Published in final edited form as:

Polymer (Guildf). 2011 September 29; 52(21): 4920–4927. doi:10.1016/j.polymer.2011.08.033.

Thermo-Mechanical Properties of Semi-Degradable Poly(β -amino ester)-co-Methyl Methacrylate Networks under Simulated Physiological Conditions

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Abstract

Poly(β -amino ester) networks are being explored for biomedical applications, but they may lack the mechanical properties necessary for long term implantation. The objective of this study is to evaluate the effect of adding methyl methacrylate on networks' mechanical properties under simulated physiological conditions. The networks were synthesized in two parts: (1) a biodegradable crosslinker was formed from a diacrylate and amine, (2) and then varying concentrations of methyl methacrylate were added prior to photopolymerizing the network. Degradation rate, mechanical properties, and glass transition temperature were studied as a function of methyl methacrylate composition. The crosslinking density played a limited role on mechanical properties for these networks, but increasing methyl methacrylate concentration improved the toughness by several orders of magnitude. Under simulated physiological conditions, networks showed increasing toughness or sustained toughness as degradation occurred. This work establishes a method of creating degradable networks with tailorable toughness while undergoing partial degradation.

Keywords

Poly(β -amino ester); Degradable; Toughness

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1. Introduction

Methacrylate copolymer networks represent a class of materials that can exhibit a wide range of mechanical properties from rubbery gels to brittle glasses. Mostly amorphous, these networks' thermo-mechanical properties are mainly governed by two parameters, the glass transition temperature (T_g) and the crosslinking density [1-5]. For instance, the crosslinking density can influence the rubbery modulus of the network and can be changed by altering the functionality and molecular weight of the crosslinker [1, 6]. These networks can be further altered with additional functional groups that allow for partial or full degradation of the polymer networks making them suitable candidates for tissue scaffolds or drug releasing materials [7-10]. This added dimension of functionality allows for a broad range of mechanical properties and degradation processes. Toughness is an important parameter that is not often considered in regards to developing polymer-based biomaterials that undergo extensive mechanical loading and deformation *in vivo*. Intrinsic toughness is the energy required to induce failure in a material, which can be measured by the area under a stress-strain curve. While matching the modulus of the implant and native tissue is important, the material must be tough in order to survive *in vivo* and not fail prematurely. Previous ways of improving mechanical properties of crosslinked networks have included using particle reinforcement, lowering crosslinking density, hyperbranching, or using double networks [11-14].

Understanding the intrinsic structure-property relationships that control mechanical properties is critical for developing biomedical materials. General relationships between chemical structure and the T_g exist where long, flexible sidegroups decrease T_g and short bulky sidegroups increase T_g by increasing steric hindrance. Additionally, failure strain is inversely related to the rubbery modulus or crosslinking density [15]. As crosslinking density decreases, chemical structure has a greater influence on mechanical properties, thus chemical structure can dictate both T_g and failure strain [16]. In previous studies, failure strain was shown to be highly temperature dependent, where a maxima in deformation capacity is present at a temperature beneath the T_g as determined by $\tan \delta$ [16-17]. Because a material's toughness is strongly dependent on failure strain, as well as its strength, a network's toughness also experiences a similar maxima that is temperature-dependent [18]. At low temperatures below T_g , the material is brittle with high strength and low strain capacity. At high temperatures, the material is rubbery, thus may have moderate strain capacity, but low strength. At a temperature near T_g , the network is viscoelastic and thus has some chain mobility allowing for deformation without significant loss in rigidity allowing for maximum toughness. It has been shown that toughness is partially dependent upon the chemical structure, where aromatic sidegroups cause higher toughness than aliphatic sidegroups [16]. Similar to changing temperature, the presence of water may increase or decrease the networks' mechanical properties in a manner dependent upon the chemical structure of the network. For instance, the degree of swelling and thermodynamic state of water shifts the T_g , which subsequently changes the temperature at which maximum toughness can be achieved under aqueous conditions [19].

Poly(β -amino ester) networks have been proposed as a family of synthetic biodegradable networks that undergo hydrolytic cleavage. Due to the diversity of chemical structures, a wide range of mechanical properties and degradation rates are possible [1]. Briefly, the biodegradable crosslinkers form from a step-growth polymerization of a diacrylate with a primary amine, where the diacrylate is in molar excess. Then a free-radical photopolymerization of the crosslinkers with acrylate ends results in a crosslinked network consisting mostly of difunctional molecules. The glass transition temperature of these networks is often sub-ambient due to the high molecular weight of the crosslinkers and the lack of steric hindrance due to a lack of rigid sidegroups. Increasing the diacrylate to amine

ratio lowers the macromer molecular weight, thus increasing the crosslinking density and modulus [20-21]. There has been limited evaluation of the mechanical properties of these networks to date, as they do not possess the inherent toughness for use as load bearing materials. Profiles of modulus under immersed conditions have been studied, where the modulus is shown to decrease due to the degradation of the crosslinks [20]. Typically, the testing has been done in compression; however, the modulus decrease has been similarly observed in tension as well [20]. Specifically, the failure strain and toughness have yet to be systematically reported for these polymer networks.

The current poly(β -amino ester) networks have low T_g and high crosslinking density, thus may be lacking in sufficient toughness to survive implantation in load bearing applications. In order to increase their mechanical properties, for some applications, additional monomers will be added to the network to increase T_g and decrease crosslinking density. Therefore, the purpose of this work was to determine the effect of varying methyl methacrylate (MMA) concentration on the thermo-mechanical properties of poly(β -amino ester)-co-MMA networks under simulated physiological conditions. Crosslinkers of varying diacrylate to amine ratio were used to determine the effect of varying the crosslinking density on these poly(β -amino ester)-co-MMA networks. The effect of MMA concentration on glass transition temperature, rubbery modulus, failure strain, toughness, and degradation is studied through systematic variation of MMA concentration. Select networks were tested in a simulated physiological environment in order to understand the effect of degradation on the mechanical properties. The use of MMA to increase the T_g and decrease the crosslinking density should enhance the toughness of the networks during degradation, but will also render the networks only partially degradable.

2. Experimental

2.1 Materials

Hexanediol diacrylate (HDDA), 3-methoxypropylamine (3MOPA), methyl methacrylate, and 2-hydroxy-1-[4-(hydroxyethoxy)phenyl]-2-methyl-1-propanone (Irgacure 2959) were purchased from Sigma Aldrich and used as received.

2.2 Synthesis

Following previous work, HDDA was mixed with 3MOPA at varying molar ratios from 1.05:1 to 1.20:1 at 200 rpm at 90°C for 24 hours on a JKEM reaction block (RBC-20 with BTS-1500 shaker) [21]. This step-growth polymerization produced the biodegradable crosslinkers. The macromers/crosslinkers will be stated by HDDA then their molar ratio, such that HDDA:3MOPA 1.15:1 will be called HDDA 1.15. The resulting crosslinkers were either stored at 4°C or mixed with varying weight ratios (0 to 75 wt%) of MMA. Then, 0.5 weight percent Irgacure 2959 was added for photopolymerization. The mixtures were injected between glass slides with 1mm thick glass spacers, and polymerized for 45 minutes by a UVP Blakray lamp ($\sim 10\text{mW}/\text{cm}^2$) to form chemically crosslinked networks.

2.3 Thermo-mechanical Characterization

Dynamic mechanical analysis (DMA) (DMA Q800 TA Instruments) was run on rectangular samples (25mm $\times$ 4mm $\times$ 1mm) at a strain of 0.1% at 1 Hz. The thermal sequence was to equilibrate at -100°C, isotherm 2 minutes, then ramp at a rate of 3°C/min to 200°C. The glass transition temperature (T_g) was defined at the peak of the tan delta [22]. The rubbery modulus was defined as the modulus in the rubbery plateau at a temperature of T_g + 75°C. From the theory of rubber elasticity, the crosslinking density, ν , can be determined from Equation 1, where E is the modulus in the rubbery regime at a temperature T (Kelvin), R is

the gas constant, assuming the Poisson's ratio is 0.5 [23]. The mean $\pm$ standard deviation with $n=3$ is given for each data point.

$$\nu = \frac{E}{3RT}$$

Equation 1

Differential scanning calorimetry (DSC) (DSC Q100 TA Instruments) was run to determine the T_g of samples that had been degraded in phosphate buffered saline (PBS), pH 7.4, and subsequently dried for 24 hours. The thermal sequence was to equilibrate at -90°C , isotherm for 2 minutes, then ramp at a rate of $3^\circ\text{C}/\text{min}$ to 200°C . The T_g was determined from the midpoint of the intersecting line method of the second order thermal transition. The mean $\pm$ standard deviation with $n=3$ is given for each data point.

2.4 Mechanical Behavior

Strain to failure tensile tests were performed on a MTS Insight 2 with a 100N load cell using ASTM D-638 Type IV half-sized dog-bone samples at a strain rate of 10^{-3} s^{-1} . A thermal chamber (Thermcraft, Inc., model LBO-14-8-5.25-1X-J8249_1A) was used to test samples in a dry atmosphere at 37°C . Dog-bones were soaked in PBS in an incubator at 37°C for up to 8 weeks, and then strained to failure in a custom environmental chamber that was filled with PBS and held at 37°C for testing under simulated physiological conditions. Toughness was calculated using the trapezoidal rule from the area under the stress strain curve and given in MJ/m^3 . Modulus was determined from the slope of the initial linear region of the stress-strain curve. If the material was in the rubbery regime, then crosslinking density was calculated via Equation 1. The mean $\pm$ standard deviation with $n=4$ is given for each data point.

2.5 Degradation and Water Content

In order to determine the degradation rate and water content, each material was soaked for varying amounts of time. Each material was cut from a 1 mm thick sheet into a 1 cm^2 square and placed into a well plate filled with PBS. The well plates remained in an incubator at 37°C . Samples were patted dry to remove excess water to obtain the wet sample mass. The samples were dried for 24 hours and the mass taken. The water content of each material will be defined by Equation 2, where M_{wi} is the wet mass at time i and M_{di} is the mass at time i after 24 hours of drying. The mass loss will be defined by Equation 3, where M_{di} has been previously described and M_o is the initial mass.

$$\text{WaterContent} = \frac{M_{wi}}{M_{di}} - 1$$

Equation 2

$$\text{MassLoss} = 1 - \frac{M_{di}}{M_o}$$

Equation 3

3. Results

3.1 Thermo-mechanical Properties

The initial T_g and storage modulus at different temperatures were characterized via DMA. The T_g as a function of wt% MMA is shown in Figure 1. By adding MMA to the degradable

crosslinkers, the T_g increased from -44°C to 80°C in a manner independent of the diacrylate to amine ratio.

The storage modulus at 37°C increased over two orders of magnitude as MMA concentration increased as shown in Figure 2A indicative of the networks transitioning from a rubbery state to a glassy state. At MMA concentrations less than 45 wt%, the storage moduli of the networks were affected by their diacrylate to amine ratio. The modulus increased as the diacrylate to amine ratio increased in this lower MMA concentration region. As the MMA concentration increased above 45 wt%, the diacrylate to amine ratio did not influence the storage modulus at 37°C . In contrast, the storage modulus at a temperature of $T_g+75^{\circ}\text{C}$ decreased as wt % MMA increased, as shown in Figure 2B. Also, the crosslinking density at a temperature of $T_g+75^{\circ}\text{C}$ decreased as wt% MMA increased, as shown in Figure 2C. The crosslinking density values are similar to other poly(β -amino ester) networks [23]. While Figure 2A shows the effect of MMA concentration at one temperature, Figures 2B and 2C illustrate the effect of MMA concentration on the network structure. At a given MMA concentration, the modulus decreased due to the decrease in crosslinking density as the diacrylate to amine ratio decreased. Similarly, at a constant diacrylate to amine ratio, the modulus decreased as the crosslinking density decreased due to higher percentages of MMA.

3.2 Mechanical Properties

The initial mechanical properties of these networks were characterized using tensile tests at 37°C under dry conditions. Representative stress-strain curves of the HDDA 1.15 networks with varying wt% MMA are shown in Figure 3. As wt% MMA increased, the failure strain increased, but there was little variation in the modulus of the material until high concentrations of MMA were used (75%), where the material acted in an elastic-plastic manner as demonstrated by a high modulus and yield point.

Mean toughness and elastic modulus values were derived from stress-strain curves. Toughness as a function of elastic modulus was plotted for networks composed from one of the four HDDA crosslinkers at varying wt% MMA and is shown in Figure 4A. As the wt% MMA increased, the toughness increased by nearly three orders of magnitude due to an increase in failure strain. As the wt% MMA increased, the modulus increased due to the transition from a rubbery to a glassy state. The toughness as a function of the difference between the T_g of each network and testing temperature ($T-T_g$) is shown in Figure 4B. As the wt% MMA increased, the toughness increased due to the shift in the glass transition temperature in relation to 37°C . As the T_g shifted closer to and beyond 37°C ($T-T_g = 0^{\circ}\text{C}$), the toughness increased at least one order of magnitude.

3.3 Degradation and Mechanical Properties under Simulated Physiological Conditions

The degradation profiles of networks composed of HDDA 1.15 with varying wt% MMA are shown in Figure 5A. The mass loss decreased as the wt% MMA increased. The amount of water uptake, as shown in Figure 5B, was 20-25% of the original weight of the network, where the 55%MMA and 75%MMA networks have decreased average water contents compared to the networks with lower %MMA. The T_g as measured by DSC over 8 weeks of degradation is shown in Figure 6. The extent of change in T_g was dependent on composition. The 0% MMA networks changed by an average of 2°C in T_g over the 8 weeks. The 35%, 45%, and 55%MMA networks showed an average upward shift by 18°C , 10°C , and 20°C , respectively. The 75%MMA showed a decrease in T_g by 7°C after 8 weeks.

There is a difference between the initial T_g values from the DMA and the initial T_g values at 0 week from the DSC. DMA is a dynamic mechanical test that measures changes in storage

and loss modulus, but DSC is a static test, which measures the change in heat capacity. DSC allows for tracking the evolution of T_g (e.g. changes in T_g from a reference point). The T_g is a range of temperatures, and the value depends upon the testing method and the convention for measuring the T_g . The assignment of T_g varies even for the DMA, where the peak of tan delta, onset of storage modulus drop, or peak of loss modulus may be used. This can vary the reported value of T_g by 25°C for the same polymer since the tan delta peak is at a higher temperature than the storage modulus or loss modulus points [24]. Values from the DSC agree with the loss modulus values, thus the lower values for the DSC compared to the DMA [25]. DMA allows for confirmation of the physical state of the polymer (e.g. rubbery, viscoelastic, glassy) at the temperature of interest (37°C). Equation 1 can be applied to obtain information on the network's structure, but only if the material is in the rubbery regime.

Representative stress-strain curves for 5 HDDA 1.15-co-MMA networks of varying wt% MMA over 8 weeks of immersion under simulated physiological conditions are shown in Figure 7. HDDA 1.15 + 0% MMA has low stress and failure strain under immersed conditions (Figure 7A). As time increased, its failure strain increased and modulus decreased allowing for greater deformation, but eventually the failure strain and failure strength decreased by 8 weeks. The 35% and 45% MMA networks behave in a similar manner, where the modulus decreased and the failure strain increased up to 6 weeks, but at 8 weeks, the modulus increased. The 55% MMA network showed minimal change in mechanical properties over time, maintaining high failure strains for all 8 weeks. The 75% MMA network shows initial softening as the modulus decreases and the failure strain increases during the initial weeks. At 8 weeks, the material returns to a more glassy state, absent of yielding and plastic deformation.

Failure strain and toughness profiles as a function of time for the HDDA-co-MMA networks are shown in Figure 8. As wt% MMA increased, failure strain increased by two orders of magnitude. As immersion time increased, the failure strain behavior was composition dependent. The 0% MMA network saw an increase in failure strain at 4 weeks from 2% to 14%, but then a decline by 8 weeks to 4%. The 35% and 45% MMA networks show an overall average increase in failure strain during the 8 weeks from 14% to 29% and 34% to 47%, respectively. The failure strain of the 55% MMA network remained nearly unchanged during the 8 weeks. The failure strain of the 75% MMA network had an increase from 4 to 6 weeks, but then a decrease by an order of magnitude at 8 weeks from 240% to 5%. As the wt% MMA increased, the toughness increased for all networks. The effect of immersion time on toughness varied depending on the wt% MMA. For example, the average toughness of the 35% and 45% MMA networks increased over 8 weeks from 0.06 to 0.15 MJ/m^3 and 0.2 to 0.47 MJ/m^3 , respectively. The 0% MMA network showed an increase in toughness at 4 weeks then a decrease by 8 weeks, similar to its failure strain profile. The 75% MMA network had the highest toughness for the first 6 weeks (12-27 MJ/m^3), but decreased at 8 weeks to 0.4 MJ/m^3 leaving the 55% MMA network with sustained average toughness near 2 MJ/m^3 for the 8 weeks. The toughness of the 75% MMA network approached values close to pure PMMA [18].

Since the T_g of the 0% MMA and 35% MMA networks were well below the testing temperature ($T_g \ll 37^\circ\text{C}$), the networks were in the rubbery regime. Select networks' structural changes were characterized by a modulus profile, and resulting crosslinking density profile via Equation 1, as shown in Figures 9A and 9B, respectively. The 0% MMA network showed a decrease in modulus and crosslinking density over the 8 weeks, which is expected for a network made of only degradable crosslinkers. The 35% MMA network showed a decrease in modulus and crosslinking density for the first 6 weeks; however, at the

8 week timepoint the modulus increased from 1.4 to 3.6 MPa and the crosslinking density increased from 187 to 469 mol/m³, respectively.

4. Discussion

In this study, the effect of varying the MMA concentration on the mechanical properties of poly(β -amino ester) networks was explored. Biodegradable acrylate-based crosslinkers of varying diacrylate to amine ratios were used to determine if the crosslinking density had an effect on the mechanical properties. The results show that adding MMA can significantly change the thermo-mechanical properties of the networks, particularly in the presence of PBS. The poly(β -amino ester) system of HDDA:3MOPA was chosen because its degradation rate and elastic modulus can be tailored independently and its potential biocompatibility has been previously demonstrated [20-21]. MMA was chosen as the linear chain builder because of its wide clinical use as bone cement [26]. While, PMMA is not hydrolytically degradable, mass loss was expected due the biodegradable HDDA crosslinker. The impact of wt% MMA on poly(β -amino ester) networks' mechanical and thermal properties, especially under simulated physiological conditions, will be discussed below.

DMA was used as a screening method to evaluate the initial thermo-mechanical properties of the networks. Since a rule of mixtures often applies to Tg of acrylate networks, adding MMA is expected to increase the Tg, thus allowing the networks to transition from a rubbery to glassy state, as seen in Figure 1 and Figure 2A. Similar poly(β -amino ester) networks used tert-butyl acrylate to increase Tg above body temperature, but required near 80 wt% tert-butyl acrylate to achieve this high Tg, which would significantly hinder degradation [27]. Use of monomers, such as MMA, with Tg above 100°C allows for Tg to reach body temperature without high concentrations of single functional monomer. Each network, at a constant wt% MMA had the same Tg regardless of the diacrylate to amine ratio of the HDDA crosslinker. Changing crosslinking density by changing the diacrylate to amine ratio did not alter Tg of the networks; however, the crosslinking density influenced the rubbery modulus, as seen in Figures 2B and 2C. This is expected since crosslinking density is proportional to the modulus for rubbery materials [28]. At a constant diacrylate to amine ratio, increasing the wt% MMA not only increases Tg, but decreases crosslinking density (Figure 2C), which is measured through a decrease of the rubbery modulus in Figure 2B. Lowering the crosslinking density can enhance the mechanical properties by allowing larger strain capacity, and hence toughness [12].

After establishing the changes in Tg and rubbery modulus, further bulk mechanical properties were examined to determine the relationship between MMA concentration and toughness. First, adding MMA to the networks increased the failure strain up to a certain point, thus increasing the toughness, but did not alter the modulus until high concentrations at the glassy state. Thus, the addition of MMA allows for unique tailoring of mechanical properties, where toughness can be maximized within a specific range of moduli. Figure 4A further demonstrates the tailorability of this system, where the diacrylate to amine ratio controls the modulus and the wt% MMA controls the toughness. Further tailoring can be achieved by considering the testing or environmental temperature. Toughness is dependent upon the Tg of the network as seen in Figure 4B. With a set operating temperature and varying Tg, the networks with Tg above the operating temperature had the highest toughness. A previous study has shown that toughness peaks when the Tg is closely aligned with the testing temperature for methacrylate networks [18]. It is expected that if the network's Tg continued to shift above the operating temperature, eventually the material would become brittle and have a lower toughness, such as PMMA which has a toughness near 1 MJ/m³ [18]. Using a pure PMMA network or lightly crosslinked network would not

maximize toughness at 37°C, since PMMA has maximized toughness near 100°C. Therefore, it is critical to tailor both the T_g and the crosslinking density in order to tailor toughness for the desired application.

For the networks under simulated physiological conditions, the HDDA 1.15 + 0% MMA composition with no MMA had limited degradation over 8 weeks, exhibiting only a 20% loss in mass over 12 weeks (Figure 5A) [21]. Networks of 35% and 45% MMA show similar trends in degradation rate. Theoretically, if enough crosslinks degraded, a network that is water soluble would be expected to degrade and erode; however, if the number of degradable crosslinks were not sufficient as possible in the 55% and 75% MMA networks, the networks may not cleave enough to enable complete degradation with non water soluble MMA backbones present. As degradation occurs, albeit slowly due the hydrophobic nature of HDDA based crosslinkers, the network is able to further swell, thus the increase in water content over 8 weeks. This increased water content will allow for further hydrolytic cleavage and increased degradation until the number of hydrolytically cleavable sites has been exhausted. In a previous study, networks with 15% poly(β -amino) ester-co-85% tert-butyl acrylate also experienced slow degradation similar to the 55% and 75% MMA networks due to the limited water uptake [27]. The increase in water content in the networks suggests more degradation is possible as the increased water content further promotes the hydrolytic cleavage of the crosslinkers. Further long term testing will be necessary to reveal the structural changes over extended periods of time.

While many biomedical devices are typically grouped into two categories, degradable or nondegradable, these HDDA-co-MMA networks represent a possible semi-degradable option, which would allow for some degradation for therapeutic agent release while also still maintaining sufficient mechanical properties to offer structural support. Another semi-degradable system was formed from a composite of a nondegradable poly(vinyl alcohol) scaffold with degradable PLGA microspheres for articular cartilage repair [29]. This approach combined nondegradable and degradable components at the microscale, where the current approach creates a semi-degradable polymer at the polymer network-scale. Processing with a porogen for incorporating microspheres would no longer be an issue, since the matrix itself would have degradable linkages to allow tissue incorporation and release of therapeutic agents.

Since the T_g of the networks is intrinsically related to both the chemical structure and the mechanical properties, it was necessary to understand the evolution of T_g during the course of degradation. The T_g increased in several networks due to the degradation of the low T_g component, the biodegradable HDDA crosslinker, thus leaving more MMA with a higher T_g in the network. The 75% MMA network represented a different case, where due to its low degradation and water content, the T_g decreased due to the absorption of water, as previously observed in nondegradable methacrylate based networks [19]. In the other networks, the upward shift in T_g and the decrease in crosslinking density verifies the loss of the degradable crosslinker, but this shift in T_g also has effects on the mechanical properties. The shift in T_g caused an increase in the 35% MMA network's modulus, thus increasing the crosslinking density at 8 weeks (Figure 9). This shift in T_g caused the network to shift from the rubbery regime to the viscoelastic regime even though degradation occurred, which allows for a self-stiffening material.

Instead of experiencing loss in mechanical properties as would be expected, adding MMA to the network resulted in an increase in both modulus and toughness. The combination of having MMA, a non-degradable high T_g monomer, in the network and the loss of the low T_g crosslinker caused the T_g to shift with time allowing the mechanical properties to become enhanced over time in a manner that was composition dependent. For example, the

35% MMA network showed an increase in both Tg and toughness during the 8 weeks, indicating that networks that experience an increase in Tg under simulated physiological conditions can potentially also have enhanced and sustained toughness. The 55% MMA network showed minimal change in mechanical properties over time due to the lack of degradation and low water content. This composition may represent a balance point, where degradation would cause a decrease in modulus, but the increase in Tg would increase the modulus. While networks with moderate concentrations of MMA (35-55%) displayed sustained toughness, the 75% MMA network became brittle by 8 weeks. The loss of the low Tg component and increasing water uptake would have made the structure more similar to pure PMMA, thus the decrease in failure strain and toughness, similar to other methacrylate networks that lost mechanical properties during long term immersion [30].

This study demonstrates that poly(β -amino ester)-co-MMA networks may be used to independently tailor the modulus and toughness, and the choice of linear chain builder and the resulting shifts in Tg can enhance mechanical properties over sustained degradation. Several modifications to these networks could allow for further tailorability. The Tg of the linear chain builder could range from subambient to well over 100°C. This would allow further control of the shift in Tg as the biodegradable crosslinker leaves the network. Other linear chain builders could augment the hydrophilicity and water content, thus increasing the degradation rate, the shift in Tg, and the resulting change in mechanical properties. This method of combining a low Tg degradable crosslinker with a nondegradable linear chain builder establishes a platform for the design of various semi-degradable materials. The biodegradable crosslinkers can be produced in a facile manner and readily mixed with any number of methacrylates to produce semi-degradable networks that would allow possible therapeutic release and sustained mechanical properties in the *in vivo* environment.

5. Conclusion

In this study, poly(β -amino ester)-co-MMA networks were polymerized to create semi-degradable tough networks. The effect of the diacrylate to amine ratio on the modulus is limited to when the material is in its rubbery state. MMA concentration has a dual effect of increasing Tg and decreasing crosslinking density, which toughens the networks by several orders of magnitude. Under simulated physiological conditions, the mechanical properties change in a composition dependent manner, where select networks increase in toughness or sustain toughness while undergoing degradation. These networks establish a platform for designing semi-degradable networks with improved mechanical properties during degradation, which would allow for therapeutic release from tough implants.

Acknowledgments

The authors thank Dr. Kathryn Smith and Dr. Christopher Yakacki for discussions.

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Figure 1.

The glass transition temperature as determined by the tan delta from DMA. Biodegradable crosslinkers formed from four varying HDDA:3MOPA ratios were mixed with varying amounts of MMA.

Figure 2.

(A) The storage modulus at 37°C, (B) The storage modulus at a temperature of $T_g + 75^\circ\text{C}$, and (C) The crosslinking density at a temperature of $T_g + 75^\circ\text{C}$ as a function of MMA concentration. Networks composed from one of four HDDDA:3MOPA crosslinkers at varying diacrylate to amine ratio were mixed with MMA.

Figure 3.
Stress-strain curves of HDDA 1.15 networks at varying wt% MMA.

Figure 4. (A) Toughness as a function of elastic modulus for four different crosslinkers of varying diacrylate to amine ratios and varying wt% MMA. (B) Toughness as a function of temperature relative to the glass transition temperature where the testing temperature was 37°C.

Figure 5. (A) Degradation profile and (B) the water content profile over 8 weeks for 5 networks of varying %MMA.

Figure 6.
The glass transition temperature as determined by DSC for the 5 networks over 8 weeks of degradation in PBS.

Figure 7.
(A-E) Stress-strain curves of HDDA 1.15-co-MMA networks over 8 weeks. The wt% MMA varied from 0 to 75%.

Figure 8. (A) Failure strain profile and (B) the Toughness profile over 8 weeks for HDDDA-co-MMA networks.

Figure 9. (A) Elastic modulus profile and (B) Crosslinking density profile over 8 weeks for 0% MMA and 35% MMA networks.