Adhesion properties of polymer/silicon interfaces for biological micro-/nanoelectromechanical systems applications

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Polymers are used in biological micro-nanoelectromechanical systems (BioMEMS/NEMS) applications due to their desirable mechanical properties, biocompatibility, and reduced cost relative to silicon microfabrication processes. Understanding the interfacial properties of the films that are used in BioMEMS/NEMS serves as a useful tool in obtaining higher device yield and greater mechanical reliability. In this study, polystyrene (PS) and glycidyl-ether-bisphenol-A novolac polymer (SU8) on silicon substrates were investigated. SU8 is a commonly used material in MEMS/NEMS fabrication, while PS is evaluated for its potential use in BioMEMS/NEMS for interaction with biological cells. The aim is to examine the delamination of the interfaces. Nanoindentation was employed on the PS/Si and SU8/Si film systems coated with a thin metallic layer of Cr to facilitate delamination. The interfacial adhesion energy was determined from measuring the size of the resulting delamination and the contact radius. Scale effects were investigated by comparing the behavior of thin and thick PS and SU8 films, where a thickness dependence on the interfacial adhesion energy was observed. In addition to room temperature testing, film delamination experiments were conducted at 50 and 70 °C by fitting the nanoindenter with a heating stage in order to study temperature effects. Nanoindentation-induced delamination is demonstrated for microstrips of PS and SU8 and the measured interfacial adhesion energy is compared to those obtained from films. © 2007 American Vacuum Society. [DOI: 10.1116/1.2435390]

I. INTRODUCTION

Biological micro-/nanoelectromechanical systems (BioMEMS/NEMS) applications have expanded into a variety of areas such as drug delivery, microfluidics, and chemical and biological sensing. Various structures have been created either by traditional silicon-based batch processes or the newer polymer-based fabrication techniques. Polymers are advantageous for certain device applications because they offer a number of advantages relative to silicon. In addition to being relatively inexpensive, many polymers are also biocompatible, so they may be integrated into biomedical devices with minimal detrimental effects to the host or on the biofluids. An improvement in device functionality (relative to silicon) is also possible due to the mechanical properties of the polymer used, which have properties much closer to biological tissues. An example is shown in Fig. 1, which is a polystyrene (PS)-based sensor for measuring the mechanical forces exerted by a cell. The adhesive and contractile forces generated by the cell cause the cantilevers in the sensor to deflect which is detected by an optical method. For this application, large, detectable deflections can be observed by selecting a polymer over silicon, thus providing sensitivity to very low magnitude forces.

Two polymer materials were investigated in this study, PS and glycidyl-ether-bisphenol-A novolak (referred to by its trade name SU8). The chemical structure and physical properties of these materials are listed in Table I. Polystyrene is particularly desirable for BioMEMS due to its ubiquitous use in tissue culture applications. Plasma treated polystyrene is the most commonly used material for in vitro cell biology studies of adherent cells. Thus, there is a wealth of knowledge about cellular behavior on polystyrene. In addition to tissue culture polystyrene, various surface modification techniques can be employed for functionalizing the PS surface in order to promote cell attachment and proliferation. Oxygen plasma modified polystyrene has been shown to improve cell growth, proliferation, and expression of cellular adhesion proteins proportional to the surface oxygen concentration.
The previous knowledge of cellular interactions with polystyrene makes it a logical choice to use in BioMEMS/NEMS devices for cellular interactions.

SU8 is a photosensitive material commonly used in MEMS processing. It is employed as a structural material in fabrication of MEMS components. Previous studies have shown that SU8 exhibits relatively good biocompatibility, making it a strong candidate for biological applications. SU8 has been employed in a number of BioMEMS devices. An example of an SU8 application is shown in Fig. 2. The micrograph shows an image of an SU8 microgripper used for manipulation of single cells in solution. The use of SU8 as the structural material allows the gripper to be electrothermally actuated at low voltage, thus providing an attractive actuator mechanism for operation in ionic solutions such as saline and cell culture medium.

The integrity of the interface between the structural components of MEMS devices and the underlying substrate is an important technological issue. Numerous studies are available for the adhesion characterization of metals and inorganic materials due to the demands of various industries such as aerospace and microelectronics. Examples include peel testing, double cantilever beam testing, and four-point bending. The latter two methods have been developed for macroscopic sandwich structures, where two substrates (one containing the film of interest) are bonded together prior to adhesion assessment. From these techniques, either the strain energy release rate (G) (referred to in this article as the interfacial adhesion energy) or the stress intensity (K) is reported.

Indentation can also be employed for thin film adhesion tests. In this technique, the applied load from the indenter initiates and propagates a delamination, the extent of which is related to the measured loads and displacements. An advantage of this method is the ease of sample preparation, relative to the traditional sandwich specimens. The induced fracture allows quantitative assessment of adhesion. However, it is difficult to induce delamination if the material is ductile or if the film is strongly adhering. To overcome this limitation, an overlayer is deposited over the film of interest. The overlayer is a material with a high intrinsic stress such as Cr which provides additional driving force for delamination. This procedure has been successfully employed for spontaneous delamination of metallic thin films and microstrip structures as well as for indentation-induced delamination of films.

There is a paucity of adhesion studies on materials and structures that are relevant for BioMEMS/NEMS applications. Understanding the interfacial properties of the films that are used in BioMEMS/NEMS serves as a useful tool in obtaining higher device yield and greater mechanical reliability. In future device generations, features with dimensions on the order of 100 nm or less will be integrated, so it is important to understand how length scale affects interfacial adhesion. It is important to understand the adhesion properties of three-dimensional structures such as the microstrips since potential BioMEMS/NEMS applications contain thin features on a substrate. The appropriate materials selection, design, and processing changes can be envisaged if the strength of the component interfaces is quantified.

This article presents the adhesion characterization between silicon and two polymers that are used in BioMEMS/NEMS applications, PS, and SU8. The polymer SU8 was evaluated due to its common application as a structural material in MEMS/NEMS and PS was evaluated as a nontraditional material for use in BioMEMS/NEMS for cellular interactions. A Cr overlayer on the polymer films was employed to facilitate the nanoindentation-induced delami-
nation process. Atomic force microscopy (AFM) was used to characterize the size of the resulting blisters, which was then used to calculate the interfacial adhesion energy (G) (or the amount of energy available to form two new surfaces as the crack expands). The delamination response of these films was examined at elevated temperatures by fitting a heating assembly into the indenter’s sample stage. Ultimately, the goal is to apply this delamination method to polymer microstrips, which represents structures that can be incorporated in a BioMEMS/NEMS device. Indentation-induced delamination of PS and SU8 microstrips on silicon substrates is demonstrated for the first time.

II. EXPERIMENTAL DETAILS

A. Materials and sample preparation

Four types of samples were prepared for evaluation, namely, films of PS and SU8 and microstrips of PS and SU8, all on silicon substrates. For PS, the films investigated were 7100, 870, 300, and 120 nm thick. The SU8 films studied were 7400, 1470, 440, and 220 nm thick. For all of the SU8 films and the 7100 nm PS film, a razor blade was used to carefully remove a portion of the film and a stylus profiler (Veeco, Inc.) was used to measure the step height. For the other PS films, a Nanometrics Nanospec 3000PHX film thickness measurement system was used to determine the film thickness. Profiler data are the average of three measurements and Nanospec data are the average of five measurements. This thickness range was selected to investigate the effect of the length scale on the adhesion characteristics of these polymers. Studies on films serve as the benchmark for establishing the appropriate sample preparation and testing procedure for the microstrips. The latter are of interest since these structures can be found in a potential BioMEMS/NEMS device (such as the cell force sensor shown in Fig. 1).

For PS, the microstrips were 3400 and 1300 nm thick, while for SU8, they were 3500 and 2400 nm thick, which are within the thickness range that can be employed in device structures. The thickness of the microstrips was measured using a stylus profiler.

To prepare PS films on silicon substrate, PS pellets (melt flow index of 4.0, Sigma-Aldrich) were dissolved in anisole (Acros Organics) to prepare three solutions of varying concentrations (20%, 10%, and 3% wt/wt) in order to achieve the desired film thickness. Silicon substrates with dimensions of 20 × 20 × 0.5 mm² were cut from precleaned silicon wafers. Films with four different thicknesses were prepared by spin coating the solutions on the silicon substrates. The 20% wt/wt solution was spin coated at 1000 rpm to prepare 7100 nm thick film, the 10% wt/wt solution was spin coated at 3000 rpm for the 870 nm thick film and the 3% wt/wt solution was used to create 1470, 440, and 220 nm thick films, respectively. The coated films were exposed to UV and baked to promote cross-linking of the SU8. The SU8 films were prebaked, exposed to UV, and postexposure baked according to the parameters provided in the manufacturer’s data sheets.

SU8 microstrips were fabricated using standard photolithography techniques. Photoresist was spin coated on p-type (100) silicon wafers. Films were baked to remove the residual solvent and exposed to UV through a chrome/glass photomask with 5 μm features to create the microstrips on the exposed areas. These were then processed according to the manufacturer’s suggested parameters. After processing, feature heights were measured using a profiler.

Polystyrene microstrips were fabricated using the soft-lithography-based procedure shown in Fig. 3. Standard photolithography cannot be used because PS, unlike conventional photoresists, is not photosensitive. In addition, PS is thermoplastic such that micromolding is a suitable processing technique. A poly(dimethylsiloxane) (PDMS) mold was first fabricated from an SU8 photoresist master as described previously. The SU8 master consisted of 5 μm wide lines separated by 45 μm. Two different mold depths were used in this work. One mold had 2.3 μm deep channels and the other had 5.3 μm deep channels. A 10:1 ratio of T-2 PDMS translucent base and curing agent (Dow Corning) were mixed thoroughly and poured over the photoresist master to transfer the pattern into the PDMS. The mold was allowed to cure for 48 h before removal from the master.

Solutions of PS in anisole were then spin coated on the PDMS molds. PS solution concentrations ranged from 3% to 7.5% and spin speeds ranged from 1000 to 4000 rpm. After coating, the molds were brought into contact with a glass slide heated to 200 °C to remove the PS from the raised portions of the PDMS mold. Slight pressure of 17 kPa (2.5 psi) was applied to the top of the mold during the first stamping. To remove the PS from the recessed channels, the mold was applied to a silicon substrate heated to 125 °C. For the 5.3 μm deep mold, pressures ranging from approximately 550–1100 kPa (80–160 psi) were used to remove the polymer from the mold. For the 2.3 μm deep mold, pressures ranging from 280 to 350 kPa (40–50 psi) were used. For the 5.3 and 2.3 μm deep molds, feature heights ranged from 2.3 to 4.3 and 1.3 to 1.7 μm, respectively, depending on the solution concentration. Feature heights were measured using a profiler.

All the film and microstrip samples were coated with an overlayer of 600 nm thick Cr using an electron beam (e-beam) evaporator (Denton Vacuum).

B. Nanomechanical characterization

The hardness and elastic modulus of PS and SU8 films were evaluated using a Nano Indenter II® (MTS Systems Corp.) in the continuous stiffness mode (CSM) equipped
with a diamond Berkovich tip. In this mode, the dynamic load and displacement of the indenter are monitored during indentation with a force resolution of about 75 nN and displacement resolution of about 0.1 nm. The CSM technique has been described in detail in Li and Bhushan. Briefly, a harmonic force, \( F = F_0 e^{i\omega t} \), is added to the nominally increasing load \( P \) on the indenter. 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 contact depth, which is the actual depth that the sample is in contact with the indenter tip. The contact depth is obtained by subtracting the elastic displacement of the surface at the contact perimeter from the indentation displacement. In this article, the maximum indentation displacement was controlled to 500 nm and the peak-to-peak load amplitude was 1.2 \( \mu \)N at a frequency of 45 Hz.

**C. Adhesion testing**

Four PS (thicknesses of 7100, 870, 300, and 120 nm) and four SU8 films (thicknesses of 7400, 1470, 440, and 220 nm) were indented at room temperature. The delamination was induced using a conical diamond tip with 90° included angle. The range of applied load was 100–600 mN.

The 870 nm thick PS film and the 1470 nm thick SU8 film were also indented at 50 and 70 °C using the same load used in the room temperature test. In order to heat the sample, the wafer with the film was mounted on a heating stage. The heat-generating elements of the heating stage were Ohmic resistors encapsulated in a steel holder. 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.

The indentation-induced deformation of the films was characterized using a NanoScope III (Veeco, Inc.) AFM. The AFM was operated in contact mode. From the section analysis feature of the NanoScope III software, the dimensions of the contact, plastic zone, and delamination radii were obtained.

Two PS and two SU8 microstrip samples were indented at room temperature. The PS samples were 3400 and 1300 nm thick, while the SU8 samples were 3500 and 2400 nm thick. The applied load range was from 100 to 200 mN.

**III. RESULTS AND DISCUSSION**

**A. Nanomechanical characterization**

Hardness \( (H) \) and elastic modulus \( (E) \) plots as a function of contact depth are presented in Fig. 4. Three indents were made on the PS and SU8 films, and the representative data are shown. These measurements are necessary to calculate the interfacial adhesion energy \( (G) \) as described later. The \( H \) and \( E \) values at the near surface are slightly elevated primarily due to error in the calculation of the contact depth at small indentation sizes. This error is based on the assumption that the materials behave as a rigid plastics as opposed to the viscoelastic nature of the materials used in this study. Surface roughness of the sample can also lead to errors in the contact depth for shallow indents. A plateau is reached (approximately at a contact depth of 200 nm) as the indenter penetrates deeper into the material. For subsequent calculations, these plateau values were used as the data at the corresponding depth range are more reproducible. Data at 22, 50, and 70 °C are shown since the adhesion characteristics at
higher temperatures were also examined. For both PS and SU8, H and E only dropped slightly (<5%) at 70 °C. No significant decrease was observed since the test temperatures are well below the glass transition temperature for each polymer.

B. Adhesion testing of films

Indents were made at three different locations on the same film sample. The load-displacement plots, postindent AFM images, and the height profile across the center of the indents for one representative indent for the PS and SU8 films are presented in Fig. 5. The load-displacement curves are for the lowest possible loads where an abrupt change in displacement at constant load is observed. This is generally taken as the onset of film delamination. At loads lower than the excursion point, only a small amount of hysteresis is present and blisters are not seen, indicating that delamination did not occur. Comparing PS to SU8, it is observed that higher loads were necessary to observe delamination in the latter. With regards to film thickness, a substrate interaction effect is observed for both PS and SU8 films thinner than 500 nm. These films appear stiffer since the deformation response is present from the film and the stiffer silicon substrate. As a result, a higher load is necessary to induce delamination in the thinner films.

The method for evaluating the interfacial adhesion energy (G) of the film is based on the model of Thouless where the debonded portion of the film is modeled as an elastic disk with plane stress conditions in the axial direction. The stress distribution in the disk is determined by applying elasticity solutions with a zero-displacement boundary condition at the outer diameter of the delamination (r=a) and a fixed radial stress σ0 at the inner diameter of the delamination, which is the outer edge of the contact (r=c). The stress distributions in the delaminated portion of the film are

\[ \sigma_r = \sigma_0 \left[ 1 + \frac{c^2}{\alpha^2} \right] \left[ 1 + \frac{c^2}{a^2} \right], \]

\[ \sigma_\theta = \sigma_0 \left[ 1 - \frac{c^2}{\alpha^2} \right] \left[ 1 + \frac{c^2}{a^2} \right], \]

(1)

where c is the delamination radius and a is the contact radius (both obtained from the AFM images), r is the radial position, the subscripts ρ and θ refer to the radial and circumferential directions, α = (1−ν)/(1+ν), and ν is Poisson’s ratio of the film. The radial stress \( \sigma_\rho \) is obtained by applying the Tresca yield criteria to the plastically deformed contact zone. The vertical stress in the contact zone is equal to the mean indenter contact pressure, taken to be the film hardness H. From the Tresca yield criteria, the radial stress \( \sigma_\rho \) is then equal to \( \sigma_{ys} - H \), the difference between the yield strength of the film (\( \sigma_{ys} \)) and the hardness (Hf) of the film.31 The strain energy U is obtained by evaluating the following integral over the film volume:

\[ U = \frac{\pi r_f}{E_f} \int_a^{r_f} \left[ \sigma_r^2 + \sigma_\theta^2 - 2\nu_\rho \sigma_r \sigma_\theta \right] r dr, \]

(2)

where \( r_f \) is the film thickness and \( E_f \) is the film’s elastic modulus.

The interfacial adhesion energy G is found by differentiating the strain energy with respect to crack area30 and is given as

\[ G = \frac{2(1-\nu_\rho)(\sigma_{ys} - H_f)^2 r_f}{E_f(1+\nu_f + (c/a)^2(1-\nu_f))}, \]

(3)

One approach to calculate yield strength of the film is by the imaging method, where \( \sigma_{ys} \) is calculated from the pileup radius b of an indent with a maximum imposed load \( P_{\max} \) as follows:32

\[ \sigma_{ys} = \frac{3P_{\max}}{2\pi b^2}. \]

(4)

Since the in situ determination of the pileup radius is not possible with the indenter system used in this study, it is necessary to use another approach. In another approach used here, the yield strength of PS and SU8 was approximated based on the known relationship between \( \sigma_{ys} \) and the material hardness.33 The yield strength is related to the hardness by a material-specific constant k such that \( \sigma_{ys} = H/k \). The

| Table II. Summary of measured interfacial adhesion energies (G) as a function of film thickness for PS and SU8 at room temperature. |
| --- | --- | --- | --- |
| | PS | | SU8 |
| Film thickness (nm) | G (J/m²) | Film thickness (nm) | G (J/m²) |
| 7100 | 0.76 | 7400 | 1.70 |
| 870 | 0.21 | 1470 | 0.20 |
| 300 | 0.09 | 440 | 0.08 |
| 120 | 0.02 | 220 | 0.06 |
constant $k$ depends on the state of strain underneath the indenter. For various glassy polymers, it has been found that $k$ ranges from 2 to 3. In this study, the $k$ for PS and SU8 is taken as 2.5.

Table II summarizes the room temperature interfacial adhesion energies of PS and SU8 as a function of film thickness by using measured values of $H_f$, $E_f$, $c$ and $a$, and reported value of $v_f$ in Table I. The observed value of $G$ for
PS/silicon is close to that reported for PS/glass (0.6 J/m² for a 3.5 μm film) using Eq. (3). For both polymers, thicker films had higher $G$ values, indicating a length scale effect. This is because the interfacial adhesion energy (or the strain energy release rate) has two component mechanisms. First is the work of adhesion, which is the energy of bond separation across the interface. Second is the plastic energy dissipation within the polymer. For the thicker films, the plastic zone increases and this contributes to a higher measured value for the interfacial adhesion energy.
Table III. Summary of measured interfacial adhesion energies ($G$) as a function of temperature for PS and SU8 films.

<table>
<thead>
<tr>
<th>Temperature (°C)</th>
<th>870 nm PS</th>
<th>1470 nm SU8</th>
</tr>
</thead>
<tbody>
<tr>
<td>22</td>
<td>0.21</td>
<td>0.20</td>
</tr>
<tr>
<td>50</td>
<td>0.23</td>
<td>0.18</td>
</tr>
<tr>
<td>70</td>
<td>0.32</td>
<td>0.22</td>
</tr>
</tbody>
</table>

Figure 6 presents the load-displacement curves, AFM images, and the corresponding section profiles for indents on PS (870 nm thick) and SU8 (1470 nm) at 22 °C (room temperature), 50 °C, and 70 °C. As summarized in Table III, the calculated $G$ for the PS films is similar at room temperature and 50 °C, and then exhibits a slight increase at 70 °C. For SU8, the $G$ is practically the same at all test temperatures. This observed elevation of the interfacial adhesion energy of polystyrene at higher test temperatures can be accounted for by the increased plastic energy dissipation as polymer chain motion is more likely at higher temperatures. However, this is not applicable to SU8, since it is a cross-linked polymer and chain motion is suppressed. In addition, SU8 has a much higher glass transition temperature (Table I). These reasons account for the insensitivity of the $G$ value of SU8 to these test temperatures.

Figure 7 shows the selected high magnification scanning electron microscopy (SEM) micrographs of 120 nm thick PS and 220 nm thick SU8 films after indentation. A crater is formed on the material that was in contact with the indenter tip. Around this is the blister which appears to be raised relative to the rest of the film. Along with the AFM images (Fig. 5), this indicates that both the PS and SU8 delaminated in a three-stage mechanism similar to that proposed by Li and Bhushan to account for the delamination in diamond-like carbon coatings on silicon, as shown in Fig. 8. In the figure, $t$ is the film thickness, $V_i$ is the volume displaced by the indenter tip, and $\sigma_i$ is the stress induced by the indentation, which is counteracted by the residual stress $\sigma_r$. In this mechanism, the first stage is the compression of material induced by the application of the load, followed by delamination and buckling of the material around the contact area. The third stage is the formation of a ringlike crack on the delaminated film.

Radial cracking is observed on the surface. This is attributed to the high applied loads that are necessary to induce delamination. From these images alone, it is difficult to assess whether the radial cracks run through the thickness of the polymer film, or if it is only on the Cr layer.

C. Adhesion testing of microstrips

Ultimately, the goal is to characterize microstrips of PS and SU8 since these structures can delaminate from the substrate of a potential BioMEMS/NEMS device due to various processing or service conditions. Figure 9 presents the load-displacement curves and SEM images of the PS and SU8 microstrips taken after indentation, where failure of the polymer/silicon interface was observed in all cases. The loading portion of the load-displacement plots starts with a low stiffness (i.e., low slope) that gradually increases as the thickness of the microstrip is approached. This corresponds to the gradual compression of the area being indented in the microstrip prior to failure of the polymer/silicon interface. From the load-displacement plots, it is observed that a lower load is sufficient to induce observable deformation in the thinner PS microstrip (1300 nm thickness) compared to the rest of the sample set. This can be correlated to the extent of indenter penetration into the thickness of the microstrip. The other samples were thicker and required a higher imposed load.
load to enable compression of the material. Unlike the thin film samples, a substrate interaction effect was not observed.

In the SEM images, the observed cracks at the microstrip/substrate interface originating from the edge of the indent indicate delamination. Certain areas of the microstrips were raised, indicating that buckling also occurred. These images demonstrate that indentation can be successfully employed to induce interfacial failure in these polymer microstrip structures. In addition, the images also imply that these microstrips fail in a manner similar to films, as the failed areas of

![Graph and SEM images of PS and SU8 microstrips.](image-url)
the microstrips are similar to stage 2 in the schematic shown in Fig. 8 (which was originally proposed to account for film delamination).

Table IV lists the calculated $G$ values for these four microstrips. The values were determined, using Eq. (3), using the SEM images to determine the contact and delamination radii ($a$ and $c$, respectively). It is noted that these $G$’s should be considered as estimates since the model for Eq. (3) is for the evaluation of film adhesion. The microstrip $G$ values are expected to be lower compared to that obtained from the films. It is expected that these strips would be easier to delaminate since a smaller volume of material needs to be displaced. This explains the results for the SU8 microstrips relative to the thin films. Compared to SU8, $G$ values are higher for the PS microstrips. The increased adhesion strength in PS is attributed to the applied heat and pressure in the microstrip stamping process that could lead to a stronger polystyrene/silicon interface.

### IV. CONCLUSIONS

Delamination of PS and SU8 films on silicon substrates has been induced through indentation and the interfacial adhesion energy ($G$) has been quantified. For both materials, thicker films had higher interfacial adhesion energies due to plastic dissipation in the polymer. The same phenomenon is proposed to account for the observed slight increase in the adhesion energy at elevated temperatures in the non-cross-linked polymer (PS). The SU8 film is not sensitive to the test temperatures used because it has a higher glass transition temperature and is cross-linked.

This delamination method, which was originally developed for films, can be applied to quantify the adhesion properties of polymer microstrip structures. Indentation-induced failure of microstrips of PS and SU8 has been demonstrated and characterization of the failed areas reveals that these structures fail in a manner similar to thin films. The higher $G$ of the PS microstrips relative to SU8 is attributed to the difference in the fabrication method used.

### ACKNOWLEDGMENT

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**Table IV. Summary of measured interfacial adhesion energies ($G$) as a function of thickness for PS and SU8 microstrips.**

<table>
<thead>
<tr>
<th>Thickness (nm)</th>
<th>PS $G$ (J/m$^2$)</th>
<th>SU8 $G$ (J/m$^2$)</th>
</tr>
</thead>
<tbody>
<tr>
<td>3400</td>
<td>0.53</td>
<td>0.11</td>
</tr>
<tr>
<td>1300</td>
<td>0.42</td>
<td>0.08</td>
</tr>
</tbody>
</table>

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