbone samples showed less water uptake than the PCL discs. Weight loss, measured after tensile testing and vacuum desiccation (*Figure 8*), was also quite consistent with the results from the disc experiments with the exception of the PCL network samples, which showed very little weight loss after 29 days.

The plot of tensile strength *versus* time of immersion for each network is shown in *Figure 9*. As seen earlier in *Figure 3*, the PLTMC network displayed an initial tensile strength four times higher than any other sample (with the exception of the PL network, which unfortunately was not subjected to a study of tensile strength *versus* time). It also showed the most dramatic absolute loss in tensile strength over time, and showed one of the higher fractional losses, along with the PGTMC and PCL networks. It is interesting, however, that this decrease in tensile strength of the PLTMC network was not accompanied by a corresponding change in the strain at break, which showed only a small increase (*Table 8*). *Tables 6* to *11* list the degradation time, weight gain and weight loss, modulus, strain at break, and tensile strength values for all of the networks subjected to the tensile degradation analysis.

The actual stress *versus* strain curves for PLTMC, shown in *Figure 10*, show a dramatic drop in tensile properties for the network over a period of 24 days of degradation. It is interesting to note that the modulus

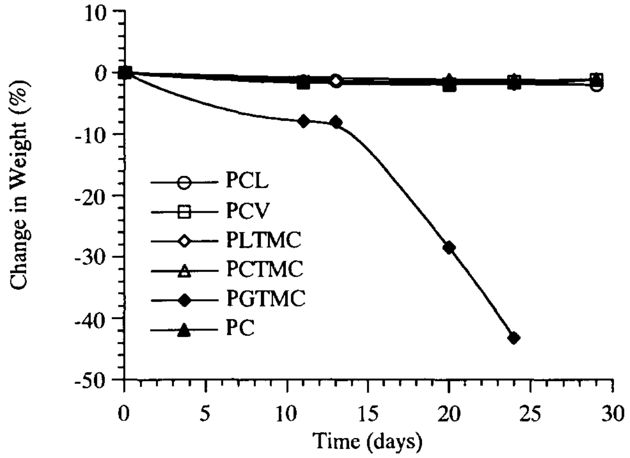

Figure 8 Change in weight of network microdogbones degraded in phosphate-buffered saline (pH = 7.4)

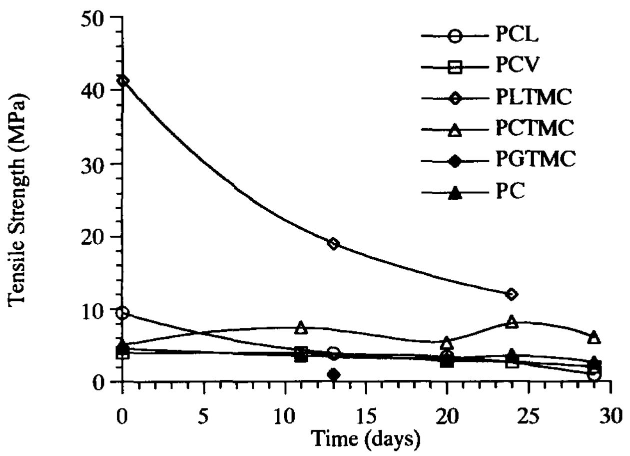

Figure 9 Tensile strength *versus* time for network microdogbones degraded in phosphate-buffered saline (pH = 7.4)

Table 6 Effect of degradation in phosphate-buffered saline (pH = 7.4) on the mechanical properties of a PCL 55/45 polyester triol–TDI urethane film

Time (days)	Weight gain (%)	Weight loss (%)	Modulus (MPa)	Strain at break (%)	Tensile strength (MPa)
0	0.00	0.00	1.44	929	9.50
13	3.09	1.31	0.33	1091	3.88
20	4.45	1.70	0.46	1323	3.42
29	10.71	1.89	0.26	1017	0.94

Table 7 Effect of degradation in phosphate-buffered saline (pH = 7.4) on the mechanical properties of a PCV 60/40 polyester triol–TDI urethane film

Time (days)	Weight gain (%)	Weight loss (%)	Modulus (MPa)	Strain at break (%)	Tensile strength (MPa)
0	0.00	0.00	2.68	289	3.97
11	0.39	1.57	3.29	261	3.95
20	0.03	1.85	2.80	223	2.84
24	0.19	1.46	3.1	196	2.69
29	0.59	1.09	2.48	153	1.97

Table 8 Effect of degradation in phosphate-buffered saline (pH = 7.4) on the mechanical properties of a PLTMC 45/55 poly(ester–carbonate) triol–TDI urethane film

Time (days)	Weight gain (%)	Weight loss (%)	Modulus (MPa)	Strain at break (%)	Tensile strength (MPa)
0	0.00	0.00	767	215	41.27
13	0.51	1.33	319	295	18.98
24	0.17	1.81	128	307	11.96

Table 9 Effect of degradation in phosphate-buffered saline (pH = 7.4) on the mechanical properties of a PCTMC 55/45 poly(ester–carbonate) triol–TDI urethane film

Time (days)	Weight gain (%)	Weight loss (%)	Modulus (MPa)	Strain at break (%)	Tensile strength (MPa)
0	0.00	0.00	2.97	247	5.06
20	0.74	1.11	2.36	379	5.38
24	0.69	1.20	2.46	493	8.21
29	0.75	0.97	2.51	423	6.07

Table 10 Effect of degradation in phosphate-buffered saline (pH = 7.4) on the mechanical properties of a PGTMC 45/55 poly(ester–carbonate) triol–TDI urethane film

Time (days)	Weight gain (%)	Weight loss (%)	Modulus (MPa)	Strain at break (%)	Tensile strength (MPa)
0	0.00	0.00	2.08	589	4.67
13	13.9	8.00	0.1	930	0.93
20	164.1	28.40	–	–	–

Table 11 Effect of degradation in phosphate-buffered saline (pH = 7.4) on the mechanical properties of a PC polyester triol–TDI urethane film

Time (days)	Weight gain (%)	Weight loss (%)	Modulus (MPa)	Strain at break (%)	Tensile strength (MPa)
0	0.00	0.00	3.46	273	4.61
11	0.37	1.61	3.00	220	3.54
20	0.23	1.41	3.58	183	3.13
24	0.11	1.60	3.29	215	3.62
29	0.48	1.06	2.95	185	2.61

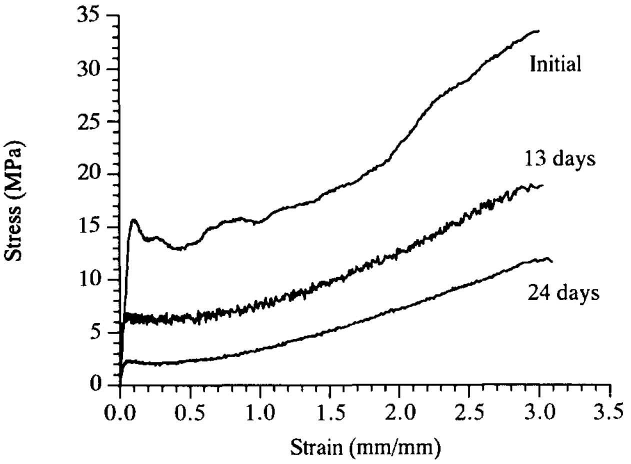

Figure 10 Stress *versus* strain curves of PLTMC network microdogbones degraded in phosphate-buffered saline (pH = 7.4)

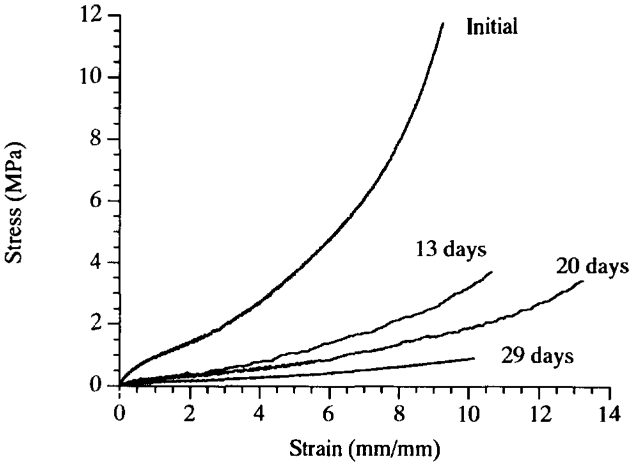

Figure 11 Stress *versus* strain curves of PCL network microdogbones degraded in phosphate-buffered saline (pH = 7.4)

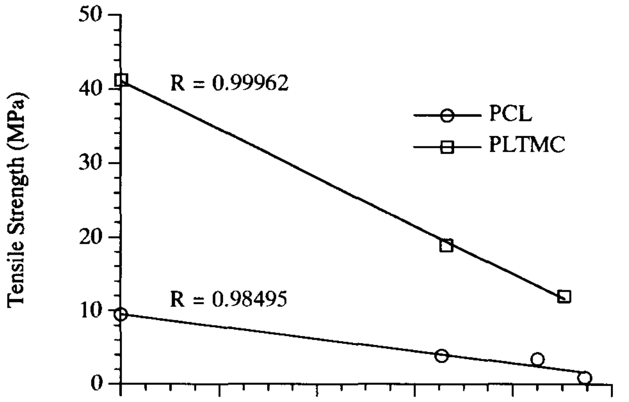

Figure 12 Tensile strength *versus* weight loss for PCL and PLTMC networks degraded in phosphate-buffered saline. The value for the correlation coefficient *R* is indicated next to each curve

and tensile strength decreased by approximately 50% over the first 13 days and then an additional 50% after a total of 24 days. The PCL network also showed a significant decrease in tensile strength during degradation owing to the quantity of buffer absorbed by the network, which ultimately resulted in hydrolysis of crosslinked chains (*Figure 11*). Most of this decrease in tensile strength occurred during the initial 13 days of degradation, followed by a more gradual loss thereafter.

Inspection of the tensile strength *versus* weight loss curves for the PCL and PLTMC networks, shown in *Figure 12*, reveals a nearly linear loss of strength with respect to loss of weight during the first month of degradation. Such uniform behaviour, indicating a homogeneous degradation process, was predicted to occur for these network samples. Semi-crystalline materials, which possess microheterogeneous morphologies, may not display such uniform strength *versus* weight loss curves, since the amorphous 'tie' molecules may tend to degrade much faster than the crystalline regions, leading to more catastrophic losses in strength. Curves like those in *Figure 11* could not be generated for the other networks because their rates of degradation were either much too fast (PGTMC) or too slow (PC, PCTMC, and PCV).

There was such a small change in the tensile properties of the PCV and PC networks that their stress–strain curves taken at various stages of degradation nearly coincided, only showing small differences in tensile strength and strain at break. The PCTMC network showed an increase in the strain at break with respect to degradation time, resulting from small amounts of chain scission (*Table 9*). The tensile curves for degradation samples of the PCTMC network also follow the same trend, nearly coinciding completely except for the small differences in tensile strength and strain at break. The PCV, PCTMC, and PC networks did not have significant changes in sol content (as measured by DCE extraction) during degradation, which indicates that the polymer backbones were not degraded significantly. The PGTMC network degraded so quickly that it had lost almost all of its tensile strength and modulus after 13 days in the buffer solution; in fact, the network had absorbed so much buffer solution that it was swollen to at least five times its original size.

All of the hydrated samples were very elastic, and in most cases they relaxed back to their original dimensions within approximately 5–15 min after tensile testing. The PLTMC network appeared to strain crystallize during tensile testing; however, this was not confirmed experimentally.

CONCLUSIONS

We have shown that trifunctional, hydroxy-terminated polyester and poly(ester–carbonate) prepolymers can be readily prepared from the triol-initiated, bulk homopolymerization and copolymerization of cyclic lactones and carbonates, and it is suggested that these prepolymers are potentially useful building blocks for the synthesis of amorphous, biodegradable networks and composite matrices for use in biomedical applications. The prepolymers had a wide range of T_gs within the rubbery regime from −64.8 to 14.0°C, and several were semi-crystalline; however, the urethane networks prepared from these prepolymers were completely amorphous and displayed T_gs which were on the average 25.6°C higher than those of the parent prepolymers. Equilibrium swelling results indicated that the solvent uptake of the networks correlated well with the prepolymer $M_{n(OH)}$, with the exception of the PGTMC network.

Tensile test results showed that the PL and PLTMC networks had the highest tensile strengths - 49.60 and 41.27 MPa, respectively. These networks also displayed the highest T_gs (51.3 and 21.3°C), whereas the other networks were highly flexible with low tensile strengths (12 MPa or less), consistent with T_gs well below room temperature. Incorporation of a higher percentage of D,L-lactide into the PLTMC copolymer may produce a network that has a tensile strength closer to that of the PL network, while retaining sufficient flexibility.

Hydrolytic degradation of the various networks showed that the glycolide-containing polymer had the fastest degradation rate. Those containing ε-caprolactone have very slow degradation rates, as expected. The PCL and PLTMC networks, with intermediate rates of degradation, showed a linear loss of strength with loss of weight during the first 30 days of degradation. This type of behaviour was expected for amorphous materials such as these, and although the absolute rate of degradation, for a given chemical composition, is faster than that for semi-crystalline materials, the degradation profile is more uniform, and this may be advantageous in certain biomedical applications.

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