

# Adhesion in Nanoimprint Lithography & Embossing Related Thin-Films

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## Abstract:

One of the critical issues during nanoimprint lithography (NIL) and the embossing process is the generation of defects in the mold due to the repeated removal of the mold/stamp from the polymer film. The goal of this project was to characterize the adhesion of the salient polymer-mold interfaces. The effects of interface chemistry and the geometry of patterned structures were investigated. Model polyacrylates were cured on silica surfaces with and without a fluorine-based release layer and were separated using a common fracture mechanics technique to examine the effects of interface chemistry on adhesion. The release layer reduced adhesion compared to the samples containing no release and ensured that the fracture path remained at the polymer-silica interface. Lithographically patterned arrays of Si lines coated with Teflon® were fabricated to examine the effects of mold pattern geometry on adhesion energies. Increasing aspect ratios of patterned arrays will be demonstrated to yield higher fracture energies, while crack growth orientation will be revealed to have little effect on adhesion.

## Introduction:

Embossing and NIL are promising new ways to create micro- and nano-scale features, respectively. Both processes involve pressing a mold or stamp into a polymer film, curing the film while the two are in contact, and then removing the mold. As molds are used many thousands of times, repeatedly overcoming the adhesion between the mold and polymer film often damages the mold. The first step to overcoming this problem and reducing damage to the mold is to characterize the adhesion of salient interfaces. In this study, the effects of interface chemistry on adhesion between NIL-related thin-films (with and without a release layer) as well as the effects of geometry on adhesion using embossing-related patterned molds and a relevant thin-film are investigated.

## Procedure:

Double cantilever beam (DCB) fracture specimens were fabricated for testing with cross-sections as shown in Figure 1. Two sets of the interface chemistry samples were prepared: one set with a release layer (Cytonix 4500) and one without. The polyacrylate was made up of: 92% lauryl acrylate, 5% 1,3 butanediol diacrylate,

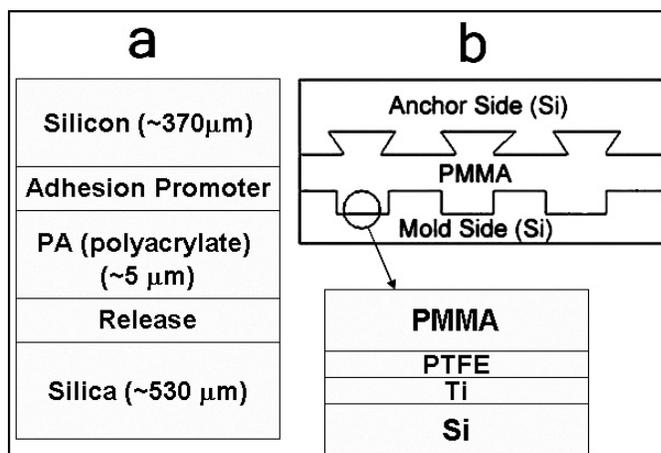


Figure 1: a) Cross-section of interface chemistry samples (release optional). b) Cross-section of geometry samples.

and 3% Ciba Darocur 4265. Half of the embossing samples with patterned structures were cut parallel to the features, and half were cut perpendicular. Testing was conducted using a micromechanical test system, which concurrently measures the load (the force applied) and the displacement (the distance between two cantilever beams). The samples were loaded and unloaded multiple times, and the slopes of the linear regions where the sample was unloaded were used to calculate the crack length, while the critical load (the beginning of the non-linear region) along with the crack length was used to calculate  $G$ , the interface fracture energy, in  $\text{J/m}^2$ . Then  $G$  vs. the crack length was graphed. X-ray photoemission spectroscopy was used to determine along which interfaces the samples debonded.

## Results and Conclusions

### Interface Chemistry:

The desired result was for the samples to debond at the polyacrylate-silica interface, which would indicate that the release layer was not functional and there was no added benefit to the release layer. The samples examined in this study debonded for the most part at the desired polyacrylate-silica interface (Figure 2),

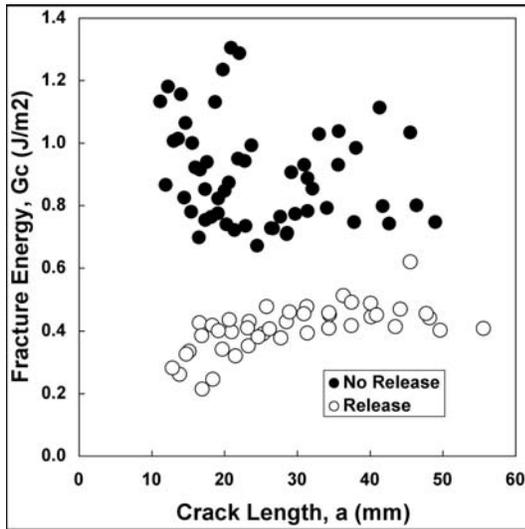


Figure 2: Release layer samples had lower adhesion than no release samples.

and the G values of the samples with the release layer were significantly less than those with no release layer (roughly in the range of 0.7-1.2 J/m<sup>2</sup> compared to 0.3-0.5 J/m<sup>2</sup>).

### Geometry:

Figure 3 compares the fracture energies measured for the 6.5 μm, 13 μm, and 18 μm tall structures where crack growth was propagated parallel to the lines. The adhesion for the 6.5 μm samples was less than those for the 13 μm, which were less than those for the 18 μm samples. This was expected because the samples with greater structure depths have more surface area in contact with the PMMA. This means that removing molds with deep structures from PMMA will cause more wear and tear than molds with shallow structures.

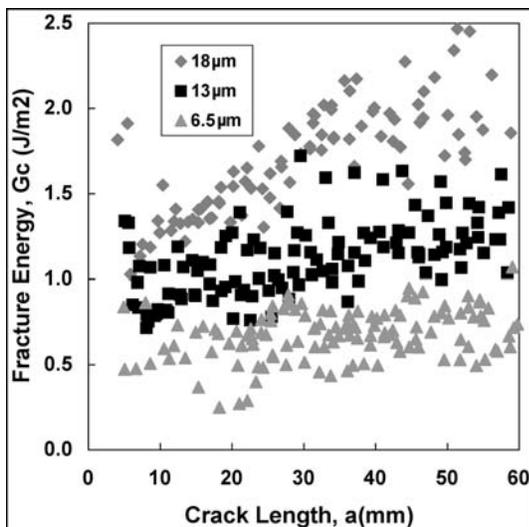


Figure 3: Fracture energy increases with aspect ratio.

Figure 4 compares how the orientation of crack growth affects the fracture energies of the 18 μm lines where cracks were propagated parallel as well as perpendicular to the lines. The majority of the G values of both types of samples are approximately within the same range (1.25-2.25 J/m<sup>2</sup>). The 13 μm perpendicular samples' G values were much more erratic than the G values for the 13 μm parallel, and the 6.5 μm perpendicular samples had G values significantly lower than its parallel counterparts. In conclusion, the direction at which the mold is separated from the PMMA is trivial at this length scale, although the data is not entirely conclusive.

### Acknowledgements:

Thanks to Professor Reinhold H. Dauskardt and Dr. Eric P. Guyer, WonHyoung Ryu and Dr. Rainer Fasching for the embossing samples, Dr. Frances Houle at IBM for the NIL samples, Dr. Michael Deal and Stanford University, the National Nanotechnology Infrastructure Network Research Experience for Undergraduates Program, and the National Science Foundation for funding.

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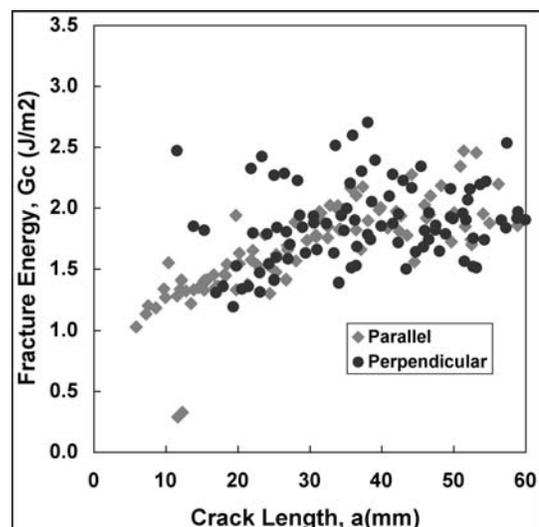


Figure 4: Fracture energies for the parallel and perpendicular samples are commensurate.