Porous polymeric elastomers for repair and replacement of the knee joint meniscus
Groot, Jacqueline Hermina de

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Chapter 5

USE OF 50/50 COPOLY(L-LACTIDE/ $\delta$-CAPROLACTONE) FOR MENISCAL RECONSTRUCTION. IN-VITRO/IN-VIVO DEGRADATION OF COPOLYMER

Summary

Porous materials of a high molecular weight 50/50 copolymer of L-lactide and \( - \)caprolactone were used for meniscal reconstruction. Two series of porous materials with compression moduli of respectively 40 and 100 kPa were implanted in the knees of dogs. A porous aliphatic polyurethane series with compression modulus of 150 kPa was implanted as comparison. Fibrocartilage formation was found to be affected by the compression modulus of the implant. The implants with a modulus of 40 kPa did not show any fibrocartilage, those with compression moduli of 100 and 150 kPa yielded 50-70 % and 80-100 % fibrocartilage respectively. Copolymer implants showed better adhesion to meniscal tissue and therefore better healing of the lesion, probably to the high degradation rate of the copolymer. Carboxylic groups formed upon degradation contribute to the adhesion between implant and meniscal tissue. The copolymer degraded through bulk degradation because cracks are easily formed. Although the L-lactide sequences crystallized during degradation, crystalline remnants of the copolymer did not appear to cause problems in the latest stage of degradation.

Introduction

Since it is beyond doubt that menisci are important structures in the knee joint\(^1\)-\(^3\), there is an increasing interest in methods which can preserve meniscal tissue to prevent degenerative changes of articular cartilage. Repair of meniscal lesions by simple suturing is limited to the vascularized outer 10-20% part of the meniscus\(^4\),\(^5\). For meniscal lesions situated in the avascular central part of the meniscus, new experimental techniques are in the process of being developed. However, reports on repair with meniscus-like tissue, fibrocartilage, are sparse but it has been observed after application of a fibrin cloth in dogs\(^6\). With this technique, repair of only small lesions is possible. Previously we have shown that healing of large lesions in the avascular part, occupying about 30% of the length of the meniscus can be accomplished by implanting a porous polyesterurethane (PU) in a connecting defect\(^7\)-\(^9\). Porous polymer implants act as temporary scaffold for the formation of blood vessels and tissue. The repair tissue appeared to be fibrocartilage, which was based on morphological and immunological grounds\(^10\)-\(^11\).

Polyurethanes are appropriate materials for this application due to the excellent physical and mechanical properties and relatively good blood compatibility. They owe their properties to a domain structure which is achieved by the phase separation of the hard and soft segments.
In the first studies a polyester-urethane, which contains 4,4'-diphenylmethane diisocyanate (MDI) in the hard segments, was used. This diisocyanate may be converted to the toxic and carcinogenic methylenedianiline after degradation\textsuperscript{12,13}. In a previous study, the aromatic PU was replaced by an aliphatic PU, containing trans-cyclohexane diisocyanate, because aliphatic diamines are known to be less toxic than aromatic diamines\textsuperscript{14}. They are, however, still more or less toxic, except putriscine. Therefore a polymer that possess good mechanical properties and releases only non-toxic degradation products is preferred.

A polymer that fulfils this requirement is a high molecular weight 50/50 copolymer of L-lactide and \(\beta\)-caprolactone. This copolymer appeared to be an elastomer with mechanical properties comparable to segmented polyurethanes due to a highly entangled polymer chains and the presence of crystallizable L-lactide sequences\textsuperscript{15}. Upon degradation the polymer will yield L-lactic acid and \(\beta\)-hydroxy hexanoic acid as the degradation products. This polymer has successfully been implanted as nerve guide, was found to be non-toxic and showed a minor foreign body reaction\textsuperscript{16,17}.

To examine the possibilities of using this copolymer for meniscal reconstruction two series porous materials with respectively compression modulus of 40 and 100 kPa were implanted. An aliphatic PU series, which was used in a previously study, with a compression modulus of 150 kPa was implanted as comparison. Additionally, the degradation behaviour of the copolymer will be discussed.

**Experimental**

**Materials**

L-lactide (Purac Biochem. Gorinchem, The Netherlands) was recrystallized from dry toluene under nitrogen atmosphere. \(\beta\)-caprolactone (Jansen Chimica, Belgium) was dried with CaH\textsubscript{2} and distilled under reduced nitrogen pressure prior to use. The catalyst, stannous octoate (Sigma Corp., USA) was used directly from the supplier without further purification.

**Polymerization**

Polymerization was carried out in silanized glass ampoules. L-lactide and \(\beta\)-caprolactone were mixed to a 1:1 mole ratio and 1x10\textsuperscript{-5} mole catalyst per mole monomer was added. Under
vacuum the ampoule was heat sealed. The polymerization mixture was homogenized at the polymerization temperature. Polymerization was continued for 10 days at 110°C.

**Polymer processing**

**Implants:** Polymer was purified and freed from unreacted monomer by precipitating chloroform solutions into a 40/60 mixture of acetone and hexane. Porous materials were prepared using a freeze-drying/salt-leaching technique. Series I: a 5 wt.-% copolymer solution in 1,4-dioxane and c-hexane (90/10) was mixed with 30 wt.-% saccharose crystals (200-400 μm). After freezing the mixture at -15 °C, the solvent was sublimated under reduced pressure and the crystals were leached out with water. Series II: after mixing a 5 wt.-% copolymer solution in 1,4-dioxane and c-hexane (90/10) with 30 wt.-% saccharose crystals (200-400 μm), the polymer concentration was increased to 8% by slowly controlled evaporating the solvent under reduced pressure at 60°C. Then the mixture was frozen at -15°C and the solvent was removed at reduced pressure by sublimating. Afterwards, the crystals were leached out.

*Figure 1.* A 1-2 Old technique. A wedge-shape full thickness defect is made in the lateral meniscus to connect the experimental longitudinal lesion to the vascular periphery. B 1-2 New technique. A rectangular partial-thickness defect is made, leaving the lower layer of the meniscus intact.
with water. Series III: the porous aliphatic PU materials were prepared as described elsewhere. Implants were disinfected with 70/30 vol.% ethanol/water.

Films: Copolymer, PLLA and P(-CL) films were casted from dioxane solutions (3 wt.-%) resulting in a thickness of 0.15 mm. Additionally, copolymer films were obtained by compression moulding bulk polymerized polymer samples at 170°C followed by quickly cooling in a cold press (thickness 1.5 mm).

Surgery
Experiments were performed as described elsewhere on 26 lateral menisci of 13 dogs. A new surgical technique was used as shown in figure 1. Full thickness lesions were made which had a longitudinal extension both into the anterior and posterior side of the meniscus. A rectangular partial-thickness defect is made, leaving the inferior meniscal part intact (figure 1B/1,2). In this defect an implant was sutured using dexon sutures. Three series implants were implanted: series I in 8 knees, series II in 6 knees and series III in 12 knees.

Degradation
In-vivo: Porous copolymer (series I) samples (40x6x6 mm\textsuperscript{3}) and polymer films were subjected to degradation at 37 ± 1 °C in phosphate buffer, pH=6.9.

In-vivo: Porous copolymer (series I) samples (15x15x5 mm\textsuperscript{3}) and solid as polymerized copolymer samples (half disks: diameter 15 mm, thickness 4 mm) were subcutaneously implanted on four sides of the back of 20 Wistar albino rats (figure 2). The samples were removed after sacrifice of the rats. Follow-up periods were 1, 8, 16, 32 and 56 weeks. Porous samples for histological study were fixed in glutaraldehyde and other samples were washed with water and dried.

Characterization
The intrinsic viscosities were measured in chloroform at 25°C with an Ubbelohde viscosimeter. Size exclusion chromatography of polymer samples was carried out at 30°C on a Waters GPC 150 with chloroform as eluens. An estimate of the molecular weight was obtained by using the Mark-Houwink constants of polystyrene (K=4.9x10\textsuperscript{-3} dl/g and a=0.790).
Figure 2. Two porous copolymer samples (series I) and two solid as-polymerized copolymer samples were implanted subcutaneous on the back of rats.

Calorimeter studies were carried out with a Perkin Elmer DSC 7 calorimeter. The scanning rate was 10°C per minute in a range of -100°C to 260°C.

Changes of polymer composition were determined by 300 \(^1\)H NMR (Varian VXR-300) solutions in deuterated chloroform. In-vivo porous polymer samples were extracted with deuterated chloroform and filtered before measurement.

FT-IR internal reflection spectra were collected using a Bruker IFS88 FTIR spectrophotometer with a MCT-A detector at an incident angle of 45°. Casted films (thickness 0.15 mm) were pressed against KRS-5 reflection plate.

Compression curves were determined at room temperature using an Instron (4301) tensile tester equipped with a 100N load-cell at a cross-head speed of 12 mm/min. Cylindrical specimens with a diameter of 10 mm and a length of about 8 mm were cut out of the foams by cooling them with liquid nitrogen.

An ISI-DS-130 scanning electron microscope was used for studying the pore structure of the porous materials and the surface of the films.

For light microscopy, meniscal reconstruction implants and rat implants were fixed in formaldehyde and embedded in glycol methacrylate. Sections (2 \(\mu\)m) were stained with toluidine blue and Giesma.

Mass loss of in-vivo porous polymer samples as a function of implantation time was determined upon stained slides using the Quantimet 520 Image Analysis System. The percentage polymer was measured using a magnification of two. The average value was assessed out of seven measurements. The percentage polymer before implantation was set on 100%. Mass loss of in-vivo solid and in-vitro porous polymer was determined through weighing.
Results and discussion

L-lactide and \( \beta \)-caprolactone (figure 3) can be polymerized to high molecular weight with intrinsic viscosities \([\eta]\) up to 9.9 dl/g\(^{15}\). The average sequence length of L-lactide appeared to be 8.5. The presence of these relatively long crystallizable L-lactide sequences are responsible for the excellent mechanical properties of the material. The material is able to crystallize under strain and the tensile strength is 34 MPa, which is comparable to tensile strength of the aliphatic PU used in a previous study. The combination of good mechanical properties and the fact that upon degradation only non-toxic products are released, make this polymer suitable for the use as a strong, degradable, biomedical elastomer.

\[\text{Figure 3. L-lactide and } \beta\text{-caprolactone.}\]

Implant preparation

Implants were prepared using a freeze-drying/salt-leaching technique\(^8\). A polymer solution was mixed with saccharose crystals (200-400 \( \mu \)m) and frozen. After removing the solvent by sublimating, the crystals were leached out with water. The porous structure contained channel-like micropores (<50 \( \mu \)m) as a result of freeze-drying the solvent\(^{18}\) and macropores as a result of leaching out the crystals. These macropores have proven to be very important for the formation of fibrocartilaginous tissue\(^7,19\). Due to high molecular weight of the polymer, the maximal concentration of the polymer solution to which saccharose crystals could be added homogeneously was only 5%. The modulus of these materials was 40 kPa and is much lower than the modulus of implants described before (250 and 150 kPa). To increase the modulus,

\[\text{Table 1.}\]

<table>
<thead>
<tr>
<th></th>
<th>Series 1</th>
<th>Series 2</th>
<th>Series 3</th>
</tr>
</thead>
<tbody>
<tr>
<td>Polymer</td>
<td>copolymer</td>
<td>copolymer</td>
<td>aliphatic PU</td>
</tr>
<tr>
<td>Porosity</td>
<td>96 %</td>
<td>90 %</td>
<td>86 %</td>
</tr>
<tr>
<td>Microporosity</td>
<td>75 % &lt;50( \mu )m</td>
<td>62 % &lt;50( \mu )m</td>
<td>34 % &lt;30( \mu )m</td>
</tr>
<tr>
<td>Macroporosity</td>
<td>25% 200-400( \mu )m</td>
<td>38% 200-400( \mu )m</td>
<td>23% 50-90( \mu )m</td>
</tr>
<tr>
<td>Compression modulus</td>
<td>40 kPa</td>
<td>100 kPa</td>
<td>150 kPa</td>
</tr>
</tbody>
</table>
the polymer concentration had to be increased since the modulus $E$ and density $\rho$ are related through $^{20}$.

Figure 4. Scanning electron micrographs of a: copolymer implant series I, b: copolymer implant series II and c: PU implant series III.
The polymer concentration could be increased to 8%, after mixing the solution with crystals, by slowly controlled evaporating of the solvent. The maximal compression modulus of porous copolymer using this technique was 100 kPa.

Two copolymer series were implanted into knees of dogs. Series I and II implants had respective compression moduli of 40 kPa and 100 kPa and a respective porosity of 96% and 90%. As comparison a previously used series aliphatic PU implants (series III), with compression modulus of 150 kPa and porosity of 86%, was implanted (table 1). Figure 4 shows the scanning electron micrographs of the porous structures of the three series.

The materials were implanted by a new suturing technique. Using the old technique shown in figure 1 A1-2, lack of adhesion on one side of the implant and therefore lack of healing of the lesion was often observed. Therefore instead of creating a full-thickness defect to connect the lesion to the periphery, a partial thickness defect was made, leaving the inferior meniscal part intact (figure 1 B1-2). Thereby, less movement between implant and meniscus is possible and thus tearing of the connection and gap formation should be prevented.

Series I showed fastest ingrowth of fibrous tissue probably due to the high porosity. After 12 weeks ingrowth was complete. Ingrowth of fibrous tissue of series II and series III were comparable and was complete after 20 weeks. The tissue response was equal for all three series.21

Although series I showed the fastest ingrowth of fibrous tissue, fibrocartilage formation in the implant was never observed. In series II and series III maximal 50 - 70 % and 80 - 100 % fibrocartilage was observed after 12 weeks, respectively. It appears that the formation of fibrocartilaginous tissue is affected by the modulus and density of the implant.22

It is known that fibrocartilage is formed after metaplasia of fibrous tissue and apparently this differentiating takes does not take place in implants with high porosity and low compression modulus. There are several factors affecting the differentiation. In-vitro formation of chondrocytes, cell that produce cartilage, was determined by chick limb mesenchymal cells.23 These are cells that can differentiate to chondrocytes under the right circumstances. The differentiation was controlled by initial plating density. At high density most cells became chondrocytes, at intermediate density only a few chondrocytes developed and at low densities no development could be observed. The differentiation was affected by the oxygen level and chemical factors. It was also shown that exposed to intermitted compressive forces, high-density chondrocytes cultures showed an increase in cartilage
production\textsuperscript{24}. Additionally, chondrocytes can be grown in a three-dimensional matrix of an agarose\textsuperscript{25}, collagen gel\textsuperscript{26,27} or porous polymer scaffold\textsuperscript{28} but when they are grown in a monolayer they differentiate in cells with a fibroblast-like appearance. It may be possible that an implant stimulate chondrogenesis from precursor cells by local alteration of cell density, by offering the cells a three-dimensional matrix to grow and by adding compressive forces to the cells. In addition, when the implant is left out, the repair tissue appeared to be fibrous tissue\textsuperscript{22}. Probably, due to the very high porosity of the implant series 1, this situation is approached.

Figure 5 show the fibrocartilage formation in a copolymer implant (series II) and PU implant (series III) after 20 weeks. Noteworthy is the very small amount of polymer left in the case of copolymer implant, while in the case of PU implant the initial porous structure is still visible. The degradation rate of the copolymer apparently is much higher.

Gap formation between implant sides and meniscal tissue was prevented with this new suturing technique but the adherence of the implant to the underlying meniscal tissue has found to be a problem. The copolymer implants provides better adhesion (series I and II) than PU implants showing complete or partial adhesion in 7/8, 5/6 and 7/12 of the cases respectively. The explanation for this phenomenon is probably the higher degradation rate of
**Figure 6.** Mass loss as function degradation time. o in-vivo porous, ? in-vivo solid and ? in-vitro porous

**Figure 7.** Porous copolymer after a: 6 weeks, b: 18 weeks and c: 24 weeks implantation. The light coloured areas represent polymer.
Figure 8. Fragmentation of porous copolymer after in-vitro degradation. a: 8 weeks, b: 26 weeks, c: 47 weeks and d: detail of a fragment after 47 weeks.
the copolymer. Upon degradation carboxylic groups are formed which are able to stick to the meniscal tissue.

Healing of the lesion using series I and III implants were inventoried. Series I provides better results than series III implants showing complete or partial healing in 75% and 25% of the cases, respectively. Adherence of implant to meniscal tissue appeared to be essential for healing of the lesions.

Degradation

The degradation behaviour of the implant material is important for several reasons. The material act as a temporary scaffold for the healing of the lesions and for the ingrowth of fibrocartilage. Fibrocartilage formation is affected by the compression modulus and density of the implant and the degradation rate determine the maintenance of these factors during implantation. The late tissue response of the polymer can be of great consequence. For instance crystalline fragments of PLLA caused tissue response 3 years after implantation.\textsuperscript{29,30} Finally, the degradation rate appeared to affect the adhesion between implants and meniscal tissue, which is essential for healing of the lesion. Therefore we decided to study the degradation behaviour of the copolymer.

\textit{In-vitro:} Porous copolymer (series I) samples and polymer films were subjected to degradation at 37 \pm 1 \degree C in phosphate buffer of pH=6.9. \textit{In-vivo:} Porous copolymer (series I) samples and solid as-polymerized copolymer samples were subcutaneously implanted on four sides of the back of rats as shown in figure 2. Changes in intrinsic viscosity, mass, polymer composition and crystallinity were monitored as a function of degradation time. The surface of copolymer films were characterized with SEM and IR during degradation.

In figure 6 the remaining mass is plotted as function of the degradation time. Difference in mass loss can be observed between in-vivo and in-vitro. It appears that the biological surrounding influences the degradation rate. After half a year, the in-vivo porous material showed a mass loss of 75%. This percentage is much higher than in case of aliphatic PU degradation which in a previously study showed a mass loss of only 30% after one year\textsuperscript{14}. Figure 7 shows transmission electron micrographs of in-vivo porous coupes at different stages of degradation. Mass loss of the polymer (light coloured) is apparent. In-vitro, this porous material showed fragmentation upon degradation as presented in figure 8. Notable is
**Figure 9.** Intrinsic viscosity of copolymer as function of degradation time: \(\bullet\) in-vivo solid and \(\triangle\) in-vitro porous.

**Figure 10.** SEC chromatogram of copolymer before degradation (——) and after 16 weeks degradation (—).
the presence of cracks on the fragments. Fragmentation is also observed for the solid material after in-vivo degradation. Before implantation the material was transparent. After implantation material appearance became opaque until the material fragmented after 56 weeks in white crystalline-like fragments.

In figure 9 the intrinsic viscosity of in-vitro porous and in-vivo solid materials as a function of degradation time is shown. The molecular weight decreased fast with degradation time and the mass loss lagged behind the loss in molecular weight which is an indication of bulk erosion rather than surface erosion. This is also confirmed by the single peak that is observed in SEC experiments for all three series. This single peak shifts towards lower molecular weight during degradation as shown in figure 10. In case of surface erosion the molecular weight of the bulk is expected not change during degradation.

In figure 11-15 the variation of respectively Tg, heat of fusion, Tm and polymer composition as a function of degradation time is shown. Due to ingrown tissue in the porous in-vivo material, the changes of the morphology of the polymer could not be determined for this case. Before degradation the materials showed a single Tg peak at -14°C between the values of P(?-CL) of -60°C and PLLA 57°C, indicating that the lactide and caprolactone sequences are mixed to a certain degree. During degradation the Tg decreased and an increasing melting endotherm appeared between 100°C and 120°C. As the fusion temperature of the poly(?-caprolactone) is 64°C, this melting temperature corresponds to crystallized L-lactide sequences. The decreasing Tg is an indication that the amorphous phase becomes richer to ?-caprolactone. Since the amorphous phase is more susceptible to hydrolysis the percentage L-lactide increases with degradation time. Notable is the difference between the definite amount of L-lactide in the case of porous in-vivo and solid in-vivo which is 68% and 85%, respectively. Furthermore, narrowing of the melting endotherm at large degradation time is observed, indicating that L-lactide crystallites remain with narrow size distribution. Upon degradation the crystallinity increased to 43 J/g.

Crystalline debris in the latest stage of implantation can be of great consequence. In the case of as-polymerized PLLA bone plates and screws used for fixation of fractures, the remnants had a lamellar appearance with minimal thickness of 22 nm. The heat of fusion was 96 J/g and the melting point 184°C. They caused swelling three years after implantation. In the case of this copolymer implanted as nerve guides, however, the tissue reaction to the polymer remnants after eighteen months was very mild. After 2 years the appearance of the
Figure 11. $T_g$ as function of degradation time. ▼ in-vivo solid and ◦ in-vitro porous.

Figure 12. $T_m$ of in-vitro porous material as function of degradation time. ▼ represents the top of the peak; the lower ◦ represent the onset of the peak and the upper ▼ represent the end of the peak.
**Figure 13.** Tm of in-vivo solid as function of degradation time. ? represents the top of the peak; ; the lower ? represent the onset of the peak and the upper ? represent the end of the peak.

**Figure 14.** Heat of fusion as function of degradation time. ? in-vivo solid and ? in-vitro porous.
Figure 15. Percentage L-lactide as function of degradation time. ○ in-vivo porous, □ in-vivo solid and ▲ in-vitro porous.

Figure 16. FTIR/ATR spectra of a: copolymer before degradation, b: copolymer after 4 weeks hydrolytic degradation, c: difference spectrum, d: poly(L-lactide), e: poly(?-caprolactone) at an incident angle of 45°C.
remnants became more rounded, probably due to swelling\textsuperscript{17}. Swelling is caused by osmotic pressure due the carboxylic groups that are formed. Apparently the lower crystallinity and smaller crystallites size of the copolymer makes them more susceptible to hydrolytic degradation. Since the percentage polymer is very low after 2 years implantation, swelling of the copolymer remnants is not likely to cause problems.

\textit{Surface degradation.} FT-IR internal reflection technique was used to provide information about the change of surface composition of casted copolymer films during degradation. The surface depth that the IR beam penetrated was about 2 \( \mu \text{m} \). Figure 16 compares the spectra of copolymer film before degradation and 4 weeks after hydrolytic degradation. Significant differences in the carbonyl stretching band can be observed before and after degradation. Before degradation two C = O stretching peaks can be observed. One at 1758 cm\(^{-1}\) corresponding to the PLLA carbonyl and a smaller one at 1737 cm\(^{-1}\) corresponding to the poly(\(-\text{caprolactone}\) carbonyl. After degradation the relative intensity of the \(-\text{caprolactone}\) carbonyl has increased, indicating that percentage \(-\text{caprolactone}\) at the surface has increased. The difference spectrum between copolymer before and after degradation strongly resembles the spectrum of poly(\(-\text{caprolactone}\)). Thus, the amorphous phase rich of poly(\(-\text{caprolactone}\)) tends to migrate to the surface. Similar phenomenon is observed for PU’s. The flexible soft segments are more abundant at the surface and the crystalline hard segments to the bulk\textsuperscript{32,33}. Since the amorphous phase is susceptible to hydrolyse it is expected that carboxylic groups can be formed easily at the surface of the copolymer and contribute to the adhesion between implant and meniscal tissue.

Figure 17 show scanning electron micrographs of the surface of a casted film after 4 weeks hydrolytic degradation. Many cracks are visible with widths ranging from 14 nm to 250 nm. Cracks are formed due to residual stresses in the material. In these regions the polymer chains are stressed and susceptible to hydrolytic degradation.

Figure 18 shows a scanning electron micrograph of the surface of a compression moulded film 4 weeks after hydrolytic degradation. The crack density is much smaller and the width of the cracks is 14 nm. Apparently during compression moulding less residual stresses are left than after solvent evaporation. The surface of both films contain spherical structures of about 30 nm but is more pronounced for compression moulded film. The sphericals seem to be a result of phase separation. (cause production process or etching due to hydrolytic degradation surface determination \( T_c \)). Figure 19 shows the surface of the compression moulded film after
4 months of hydrolytic degradation. The cracks grew very large to widths of 5-20 \( \text{?m} \) and lengths of 10-300 \( \text{?m} \). The crack morphology of the compression moulded film is very different from the casted film.

The presence of crack enables the water to reach the inner parts of material and it is therefore obvious that these materials degrade through bulk degradation rather than surface erosion.

**Conclusions**

Porous copolymer can be used for meniscal reconstruction. Gap formation between implant sides and meniscal tissue could be prevented with a new suturing technique. Fibrocartilage formation was affected by compression modulus of the implant. Implants with a compression modulus of 40 kPa never showed ingrowth of fibrocartilage while the implants with compression moduli of 100 kPa showed 50-70% fibrocartilage after 12 weeks. The largest percentage fibrocartilage of 80-100% was observed in PU implants with compression moduli of 150 kPa. Although the ingrowth of fibrocartilage is very important, it is not necessary for the healing of the lesions. Both copolymer implants induce better healing of the lesions than
PU implants. They also showed better adherence to the underlying meniscal tissue. This adherence was found to be essential for the healing of the lesions. The reason for this better adhesion is probably the higher degradation rate of the copolymer yielding a increased number of carboxyl groups.

The polymer degrades through bulk degradation rather than through surface erosion because the material easily forms cracks in the early stage of degradation. During degradation the material phase separated in an amorphous phase containing mainly caprolactone and a crystalline phase containing L-lactide. The amorphous phase tends to migrate to the surface. As the amorphous phase is susceptible to hydrolysis, carboxylic groups easily can be formed at the surface of the copolymer and contribute to the adhesion between implant and meniscal tissue. Although the material crystallizes upon degradation due to the presence of long L-lactide sequences, crystalline remnants of the copolymer are not expected to cause problems in the latest stage of degradation.

The compression modulus of 100 kPa of the copolymer implants has proven to be too low to accomplish 100% fibrocartilage ingrowth. Additionally, the degradation rate of the polymer is high and the compression modulus is expected to decrease during degradation. Therefore
material with higher compression module are necessary. However, using the freeze-drying/salt-leaching technique, only materials with a maximal compression modulus of 100 kPa can be prepared due to the high molecular weight of the copolymer. An other technique has to be developed to increase the compression modulus of the materials.

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