Microfabrication and nanomechanical characterization of polymer microelectromechanical system for biological applications

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Polymer microelectromechanical system (MEMS) devices are promising for biological applications such as development of biosensors and biomechanical devices. In order to develop polymer biological MEMS (BioMEMS), polymer microfabrication techniques are required, and the nanomechanics studies, including measurement of the nanomechanical properties of the polymer materials, must be carried out. This article presents the development of soft lithography based polymer BioMEMS microfabrication techniques and systematic studies on the nanomechanical characterization of the polymer thin films and beams. Poly (methylmethacrylate) (PMMA) and poly (propyl methacrylate) (PPMA) are used to make the polymer beams for MEMS integration. The hardness, elastic modulus and creep behavior of PMMA and PPMA thin films and microstructures were measured using continuous stiffness measurement nanoindentation technique, and the scratch resistance of the polymer thin films was measured using a nanoscratch technique. The elastic moduli of the polymer beams were also obtained from bending tests performed by nanoindentation. To simulate the working environment of the polymer BioMEMS, the beams were also soaked in de-ionized water or heated up to human body temperature, and the nanomechanical properties were measured. © 2005 American Vacuum Society. [DOI: 10.1116/1.1861937]

I. INTRODUCTION

Microelectromechanical system (MEMS) technology and fabrication processes have found a variety of applications in biology and biomedicine, leading to the establishment of an entirely new field known as BioMEMS. The ability to use microfabrication processes to develop precision devices that can interface with biological environments at the cellular and molecular level has led to advances in the fields of drug delivery,1,2 tissue engineering,3–5 and biosensor6 technology. The miniaturization of fluidic systems using microfabrication techniques has led to new and more efficient devices for medical diagnostics and biochemical analysis.7

As the field of MEMS has progressed, alternative materials, especially polymers, have established an important role in the advancement of the technology. This trend has been driven by the reduced cost associated with polymer materials. Polymer microfabrication processes, including the micromolding technique discussed here, can be orders of magnitude less expensive than traditional silicon microfabrication processes. The use of polymers in the BioMEMS field has additional functional advantages, as polymers offer a wide range of material properties to allow tailoring of biological interactions for improved biocompatibility. Figure 1 shows examples of two polymer BioMEMS structures designed to measure cellular forces. The device on the left shows cantilevers anchored at the periphery of the circular structure, while the device on the right has cantilevers anchored at the four corners on the top and the bottom. The cell adheres to the center of the structure, and the contractile forces generated in the cell’s cytoskeleton cause the cantilevers to deflect. By using a polymer material with a low modulus, relatively large deflections can be achieved. The deflection of the polymer cantilevers can then be measured optically and related to the magnitude of the forces generated by the cell. This is just one example where a shift from silicon to polymer can decrease fabrication cost, improve biological interactions, and improve device functionality. Polymer materials most commonly used for biomedical applications include poly (methylmethacrylate) (PMMA), polydimethylsiloxane (PDMS) and polycarbonate. Another material of interest due to ease of fabrication is poly (propyl methacrylate) (PPMA), which has lower glass transition temperature ($T_g$) (35–43 °C)$^8$ than PMMA (104–106 °C)$^9,10$ which permits low temperature thermal processing.

As polymers have been incorporated into microfabricated systems, new fabrication technologies have been developed to accommodate. An entire class of microfabrication methods based on soft lithography11 has been developed for fabrication of polymer microstructures. Hot embossing12 and injection molding13 have been scaled down for fabrication of microfluidic diagnostic and analytical systems. Other technologies directly from silicon MEMS fabrication have been adapted for polymeric materials. For example, reactive ion etching has been modified from silicon micromachining to allow etching of polymeric microstructures.$^14$
Despite the large increase in the use of polymers in MEMS, little work has focused on the use of polymers as the primary structural and functional components in MEMS. Polymer properties such as low elastic modulus and good inherent biocompatibility make them an attractive option for biosensors such as cantilever force sensors and biological micromanipulators. In general, the current polymer microfabrication methods are not capable of producing suspended, mechanically independent microstructures such as those found in many silicon MEMS devices. In order to achieve such devices, new microfabrication methods that meet these needs must be developed. This work presents a microfabrication process capable of producing suspended polymer structures from thermoplastic polymers.

In addition to a more versatile polymer microfabrication method, a proper understanding of the micro- and nanoscale mechanics of polymeric material is critical for their proper implementation in MEMS device. The hardness, elastic modulus, creep, adhesion and scratch resistance of polymer materials in the micro/nano scale need to be characterized in order to design polymer MEMS. In particular, the elastic modulus of polymer beams measured from bending test is very important, because that value guides the design of devices such as the polymer cantilevers discussed previously. At the same time, it is very difficult to do the bending tests on polymer beams, as the polymers are extremely soft relative to silicon or ceramic beams. The hardness and elastic modulus measurements of polymers are also challenging, considering that polymers are viscoelastic materials, and creep usually occurs during the unloading segment of the conventional nanoindentation measurements. This may lead to errors in calculating the hardness and elastic modulus. A continuous stiffness measurement (CSM) nanoindentation technique is preferable in measuring the hardness and elastic modulus of polymers because a loading-unloading cycle is applied continuously at high frequency (45 Hz), when a normal load is applied, and the hardness and elastic modulus is measured continuously in a short time. In this way, the influence of the creep of polymers on property measurements can be greatly limited. The creep behavior of polymers can also be measured by CSM technique. Nanoscratch technique has been used widely to evaluate the scratch resistance and adhesion of hard coatings, but little work has been performed on polymer thin films.

Two polymer materials selected for this study are PPMA and PMMA. PPMA was chosen due to its relatively low glass transition temperature, allowing easy thermal processing of the material during the microfabrication process. PMMA was chosen due to its wide use in commercial biomedical applications. Table I summarizes published data on these materials. The primary operating environment for polymer BioMEMS is an aqueous solution (e.g., cell culture medium). Absorption of water by the polymer matrix has the

<table>
<thead>
<tr>
<th>Structure</th>
<th>$T_g$ (°C)</th>
<th>Water absorption after 24 h (%)</th>
<th>Thermal expansion coefficient (°K)</th>
<th>Thermal conductivity (0°–50°C (W m⁻¹ K⁻¹))</th>
<th>Elastic modulus (E) (GPa)</th>
<th>Hardness (H) (MPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>PMMA</td>
<td>104–106</td>
<td>0.3–0.4</td>
<td>2–3 × 10⁻⁶ (&lt;$T_g$)</td>
<td>0.193</td>
<td>3.1–3.3</td>
<td>4.43</td>
</tr>
<tr>
<td>PPMA</td>
<td>35–43</td>
<td>…</td>
<td>…</td>
<td>…</td>
<td>…</td>
<td>…</td>
</tr>
</tbody>
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Table I. Summary of physical properties of PPMA and PMMA materials.

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FIG. 1. Schematic of two different designs for polymer BioMEMS structures.
potential to change the properties of the material. In order to determine if immersion in water has a significant effect on material behavior, tests need to be run on samples that have been subjected to an aqueous environment for a considerable time. In addition, in the case of employing BioMEMS for in vivo or elevated temperature in vitro applications, the effect of human body temperature on polymer properties may be important. The \( T_g \) of PPMA (35–43 °C) is around body temperature and the \( T_g \) of PMMA (104–109 °C) is above body temperature. It is necessary to study how the mechanical properties of PPMA and PMMA are affected when operating at body temperature as compared to room temperature.

This article presents the development of soft lithography based polymer BioMEMS microfabrication techniques and systematic studies on the nanomechanical characterization of the polymer thin films and beams.

II. EXPERIMENTAL DETAILS

A. Microfabrication of polymer beams

Polymer microbeams were fabricated using a soft lithography based micromolding process along with standard photolithography. Photolithography using SU8-25 (MicroChem Corp.) negative tone photoresist was used to define channels in the photoresist which acted as a substrate for the polymer beams. The photoresist was spin coated on silicon at 2000 rpm to obtain a film thickness of approximately 25 \( \mu m \). The photoresist was then processed according to the manufacturer’s suggested processing parameters to attain patterned photoresist consisting of 25–30-\( \mu m \)-wide channels with a separation of 500 \( \mu m \) between each channel.

After defining the channels in the substrate material, a patterned PDMS mold with the desired polymer beam geometry was fabricated from a photoresist master. A detailed description of PDMS mold fabrication can be found elsewhere. Briefly, a layer of SU8-5 photoresist was spin coated on silicon and photolithography was used to define 5-\( \mu m \)-wide photoresist features separated by 45 \( \mu m \) gaps. A 10:1 ratio of T2 PDMS translucent base and curing agent were mixed thoroughly and poured over the photoresist master to transfer the pattern into the PDMS. The mold was then placed in a vacuum dessicator to remove bubbles incorporated during mixing. The sample was removed from the vacuum periodically, and a razor blade was used to remove surface bubbles. After the bubbles were completely removed, the PDMS mold was allowed to cure at room temperature for 48 h before removing it from the wafer.

Next, the PDMS mold was selectively coated with the polymer to form the microbeams and then transferred to the substrate. Figure 2 shows a schematic of the fabrication process used for making the polymer microbeams. As shown in Fig. 2(a), the polymers were dissolved in anisole, then spin coated on a mold with 5.5-\( \mu m \)-deep features at 4000 rpm yielded the 3.5 \( \mu m \) thickness beams used for the bending experiments. The coated mold was then brought into contact with a heated glass plate to promote adhesion of the contacting polymer materials. This process removed the polymer material from the raised surfaces of the mold, resulting in the polymer remaining only in the recessed portions of the PDMS mold. The glass plate was heated to 175 °C for both materials. As shown in Fig. 2(b), the selectively coated mold was then aligned with the photolitho-

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**Fig. 2.** Schematic of the micromolding process used in polymer beam fabrication. (a) Selective patterning of PDMS mold, (i) mold is spin coated with a polymer layer, (ii) mold is inverted and brought into contact with a heated glass plate, (iii) mold is removed from the plate, transferring the surface polymer onto the glass, (iv) mold is left with polymer only in the recessed features. (b) Stamping of the polymer beams. (i) Selectively patterned mold, (ii) mold is inverted and aligned with the photolithographically patterned substrate, (iii) mold and substrate are brought into contact under heat and pressure, (iv) polymer is transferred onto the photoresist and mold is removed.
graphically patterned substrate so that the PPMA or PMMA features ran perpendicular to the channels defined in the photoresist. The photoresist substrate was then heated and pressure was applied to the top of the mold to transfer the material onto the substrate. The transfer temperatures for PPMA and PMMA were 95 and 175 °C, respectively, and the transfer pressure for both materials was around 30 psi (0.21 MPa).

After removal of the mold, the polymer beams were left suspended over the channels in the substrate materials to provide a suspended region on which bending tests were performed. The elastic modulus, hardness, and creep were also determined by nanoindentation of the polymer structures in the regions supported by the substrate. Optical microscopy and image analysis were used to characterize the beam lengths and widths, and mechanical stylus profilometry was used to determine the height of the beams. The polymer beam samples studied in the article were 2–5 μm thick, 5 μm wide and 25–30 μm long. The tested beams were characterized individually to minimize errors due to geometrical considerations. The molecular weights of the PPMA and PMMA used in the experiments were approximately 250 000 and 75 000, respectively.

Thin films of the polymer materials were mainly used to determine scratch resistance and adhesion of the polymer materials. Hardness, elastic modulus and creep measurements were also conducted on polymer thin films in order to compare the nanomechanical properties of PPMA and PMMA thin films with the beams. The polymer thin films were spin coated on SU8-25 negative tone photoresist to simulate the environment of the microfabricated polymer beams described above. PMMA and PPMA were dissolved in anisole at various concentrations and spin coated for 1 min to achieve the desired film thickness. The samples were then baked at 95 °C for 2 min to remove any residual solvent. The films were annealed at 95 and 175 °C for PPMA and PMMA, respectively, to simulate the thermal processing used in the micromolding process. Thickness of the film samples was characterized by ellipsometry.

**B. Nanomechanical characterization**

The nanomechanical properties of polymer beams and thin films were measured using nanoindentation technique with a Nano Indenter II® (MTS Systems Corp.) equipped with the CSM option. This instrument monitors and records the dynamic load and displacement of the indenter during indentation with a force resolution of about 75 nN and displacement resolution of about 0.1 nm. A three-sided pyramidal diamond Berkovich indenter (radius 100–200 nm) was used. The hardness and elastic modulus were measured using the CSM technique, which has been described in detail in Ref. 20. Briefly, a harmonic force, \( F = F_0 e^{i\omega t} \), is added to the nominally increasing load, \( P \), on the indenter [see Fig. 3(a)]. The displacement response of the indenter at the excitation frequency and the phase angle between the two, \( h = h_0 e^{i(\omega t + \phi)} \), are measured continuously as a function of depth. Solving for the in-phase and out-of-phase portions of the response results in an explicit determination of the contact stiffness, \( S \), as a continuous function of depth. In the article, the indentation depth was controlled to 500 nm and the peak-to-peak load amplitude was 1.2 μN at a frequency of 45 Hz.

For indentation creep test, which was also performed using the CSM technique, the indentation load was 30 μN and the loading rate was 3 μN/s. The tip was held for 600 s after the indentation load reached 30 μN. To measure the mean stress and contact stiffness, during the hold segment the indenter was oscillated at a peak-to-peak load amplitude of 1.2 μN and a frequency of 45 Hz.

In nanoscratch studies, a conical diamond indenter tip having a tip radius of about 1 μm and an included angle of 60°, was drawn over the sample surface, and the load was ramped up until the film was delaminated. The normal load...
corresponding to the position of the onset of film delamination is defined as the critical load, which can be used to evaluate the adhesion strength of the film to the substrate. The 500-μm-long scratches were made by translating the sample while ramping the load on the conical tip to 1.0 mN. The translating speed was 5 μm/s. A typical scratch experiment consisted of three subsequent steps: (1) approaching the surface; (2) translating the sample at ramping load; (3) final unloading of the tip. After the scratch tests, the scratch-induced damage was measured by scanning electron microscopy (SEM).

Bending experiments were carried out using the nanoindenter with a 60° conical tip (radius 1 μm). Figure 3(b) shows the schematic of the polymer beam bending test. The uncoated tip penetrated the polymer beams easily and caused considerable plastic deformation during the bending test, which led to significant errors in the measurements. To avoid this issue, the diamond tip was dip coated with PMMA (about 1–2 μm thick) by dipping the tip in the 2% PMMA (wt/wt) solution for about 5 s. The comparison of polymer coated tip with uncoated tip will be discussed in more detail in the Results and Discussion section. During the bending test, the applied normal load was 10 μN, and the loading and unloading rate was 1 μN/s. The loading position was at the center of the beam. An optical microscope with a magnification of 1500× was used to locate the loading position. Then the specimen was moved by using a lead screw until the desired load location was aligned with the indenter tip. The resolution is about ±400 nm in the longitudinal and lateral directions, which can assure that the bending tests are made in the center of the 5-μm-wide and 25–30-μm-long polymer beams. In order to maintain that resolution, the “recalibration of microscope to indenter distance” (Operating Instructions, Nano Indenter II) must be carried out frequently, because it is easy for the system to lose calibration. It should be noted that before each bending test, the “recalibration of microscope to indenter distance” was performed on a polycaprolactone (PCL) sample instead of the standard Al sample. Since the recalibration procedure requires making indents, the Al sample is not a suitable recalibration sample for the PMMA coated tip, considering that Al has higher hardness than PMMA. So PCL, which is much softer than PMMA, was used to do the recalibration.

To simulate the aqueous working environment of the polymer BioMEMS, the bending tests were also performed on soaked samples in addition to the dry samples and the results were compared. To make the soaked samples, the PMMA and PPMA beams were soaked in de-ionized (DI) water for 36 h before the bending tests. In addition, the nanoindentation tests were conducted in a range of temperature (22–37.5 °C) to study the effect of temperature on nanomechanical properties of polymer beams. In order to heat the sample, the polymer beam sample was placed on a heating stage [see Fig. 3(b)], which can increase the temperature up to 200 °C.21 The heat-generating elements of the heating stage were ohmic resistors encapsulated in a steel holder and kept in good thermal contact by using thermal paste. J-type thermocouples were used to measure the sample temperature. A thermal controller and a solid-state relay were used to control the temperature by adjusting on/off time. A glass plate was attached at the bottom of the heating stage to isolate it thermally from the lateral motion stage, where the heating stage was mounted.

III. RESULTS AND DISCUSSION

A. Hardness, elastic modulus, creep and scratch resistance

Polymer microbeams with thickness of about 2 to 5 μm were fabricated. Figure 4 shows the low and high magnification SEM images of the PPMA beams. It can be seen that the beams were transferred to the substrate with high fidelity and no sagging of the suspended portion of the beams was observed.

The hardness (H), elastic modulus (E) and creep of PPMA and PMMA beams were measured at the supported region of the beams. In Fig. 4, the indentation location, where the H, E and creep were measured, is indicated by an arrow. Figure 5(a) shows the H and E of 2.9 μm PPMA and 3.4 μm PMMA beams as a function of contact depth. The H and E were calculated by averaging the H and E values obtained at contact depth of 100 nm from five indents. The H (0.41 GPa) and E (5.0 GPa) of the PMMA beams are both higher than H (0.19 GPa) and E (3.7 GPa) of the PPMA beams. Table I compares the present data with the published data. We note that the E of PMMA is comparable to the published data. It should be noted that the H and E were also measured on 3 μm PMMA and 3 μm PMMA thin films/SU8-25, and the results were very close to the values obtained from the beams.
In the creep tests, the changes in indentation displacement, mean stress (hardness) and contact stiffness can be monitored during the hold segment with the CSM technique. Figure 5(b) shows the indentation displacement, mean stress and contact stiffness as a function of time for 2.9-μm-thick PPMA and 3.4-μm-thick PMMA beams. The indentation depths of both polymer beams increase with time. The PPMA beam exhibits a faster increase in indentation depth (from about 60 to about 90 nm) than the PMMA beam (from about 30 to about 50 nm). This indicates that a higher hardness is associated with a higher creep resistance. In contrast with indentation displacement, the mean stresses of both polymer beams decrease with time, indicating that stress relaxation occurred during the hold segment. The creep tests were also conducted on the polymer thin films and the results were similar to the polymer beams.

As mentioned in the experimental section, the polymer films were annealed at 95 and 175 °C to PPMA and PMMA, respectively, to simulate the thermal processing used in the micromolding process for polymer beam fabrication. The annealing treatment seemed to effectively simulate the beam
fabrication process, because the H, E and creep measurements show that the PPMA and PMMA thin films have similar mechanical properties with PPMA and PMMA beams. In order to evaluate the scratch resistance and adhesion of PPMA and PMMA beams/SU8-25 substrate, nanoscratch measurements were needed. Since the polymer beams were too narrow (width 5 μm) to perform nanoscratch tests, the nanoscratch experiments were conducted on 300 nm PPMA and PMMA thin films/SU8-25, assuming that they have similar scratch resistance and adhesion with PPMA and PMMA beams.

Figure 5(c) shows the scratch depth profiles, coefficient of friction, and SEM images of three regions across scratches: at the beginning of the scratch (indicated by A on the scratch depth profile), at the point of initiation of damage (indicated by B on the scratch depth profile), and towards the end of the scratch (indicated by C on the scratch depth profile) for PPMA and PMMA thin films. The scratch results of polymer thin films are very different from metallic and ceramic thin films.22,23 First, the coefficient of friction value was very high, even greater than one, and the coefficient of friction profile jumped up and down frequently. Second, after the polymer thin films were damaged, the scratch depth profile jumped up and down considerably, indicating the scratch tip moved up and down during scratching. The SEM images of PPMA show that during scratching, the materials were plowed and accumulated in front of the tip, instead of being pushed aside, because the polymers were so soft. When the plowed materials reached a certain amount, the tip was almost stuck, so the tip jumped up instead of ramping down. This may explain the oscillation in the scratch depth profile.

At region B, the coefficient of friction increased and the in situ scratch depth also changed abruptly, indicating that the film was delaminated. The critical load of PPMA and PMMA is about the same, around 0.22 mN.

B. Bending tests

The bending tests were first performed on PPMA (3.5 μm thick, 25 μm long and 5 μm wide) beams with an uncoated diamond conical tip (radius 1 μm). The load-displacement curve was shown in Fig. 6(a). It can be seen that at a low load of about 2 μN, the in situ displacement and the residual depth were both about 130 nm, indicating that severe plastic deformation occurred during the bending test. It is also interesting to find that at the displacement range of about 85–130 nm, the load was almost constant, which means creep occurred. The much greater hardness of the diamond tip relative to the polymer beam caused the tip to penetrate the beam much more than bending it, causing permanent damage to the beam. Since most of the load-displacement curve came from plastic deformation, the curve cannot be used to calculate the elastic modulus. To avoid this issue, the tip was dip coated with PMMA. Figure 6(b) shows the load-displacement curves of PPMA and PMMA beams measured with the PMMA coated tip. The loading curve was linear and the unloading curve does not exhibit much plastic deformation, which means that the bending tests were made mostly in the elastic region of the samples and the elastic modulus can be calculated from the loading curve. Any slight degree of deformation of the tip coating during the bending tests was assumed negligible compared to the beam bending.

Elastic modulus of the beams can be estimated by equations based on the assumption that the beams follow linear elastic theory of an isotropic material. For a fixed elastic beam loaded at the center of the span, the elastic modulus is expressed as

$$E = \frac{1}{\frac{l^3}{192I} m},$$

where \(l\) is the beam length, \(I\) is the area moment of inertia for the beam cross section and \(m\) is the slope of the load-

![Fig. 6. (a) Bending result on PPMA beam with uncoated diamond conical tip, (b) bending results on PPMA and PMMA beams with PMMA coated diamond conical tip.](image-url)
displacement curve during bending. The area moment of inertia is calculated from the following equation:

\[ I = \frac{b}{12} h^3, \]  

(2)

where \( b \) is the beam width and \( h \) is the beam thickness. Based on Eqs. (1) and (2) and Fig. 6(b), the calculated elastic moduli for PPMA and PMMA beams are 0.7±0.1 and 2.0±0.1 GPa, respectively. The standard deviation was calculated based on five measurements.

Both the calculated elastic moduli of PPMA and PMMA beams from bending tests are lower than the values obtained from indentation. This may be due to several causes. The last step of fabricating the polymer beams was to bring the mold in contact with the photoresist substrate and the photoresist substrate was then heated and pressure was applied to the top of the mold to transfer the polymer beams onto the substrate. Since pressure was applied in this step, the boundary between the suspended region of the beam and the supported region of the beam might be deformed to some extent due to the pressure. During the bending tests, the maximum moments were generated at the ends (boundaries) (negative moment) and under the loading point (positive moment). If the strength of the ends was decreased due to the pressure-induced deformation, then the beam would be easily bent, leading to a low elastic modulus. In addition, because the suspended region of the beam was not in contact with the substrate as the supported region of the beam was, the heat transfer for the two regions would be different, which might also affect the mechanical properties. Furthermore, although the diamond tip was coated with PMMA, there might still be some plastic deformation of the beam surface during the bending tests, which would reduce the slope of the loading curve.

C. Effect of soaking and temperature

Figure 7(a) shows the effect of soaking on elastic modulus of PPMA beams. After soaking, the calculated elastic modulus (0.5±0.1 GPa) of PPMA (soaked) was lower than the elastic modulus (0.7±0.1 GPa) obtained at dry conditions (unsoaked). But for PMMA, after soaking, the calculated elastic modulus does not change (figure not shown). The more open structure of PPMA vs PMMA might allow more water to penetrate into the matrix, thus having a greater effect on the modulus than with PMMA (closed, hydrophobic structure). The open structure of PPMA is due to the larger side chain compared to the side chain of PMMA.

Figure 7(b) shows the effect of temperature on load-displacement curves of PPMA beams. As the temperature increased from room temperature (22 °C) to human body temperature (37.5 °C), the slope of the loading curve decreased. The \( T_g \) of PPMA is 35–43 °C, so it is understandable that the elastic modulus, thus the slope of the loading curve, would decrease when the temperature increased beyond 35 °C. However, since the \( T_g \) of PMMA is about 100 °C, at 37.5 °C, the load-displacement curve of PMMA beam was similar to 22 °C (figure not shown).

![Fig. 7. (a) Soaking effect on bending of PPMA beam, (b) temperature effect on bending of PPMA beam, (c) soaking and temperature effects on H and E of PPMA and PMMA beams.](image)
The soaking and temperature effects on hardness and elastic modulus of PPMA and PMMA beams were also studied and the results are shown in Fig. 7(c). It can be seen that for PPMA beam, the hardness and elastic modulus decreased after 36 h soaking in DI water, and also decreased at 37.5 °C, which is consistent with the bending test results. For the PMMA beams, the hardness decreased after soaking, and also decreased slightly at 37.5 °C. However, the elastic modulus of the PMMA beams did not change after soaking or heating up to 37.5 °C, which is in agreement with the bending results. The lower elastic modulus of PPMA compared to PMMA makes it an attractive option for certain applications, such as cantilevers. However, the changes in the properties of PMMA as a function of aqueous environment and temperature would require careful calibration at operating conditions when implementing PPMA microstructures in a biological setting.

IV. CONCLUSIONS

This article presents the development of soft lithography based polymer BioMEMS microfabrication techniques and systematic studies on the nanomechanical characterization of the polymer thin films and beams. Based on the above discussion, the following conclusions can be drawn:

1. The well patterned polymer microbeams were fabricated by using the soft lithography based micromolding process along with standard photolithography.
2. The hardness, elastic modulus and creep behavior of PPMA and PMMA beams were measured using nanoindentation CSM technique. The hardness, elastic modulus and creep resistance of PMMA beam are higher than PPMA beam.
3. With polymer coated conical tip, the bending moduli of PPMA and PMMA beams were measured using nanoindentation technique. The modulus calculated from the bending test was lower than the modulus measured by indentation.
4. The scratch behavior of PPMA and PMMA films is different from metallic and ceramic films in that the in situ displacement oscillated and coefficient of friction are greater than one during the scratch test. This might be because the polymer is so soft and easily plowed.
5. After 36 h soaking in DI water, the hardness and elastic modulus of PPMA beam were decreased. The soaking seems to have no effect on elastic modulus of the PMMA beam. At 37.5 °C, the hardness and elastic modulus of the PPMA beam were decreased as compared to room temperature, while this temperature appears to have no effect on elastic modulus of the PMMA beam.

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26. Rohm and Haas General Information on PMMA.