Biodegradation of polyurethane under fatigue loading

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Abstract: A method utilizing expansion of a diaphragm-type film specimen was developed to study in vitro biodegradation of poly(etherurethane urea) (PEUU) under conditions of dynamic loading (fatigue). A finite element model was used to describe the strain state, which ranged from uniaxial at the edges of the film to balanced biaxial tensile strain at the center. During testing, the film was exposed to a H2O2/CoCl2 solution, which simulated in vivo oxidative biodegradation of PEUU. The extent of chemical degradation was determined by infrared analysis. Physical damage of the film surface was characterized by optical microscopy and scanning electron microscopy. Dynamic loading did not affect the rate of degradation relative to unstressed and constant stress (creep) controls in regions of the film that experienced primarily uniaxial fatigue; however, degradation was accelerated in regions that experienced balanced biaxial or almost balanced biaxial fatigue. It was concluded that the combination of dynamic loading and biaxial tensile strain accelerated oxidative degradation in this system. Chemical degradation produced a brittle surface layer that was marked by numerous pits and dimples. Physical damage of the surface in the form of cracking occurred only in fatigue experiments. Cracking was not observed in unstressed or creep tests. Cracking initiated at the dimples produced by chemical degradation, and propagated in a direction that was determined by the strain state. © 2003 Wiley Periodicals, Inc. J Biomed Mater Res 65A: 524-535, 2003

Key words: polyurethane; biodegradation; fatigue; creep

INTRODUCTION

Polyether-based polyurethane (PEU) has been utilized in a number of implantable biomedical devices.1 One of the first biomedical applications of PEU utilized the material as the outer insulating sheath surrounding cardiac pacing leads.2 From early experience with pacing leads, PEU was recognized as biocompatible and was thought to be biostable. It was not until nearly a decade after PEU was first used in pacing leads that it was recognized that a small percentage of leads were failing.3 Lead failure occurred when physical damage to the PEU, in the form of cracks and tears, compromised its ability to act as an insulator. Analyses of failed leads led to the understanding that physical damage could occur in response to residual stress or strain. This type of physical damage was labeled as environmental stress cracking (ESC).4 Annealing the PEU to reduce residual stresses, e.g., hoop stresses due to interference fits between the outer PEU insulating sheath and the conductor coil, and abandoning lead designs that caused large bending strains on the PEU, e.g., in the J-type leads, partially resolved the problem of ESC.3,5,6 It was also recognized that chemical degradation of the PEU macromolecule was involved in lead failure and that biostability was likely a function of interrelated phenomena of physical damage and chemical degradation.7 It is now understood that the mechanism of chemical degradation involves oxidative attack on the polyether soft segment, starting with abstraction of a proton from the α-methylene position and leading to chain scission and/or chemical crosslinking.8,9

Based on early experience with pacing leads, a number of in vivo and in vitro experiments were devised to further understand the influence of mechanical stress on chemical and physical degradation. It was shown that constant uniaxial stress (creep mode) affected the in vitro biostability of a poly(etherurethane urea) (PEUU) in two ways.10,11 It inhibited chemical degradation by causing stress-induced crystallization of the ether soft segment. On the other hand, stress enhanced physical damage. Stress concentration at surface
flaws, created by localized chain scission, caused the flaws to enlarge first into small pits and then into larger voids as the pits coalesced. Similar results were observed for in vitro degradation of the same PEUU under constant uniaxial strain (stress relaxation mode). Increasing uniaxial strain caused progressively more orientation of soft segment chains, which corresponded to a progressive decrease in the amount of chemical degradation. Physical damage was enhanced by constant uniaxial strain, but only at very high strains.

In addition to uniaxial strain, the affect of constant biaxial strain on biostability has also been tested. Various levels of unbalanced biaxial strain were tested. Biaxiality inhibited orientation of the ether soft segment. Orientation of the soft segment in the direction of highest strain did occur, but only at higher strains, depending upon the ratio of the level of strain in each direction. As would be expected, chemical degradation was inhibited when orientation did occur. When no orientation was detected, biaxial strain had no effect on chemical degradation. Chemical damage, particularly at high levels of biaxial strain, enhanced physical damage. Biostability in these examples can be explained by competing factors: orientation inhibited chemical degradation whereas the applied load enhanced physical damage.

Although understanding the effects of constant stress or strain on biostability helps considerably in explaining the phenomenon of ESC in pacing lead insulation, little research has been devoted to the effects of dynamic loading on biostability. The potential importance of dynamic loading can be appreciated if one considers that pumping diaphragms, present in ventricular assist devices and total artificial hearts, require a flexible membrane that exhibits long-term biostability under dynamic stress. Potential problems with using PEU as pumping diaphragms could be anticipated by studying the effect of dynamic stress on PEU biostability. It is reasonable to anticipate a significant effect of dynamic stress on the working lifetime of a PEU pumping diaphragm, considering the demonstrated effects of constant stress and strain on biostability, and well-documented observations that dynamic stresses promote chain scission, crack initiation and propagation in other environmentally challenged synthetic elastomers.

Various researchers have investigated the effects of dynamic stresses on PEUs, but most of the studies did not directly address biostability. Furthermore, most fatigue experiments were performed in uniaxial tension, whereas the diaphragm operates under multiaxial stresses. In the present study, a method was developed to study the effect of dynamic mechanical stress (fatigue) on the biostability of PEUU. This method allowed fatigue experiments to be performed in an environment that has been shown to reproduce the oxidation of PEU observed in vivo. Moreover, the experimental design permitted uniaxial and balanced biaxial fatigue to be performed simultaneously. For comparison, constant stress experiments and experiments with no applied stress were also performed. During fatigue, the PEUU was exposed to a H2O2/CoCl2 environment that simulates in vitro degradation of PEUU. Chemical degradation and physical damage were evaluated as measures of biostability.

MATERIALS AND METHODS

Materials

The poly(etherurethane urea) (PEUU) was described previously. It had a poly(tetramethyleneglycol) soft segment (Mn = 2000) and a di(4-phenyl isocyanate) (MDI) hard segment chain extended with diamines. To prevent premature failure of the PEUU at the clamped edges of the fatigue specimen, the tested films were bilayers of PEUU and a polycarbonate polyurethane that was stable to oxidative degradation. The polycarbonate polyurethane (PCU) had a hexanediol carbonate soft segment (Mn = 1939) and an MDI hard segment chain extended with diamines. Both polyurethanes were in a solution of dimethyl acetamide, 20% solids by weight.

To fabricate the bilayer film, the PCU solution was cast onto a glass substrate to a uniform thickness using a doctor blade. The solvent was removed in vacuo at ambient temperature. The PEUU was then cast over the PCU layer with a doctor blade and dried in vacuo. The PCU base layer and the PEUU top layer were both approximately 105–110 μm thick, giving a total film thickness of 210–220 μm.

In vitro fatigue and creep experiments

Testing was carried out with a custom-designed film holder coupled to a commercial Flexercell Strain Unit (Flexcell International Corp., McKeesport, PA). The design of the film holder allowed deformation of a circular section of the film by application of a vacuum to the bottom of the film (PCU side), whereas the top of the film (PEUU side) was exposed to an in vitro testing solution. The film holder consisted of two blocks, an upper block constructed of poly(oxymethylene) (POM) and a lower block constructed of aluminum. A single continuous film was inserted between the blocks, which were secured with a series of screws. The blocks contained eight specimen wells. One well is shown in Figure 1. The space above the specimen was filled with the appropriate test solution, and the film was deformed by application of a vacuum from below. The magnitude and frequency of the applied vacuum were controlled with the Flexercell Strain Unit. During a typical test, four of the wells were subjected to loading, either fatigue or creep, and the remaining four wells were not loaded and served as controls. A typical experiment lasted for 12 days.
ate time points, 3, 6, and 9 days, a section of the film from a loaded well and from an unloaded well were removed for analysis.

To reproduce the mechanism of in vivo PEUU degradation at an accelerated rate, the wells were filled with a solution of 20% H2O2/0.1 M CoCl2. As controls, wells were filled with a solution of 20% H2O2 or with distilled water. All solutions were refreshed every 3 days, and all testing was done at 37°C.

Fatigue testing was carried out at a frequency of 1 Hz with a maximum load of 50 kPa. The stress as a function of time was measured during fatigue testing with a pressure transducer. The fatigue waveform as measured in this way is shown in Figure 2. The mean load applied during fatigue testing was ~25 kPa. Creep experiments were performed at a constant applied load of either ~50 or ~25 kPa.

Characterization of strain state

Deformation varied across the film from essentially uniaxial strain at the edges to balanced biaxial strain at the center. The strain was resolved into radial and tangential components, \( \varepsilon_r \) and \( \varepsilon_t \), respectively, as shown in Figure 1(b). The deflection at the center of the film \( (d) \) was measured from the vertical displacement of a pipette. A separate determination of the strain at the center of the film \( (\varepsilon) \) was obtained by depositing gold squares on a film and measuring the changes in dimension of the squares at maximum deflection with an optical microscope (OM). The pattern was deposited by placing a flat metallic mesh on the outer PEUU surface of the film (40 lines per inch) and sputter coating the film with 200 Å of gold in a Hummer 6.2 sputtering system (Anatech Ltd., Alexandria, VA). The mesh was removed, leaving gold squares on the film surface. It was only possible to measure the strain at the center of the film in this way because of curvature during deformation.

After direct measurement of \( d \) and \( \varepsilon \) a silicone cast of the deformed film was made for determining the strain distribution across the film. An RTV11 silicone base compound and dibutyl tin dilaurate curing agent was used according to the manufacturer’s recommendation to make the cast (GE Silicones, Waterford, NY). The silicone base compound and curing agent were mixed and poured into the well. After curing for 24 h, the film and the silicone cast were removed from the loading cell. The film remained adhered to the silicone cast upon removal. Because the film was clear, the gold squares were visible through the film. With the film adhered to the silicone cast, the dimensions of the gold squares were measured by optical microscopy keeping the line of sight perpendicular to the surface of the silicone cast to ensure accurate measurements. The radial and tangential strains as a function of distance from the center of the film were measured in this way. The deflection was also measured from the silicone cast. To obtain the strain in the thickness direction, the silicone cast was bisected through the diameter. The thickness of the film as a function of distance from the center was measured from micrographs recorded with a scanning electron microscope (SEM; JEOL 840-A, Peabody, MA). Before examination in the SEM the silicone cast and the adherent film were coated with 100 Å of gold.

The deflection was measured every 3 days during the experiment. Because significant creep occurred during the course of an experiment, a finite element model (FEM) was developed to predict the strains on the film surface based on the measured deflection. The FEM also served to confirm the strain distribution that was measured with the silicone cast. Finite element modeling was done with the software package ABAQUS/Standard, version 5.6 (Hibbitt, Karlsson & Sorensen, Inc., Pawtucket, RI). The film was modeled using a total of 200 two-node linear elastic axisymmetric elements. The model was defined to be 2.5 cm in diameter and 220 μm thick.
thick. Nodes around the boundary of the model were considered pinned. Loading was distributed over the bottom surface of the elements and was applied incrementally in four steps. The software provided the positional displacement and the tangential and radial strain at each node.

Surface characterization

Before and after treatment, each film was analyzed by attenuated total reflectance Fourier transform infrared microspectroscopy (FTIR). Data were collected with a Nexus 870 FT-IR bench coupled to a Continuum microscope (Thermo Nicolet, Madison, WI). The microscope utilized an attenuated total reflectance (ATR) slide-on attachment with a germanium crystal to allow micro-ATR analysis. Spectra were collected at a resolution of 2 cm⁻¹ for 32 scans. Spectra were collected every 1 mm along a diameter over the film surface. Each spectrum sampled an area 150 × 150 μm.

Chemical degradation was quantified as the percent ether remaining after treatment compared with the untreated film. Ether content was determined from the absorbance of the 1110 cm⁻¹ peak (asymmetric stretch of soft segment ether) relative to the internal reference peak at 1595 cm⁻¹ (stretching of the carbon-carbon double bond in the aromatic hard segment). The amount of crosslinking was quantified from the absorbance of the 1174 cm⁻¹ peak (stretch of soft segment branched ether) relative to the internal reference peak. Physical damage was characterized by OM and SEM. Representative SEM images were recorded at the center of treated films and at 4, 8, and 12 mm from the center. All SEM images are of surfaces tilted by 60° to show surface morphology.

RESULTS AND DISCUSSION

Finite element model

The finite element model (FEM) was tested by comparing the displacement at the center (d) and the strain at the center of film (ε = ε₀ = ε₁) at maximum loading to the FEM prediction. The measured values of d and ε were 6.22 mm and 17.2%, respectively, at a loading of ~50 kPa. Initially, the tensile modulus from uniaxial stress-strain measurements, E = 12.1 MPa, and a typical Poisson’s ratio for an elastomer, ν = 0.49, were used in the FEM. These values, however, predicted d and ε to be much larger than the measured values (Table I). The FEM accurately predicted the measured values of d and ε with E = 14.1 MPa and ν = 0.3 (Table I).

From the silicone cast, d was measured to be 6.61 mm and ε was measured to be 19.0%. These values were higher than those directly measured from the film. This was attributed to creep of the film during the 24-h period required for complete curing of the silicone resin. Accommodating the increase in d and ε required a decrease in E from 14.1 to 12.4 MPa in the FEM (Table I).

To further test the FEM, the strain distribution on the film measured from the gold grid on the silicone cast was compared with the strain distribution from the FEM. Despite scatter in the measured ε, and ε₁ data, which was attributed to imperfect alignment of the silicone cast surface in the optical microscope, the FEM with E = 14.1 MPa and ν = 0.3 accurately predicted the distribution in radial strain (ε₀), tangential strain (ε₁), and strain in the thickness direction (ε₂) as measured from the silicone cast (Fig. 3). Representative data is plotted in Figure 3. Radial and tangential strains were maximum and equal at the center (balanced biaxial strain state). Radial strain decreased slightly toward the edge, whereas tangential strain decreased to zero at the edge (uniaxial strain state). The tensile strains were accompanied by a decrease in the thickness of the film.

In formulating the FEM it was assumed that the strains were small enough that the material behaved as a linear elastic solid. This assumption was reasonable based on the fact that the strains involved were <20% and the uniaxial stress strain curve was approximately linear up to 50% strain. It was also assumed that the modulus measured in uniaxial tension accurately described the response to biaxial deformation. This assumption may not have been valid. Some elastomers are reportedly stiffer in balanced biaxial tension than in uniaxial tension.₂⁶ This could explain why the FEM required a higher modulus than the measured uniaxial tensile modulus to produce the measured values of d and ε.

An increase in volume during deformation was indicated by the strain measurements from the silicone cast and was reflected by the Poisson’s ratio of 0.3 in the FEM. The Poisson’s ratio of polyurethanes similar to the one in this study has been measured to be between 0.45 and 0.5.₂⁷ Although using a Poisson’s ratio <0.5 was counterintuitive, a value of 0.3 accurately predicted the strain distribution in the deformed PEUU film and therefore was used in subsequent calculations.

### Table I

<table>
<thead>
<tr>
<th>Method</th>
<th>d (mm)</th>
<th>ε (%)</th>
<th>E (MPa)</th>
<th>ν</th>
</tr>
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<tbody>
<tr>
<td>Direct measure</td>
<td>6.22 ± 0.08*</td>
<td>17.2</td>
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<td></td>
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<tr>
<td>FEM</td>
<td>7.45</td>
<td>24.9</td>
<td>12.1</td>
<td>0.49</td>
</tr>
<tr>
<td>FEM</td>
<td>6.22</td>
<td>17.4</td>
<td>14.1</td>
<td>0.3</td>
</tr>
<tr>
<td>Silicone cast</td>
<td>6.60 ± 0.06*</td>
<td>19.0</td>
<td></td>
<td></td>
</tr>
<tr>
<td>FEM</td>
<td>6.61</td>
<td>19.4</td>
<td>12.4</td>
<td>0.3</td>
</tr>
</tbody>
</table>

*Average ± SD; n ≥ 4.
Strain state during deformation

The time dependence of the maximum displacement was measured over the course of fatigue, and creep experiments were carried out in different environments. Figure 4 is representative data showing the change in the deflection ($\Delta d$), defined as the difference between the maximum deflection measured at some time point and the maximum deflection measured immediately after loading commenced. In Figure 4(a), $\Delta d$ is plotted for creep loading at $-50$ kPa for films tested in water, H$_2$O$_2$, or H$_2$O$_2$/CoCl$_2$. In all three treatments, an initial rapid increase in the deflection was observed between day 0 and day 3. After day 3, there appeared to be little or no further increase in the maximum deflection. Films tested in H$_2$O$_2$ and H$_2$O$_2$/CoCl$_2$ deflected an additional 2.3 mm over the initial displacement, whereas films tested in H$_2$O deflected only an additional 1.6 mm. It is known that swelling of PEUU by either H$_2$O or H$_2$O$_2$ increases the compliance due to plasticizing effects, and the effect is more pronounced with H$_2$O$_2$, which swells PEUU by 6.7% by weight, whereas H$_2$O swells PEUU by 1%. The increase in deflection between day 0 and day 3 was attributed primarily to the increase in compliance due to swelling. This behavior has been observed in previous studies of PEUU creep behavior. One would expect $\Delta d$ to continue increasing after day 3 because of the viscoelastic nature of the PEUU. Indeed, this behavior was observed previously and was termed primary creep. However, the magnitude of the effect was small and would have been within experimental error of the $\Delta d$ measurement.

In addition to primary creep, secondary creep was previously observed in PEUU treated in H$_2$O$_2$/CoCl$_2$. Secondary creep was characterized by an increase in slope of the logarithmic creep response at longer time points. It arose from the effect of chemical degradation on the bulk compliance of the film. Secondary creep was not observed within the time scale of the present experiments. Although chemical degradation responsible for secondary creep certainly occurred in the present experiments, construction of the film with an oxidation-resistant PCU base layer and design of the experiment with only the PEUU side exposed to the in vitro solution would have minimized the effect of chemical degradation on bulk properties of the film.

The trend in $\Delta d$ observed with creep loading at $-50$ kPa was also observed in fatigue loading [Fig. 4(b)]. There was an initial rapid increase in the deflection between day 0 and day 3, with no further increase at longer times. Similarly, a larger deflection was observed after day 3 for H$_2$O$_2$ treatments than for H$_2$O treatment. Comparing the magnitude of $\Delta d$ between creep loading at $-50$ kPa and fatigue loading at $-50$ kPa in H$_2$O$_2$/CoCl$_2$ in Figure 4(c), it is seen that $\Delta d$ after day 3 was about twice as great in creep loading as in fatigue loading, 2.3 and 1.2 mm, respectively. The deflection during creep loading at $-25$ kPa was 1.2 mm, the same as in fatigue at $-50$ kPa. This meant that...
the mean load applied to the film controlled the increase in deflection between day 0 and day 3.

The FEM was used to obtain the strain distribution on the film after day 3 of fatigue and creep testing. The modulus used in the model was decreased to obtain a match with the measured deflection. An initial loading of −50 kPa produced an average displacement of 6.22 mm. Between day 3 and day 12 of fatigue in H$_2$O$_2$/CoCl$_2$, the deflection increased to an average value of 7.34 mm (Table II). The modulus in the FEM decreased from 14.1 to 10.1 MPa to match the new deflection and thereby accommodate the increase in effective compliance of the film. This corresponded to $\varepsilon = 23.3\%$. This was significantly higher than the value measured immediately after loading, $\varepsilon = 17.2\%$. The average deflection between day 3 and day 12 for creep loading at −50 kPa was 8.56 mm. Decreasing the modulus in the FEM to 7.7 MPa to match the deflection gave $\varepsilon = 30.3\%$. These results are summarized in Table II.

Subsequently, $\varepsilon_r$ and $\varepsilon_t$ were obtained from the FEM using the modulus values in Table II. In Figure 5(a), $\varepsilon_r$ and $\varepsilon_t$ are plotted against the distance from the center of the film for fatigue loading. The strains were maximum and equal at the center of the film. Toward the edge, $\varepsilon_t$ quickly decreased to zero, whereas $\varepsilon_r$ slowly decreased to $\sim 70\%$ of its maximum value. Region I was defined as the area where $\varepsilon_r$ and $\varepsilon_t$ were both greater than 85% of their maximum value. This region of primarily biaxial strain was located within 5 mm of the center of the film. Region II was the area between 5 mm from the center and the edge of the film where the strain state gradually changed from almost balanced biaxial to uniaxial. Although the magnitude of the strain in creep loading at −50 kPa was greater than in fatigue, the shapes of the $\varepsilon_r$ and $\varepsilon_t$ distributions were the same.

The ratio $\varepsilon_t/\varepsilon_r$ was a measure of strain biaxiality and is plotted against the distance from the center of the film in Figure 5(b) for fatigue loading after day 3. In region I the strain was nearly balanced biaxial with $\varepsilon_t/\varepsilon_r \geq 0.92$. In region II, the tangential strain rapidly decreased toward the edge, producing nearly uniaxial strain with $\varepsilon_t/\varepsilon_r = 0$.

The corresponding nominal radial and tangential

![Figure 4](image-url) Change in maximum deflection ($\Delta d$) over time for several experiments: (a) creep at −50 kPa, (b) fatigue, and (c) creep and fatigue.

<table>
<thead>
<tr>
<th>Method</th>
<th>$d$ (mm)</th>
<th>$\varepsilon$ (%)</th>
<th>Effective modulus, $E$ (MPa)</th>
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</thead>
<tbody>
<tr>
<td>Direct measure immediately after −50 kPa loading</td>
<td>6.22 ± 0.08$^a$</td>
<td>17.2 ± 0.09$^a$</td>
<td></td>
</tr>
<tr>
<td>Direct measure from fatigue loading and creep loading at −25 kPa</td>
<td>7.34 ± 0.05$^{a,b}$</td>
<td>23.3</td>
<td>10.1</td>
</tr>
<tr>
<td>FEM prediction</td>
<td>7.33</td>
<td></td>
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</tr>
<tr>
<td>Direct measure from creep loading at −50 kPa</td>
<td>8.56 ± 0.08$^{a,b}$</td>
<td>30.3</td>
<td>7.7</td>
</tr>
<tr>
<td>FEM prediction</td>
<td>8.55</td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

$^a$Average ± SD; $n \geq 4$.

$^b$Average deflection measured between day 3 and 12 of testing.
strain rates ($\varepsilon_\tau$ and $\varepsilon_t$, respectively) for fatigue loading are plotted as a function of distance from the center of the film in Figure 5(c). The radial and tangential strain rates were obtained by dividing the maximum strain by one-half of the loading period. Both $\varepsilon_\tau$ and $\varepsilon_t$ were highest at the center of the film; $\varepsilon_\tau$ remained within 85% of the maximum value in region I, whereas $\varepsilon_t$ decreased to zero at the edge.

**Chemical degradation**

The extent of chemical degradation was measured by the change in intensity of the normalized ether band at 1100 cm$^{-1}$. The intensity of the 1100 cm$^{-1}$ band for the unstressed control did not vary across the film, and the average intensity from several control experiments is indicated on each plot as a solid line. Consistent with previous observations, the intensity decreased progressively from 94–85 to 75–55% after 3, 6, 9, and 12 days exposure to the oxidative H$_2$O$_2$/CoCl$_2$ solution.

The relative intensity of the 1100 cm$^{-1}$ band as function of position on the film for creep loading at $\pm 50$ kPa at 3, 6, 9, and 12 days is plotted in Figure 6 for representative creep experiments. At all time points the change in ether intensity was constant across the film and showed approximately the same amount of chemical degradation as the unstressed control. Scatter in the data was always present, especially at later time points. Similar scatter was observed in the control data. This scatter was attributed to inhomogeneous degradation of the surface of the film. Regions of higher degradation associated with formation of small pits or dimples were typical of in vivo and in vitro degraded PEUU films. Creep loading at $\pm 25$ kPa produced the same results as creep loading at $\pm 50$ kPa in that the ether loss was constant across the film and was comparable in magnitude to that of the unstressed control.

Under the creep loading conditions used in the present study, neither uniaxial creep near the edge nor balanced biaxial creep at the center of the film affected the rate of biodegradation. Although previous studies have reported that uniaxial creep inhibits in vitro oxidation of PEUU due to strain-induced crystallization of the soft segments, the strains used were much higher, between 150 and 750%. In the present creep experiments, the edge of the film experienced an initial uniaxial strain of only $\sim 10\%$. This was much lower than the 150% required for the onset of strain-induced crystallization. Soft segment orientation at strains as low as $50\%$ was observed to inhibit degradation in uniaxial stress relaxation, but not in biaxial stress relaxation. If orientation occurred in the fatigue and creep experiments near the edge, it was not high enough to significantly influence degradation.

The ether intensity for a representative fatigue experiment is compared with the unstressed control at 3, 6, 9, and 12 days in Figure 7. After 3 days of fatigue, the ether intensity was constant across the film and showed approximately the same amount of ether loss as the unstressed control. After 6 days, the center of the film (region I as defined in Fig. 5) showed more ether loss than the rest of the film and showed more ether loss than the unstressed control. The edge of the fatigued film (region II in Fig. 5) showed about the same or even less ether than the unstressed control. The difference in ether intensity between the center and edge of the fatigued film became more distinct at 9 and 12 days.
Figure 8 compares the rate of ether loss at region II and at the center of the film in one fatigue experiment to the rate of ether loss of the unstressed control. In the control, ether loss was gradual and fell to about 56% of its initial value at day 12. In region II of the fatigued film, ether loss was similar to the control. Considering the experimental scatter, ether loss in the control and in region II of the fatigued film were not considered to be significantly different. In the center of the fatigue film, ether loss was significantly greater than in the control. The ether content fell to \(\sim 35\%\) of its initial value at day 12.

Figure 6. Percent ether remaining for unstressed controls (solid lines) and creep at \(-50\) kPa (data points) at (a) 3 days, (b) 6 days, (c) 9 days, and (d) 12 days.

Figure 7. Percent ether remaining for unstressed controls (solid lines) and fatigue tests (data points) at (a) 3 days, (b) 6 days, (c) 9 days, and (d) 12 days.
Previous studies demonstrated the effect of soft segment orientation and crystallization on the degradation rate of PEUU films mechanically deformed either under conditions of creep or stress relaxation. Apparently the imposed strains were too small for these effects to be observed in the present creep experiments. However, fatigue loading, especially biaxial fatigue, did accelerate degradation although the imposed strains were actually smaller than in creep. The difference between creep and fatigue results must be attributed, at least in part, to the effect of strain rate. The dependence of chemical degradation on strain rate can be tested easily by varying the frequency of the fatigue experiment. This is the subject of a future publication.

Physical damage

Surfaces of films exposed to constant stress (creep) or no stress (unstressed controls) were featureless when viewed in the optical microscope. Films fatigued up to 6 days were also featureless [Fig. 9(a)]. However, cracks appeared on the surface of fatigued films at later time points. Cracking was observed as early as day 9, although it usually did not appear until day 12 or even later. The example in Figure 9(b) shows a film that had started to crack after 9 days of fatigue.

Figure 8. Comparison of degradation rate for fatigue tests and unstressed controls, average ± SD, n ≥ 4.

Figure 9. Optical micrographs of films after (a) 3 days, (b) 9 days, (c) 12 days, or (d) 18 days of fatigue in H₂O₂/CoCl₂.
Initially, cracking was confined to a small region in the center of the film. After 12 to 15 days, cracks were more numerous, and the cracked area had spread outward from the center [Fig. 9(c)]. At later time points, 18–21 days, cracking appeared over the entire surface of the film [Fig. 9(d)]. The crack direction in the center tended to show no orientation, whereas cracks near the edge tended to show tangential orientation.

Examination in the SEM confirmed that the initially featureless surfaces remained featureless at the micron scale at day 3 [Fig. 10(a)]. However after 6 days all the surfaces exposed to H₂O₂/CoCl₂, including unstressed controls, creep and fatigue films, had developed small dimples that became more numerous at day 9 and day 12 [Fig. 10(b)].

The type of loading had no effect on the size or density of the dimples, nor was there any difference in dimpling between the center and edge of the film (Fig. 11). Dimpling preceded cracking in the fatigue experiments. The cracks appeared to initiate at the dimples, as indicated by the arrows in Figure 11(e,f). The direction of crack propagation correlated with the strain state of the fatigued film. Nearly balanced biaxial strain in the center (region I) caused the direction of crack propagation to be nonpreferential. Closer to the edge (region II) where the strain state approached uniaxial, cracks propagated in the direction perpendicular to the highest strain, i.e., tangentially.

Oxidative degradation of PEUU in vitro and in vivo results in soft segment crosslinking in addition to chain scission. The combined effects of crosslinking and chain scission produce a brittle surface layer on the PEUU film. The brittle layer was previously estimated to be on the order of 10 μm thick and cracked easily if the degraded film was bent or stretched. In the present study, crosslinking of all the films exposed to H₂O₂/CoCl₂, whether in creep, fatigue, or unstressed, was confirmed by the presence of a characteristic band at 1174 cm⁻¹ in the infrared spectrum. It is speculated that the brittle surface layer cracked under dynamic loading conditions encountered in fatigue. The importance of dynamic loading is indicated by the fact that no cracks were observed in creep experiments. Dimples facilitated crack initiation by acting as sites of stress concentration with high local strain rate.

CONCLUSIONS

A method utilizing expansion of a diaphragm-type film specimen was developed to study biodegradation of PEUU under conditions of dynamic loading (fatigue). Experiments were also carried out without loading and with a constant load (creep). The strain was not uniform over the film. A finite element model was used to describe the strain state, which ranged from uniaxial at the edges to balanced biaxial tensile strain at the center. In the range of strains tested, creep had no effect on the rate of in vitro chemical degradation of PEUU, compared with unstressed controls, regardless of the strain state. Dynamic loading did not affect the rate of degradation in regions of the film that experienced primarily uniaxial fatigue; however, degradation was accelerated in regions that experienced almost balanced biaxial fatigue. From these observations, it was apparent that accelerated oxidation resulted from the combination of dynamic loading and biaxial tensile strain in this system.

Chemical degradation produced a brittle surface layer that was marked by numerous pits and dimples. Physical damage of the surface in the form of cracking occurred only in fatigue experiments. Cracking was not observed in unstressed or creep experiments. Cracking occurred first in biaxially loaded areas of the
specimen that experienced accelerated chemical degradation. Cracks initiated at dimples produced by chemical degradation and propagated in a direction that was determined by the strain state. In biaxially loaded areas, there was no preferential direction of crack propagation. In primarily uniaxially loaded areas, the cracks propagated perpendicular to the direction of maximum tensile strain.

Figure 11. Representative scanning electron micrographs from three regions of films loaded in either creep at -50 kPa or fatigue, all after 12 days of treatment in H₂O₂/CoCl₂.

References


