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High-Cycle Fatigue of Polycrystalline Silicon Thin Films in Laboratory Air

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ABSTRACT

When subjected to alternating stresses, most materials degrade, e.g., suffer premature failure, due to a phenomenon known as *fatigue*. It is generally accepted that in brittle materials, such as ceramics, cyclic fatigue can only take place where there is some degree of toughening, implying that premature fatigue failure would not be expected in polycrystalline silicon where such toughening is absent. However, the fatigue failure of polysilicon is reported in the present work, based on tests on thirteen thin-film (2 μm thick) specimens cycled to failure in laboratory air (~25ºC, 30-50% relative humidity), where damage accumulation and failure of the notched cantilever beams were monitored electrically during the test. Specimen lives ranged from about 10 seconds to 34 days (5×10^5 to 1×10^{11} cycles) with the stress amplitude at failure being reduced to ~50% of the low-cycle strength for lives in excess of 10^9 cycles.

INTRODUCTION

Bolstered by the success of sensor applications, manufacturers of microelectromechanical systems (MEMS) are developing micromechanical components made of silicon-based structural films in actuator, power and other “safety-critical” and “high performance” applications. However, these safety-critical structures are often subjected to aggressive mechanical and chemical environments without sufficient understanding of the behavior of the material under such conditions; this is especially pertinent as the dimensions of the material components may be far smaller, i.e., micron-scale and below, than has been conventionally tested in mechanical property evaluations. Consequently, in order to ensure performance and reliability, design approaches must be employed that account for the time and cycle-dependent degradation of the material at the size-scales of interest.

Micromechanical components are routinely subjected to cyclic stresses at kilo- and mega-hertz frequencies, accumulating large numbers of stress cycles in relatively short periods of time. It is important to note that when cyclic stresses are applied, most materials degrade and can suffer premature failure due to the process of *fatigue*. Fatigue is the most commonly experienced form of structural failure, yet surprisingly is one of the least understood. The most well known form of fatigue, that of the cyclic fatigue of metals, is generally associated with the generation and motion of dislocations and the accumulation of plastic deformation; these processes can lead to the creation and advancement of a nucleated or pre-existing crack by alternately blunting and sharpening the crack tip (striation formation). However, the corresponding mechanisms of fatigue in brittle materials, such as the structural films commonly used in MEMS, are quite different. Due to their high Peierls forces, brittle materials such as ceramics and single crystal silicon have very limited dislocation mobility at low homologous temperature, making the possibility of cyclic fatigue failure far less obvious. However, premature cyclic fatigue can occur in brittle materials, e.g., polycrystalline ceramics and ordered intermetallics, but by a conceptually different mechanism.[1,2] Such failures are generally associated with kinematically irreversible deformations; specifically, crack-tip shielding mechanisms that operate primarily in the crack wake and are the basis of their fracture toughness tend to degrade under cyclic loading conditions. However, this mechanism can only take place in toughened solids where there is some degree of shielding to degrade; it would not be expected in materials such as silicon. Despite this, studies of both mono and polycrystalline silicon thin films have established that crack growth can occur in ambient air environments under cyclic loading conditions.[3,4]

Considerations of fatigue-crack initiation and growth must be a crucial part of critical component design, since the majority of the lifetime of a structural component will invariably be spent in the initiation and early growth of small flaws. As most mechanical components contain notches, the process of fatigue crack initiation and small crack behavior in the presence of notches is a critical feature to be understood. Previous studies of polycrystalline silicon by two of the authors have shown that delayed failure can occur under cyclic stresses [3-6], results which have recently been confirmed by others.[7] The objective of the present study is to investigate the susceptibility of polycrystalline silicon to the growth of flaws under cyclic loading in the vicinity of a lithographically patterned, micromachined notch.

EXPERIMENTAL METHODS

Test Techniques

Over the past decade, we have developed a specimen geometry and "test structure" for characterization of fatigue crack initiation in thin films and have explored fatigue crack initiation in a variety of materials systems. The "structure" design is based on the philosophy that underlies fatigue testing standards such as ASTM E 466. The micron-scale fatigue characterization structure shown in Figure 1 is approximately 300 μm square. This structure is analogous to a specimen, electromechanical load frame, and capacitive displacement transducer found in a conventional mechanical testing system. The specimen is a notched cantilever beam that is in turn attached to a large, perforated plate that serves as a resonant mass. The mass and beam are electrostatically forced to resonate and the resulting motion is measured capacitively. On opposite sides of the resonant mass are interdigitated "fingers" commonly referred to as "comb drives." One side of these drives is for electrostatic actuation; the other side provides capacitive sensing of motion. The specimen is attached to an electrical ground, and a sinusoidal voltage at the appropriate frequency is applied to one comb drive, thereby inducing a resonant response in the plane of the figure. The opposing comb drive is attached to a constant potential difference, and the relative motion of the grounded and biased fingers induces a current proportional to the amplitude of motion. The small induced current is converted to a direct current (DC) voltage using an analog circuit. A variety of circuits for capacitive sensing can be found in the literature.[8] The output of the circuit used in this study was calibrated using the computer microvision system developed by Freeman [9] and stresses at the notch were calculated using the commercial finite element analysis package ANSYS. A stress concentration is introduced in the beam through the mask set, as shown in Figure 1. The radius of the stress concentration and the remaining beam ligament were selected to ensure that the specimen could be broken immediately at resonance. The longer-term fatigue response can then be measured by exciting the specimen at some fraction of the short time breaking amplitude. All samples were tested until failure occurred by fracture of the beam at the notch as shown in Figure 2.

The resonant frequency is used to monitor the accumulation of fatigue damage in the specimen until failure occurs at the notch. The device is driven at resonance using the following control scheme: The first mode resonant response of the specimen is determined by sweeping a range of frequencies around the expected response and monitoring the amplitude of response.

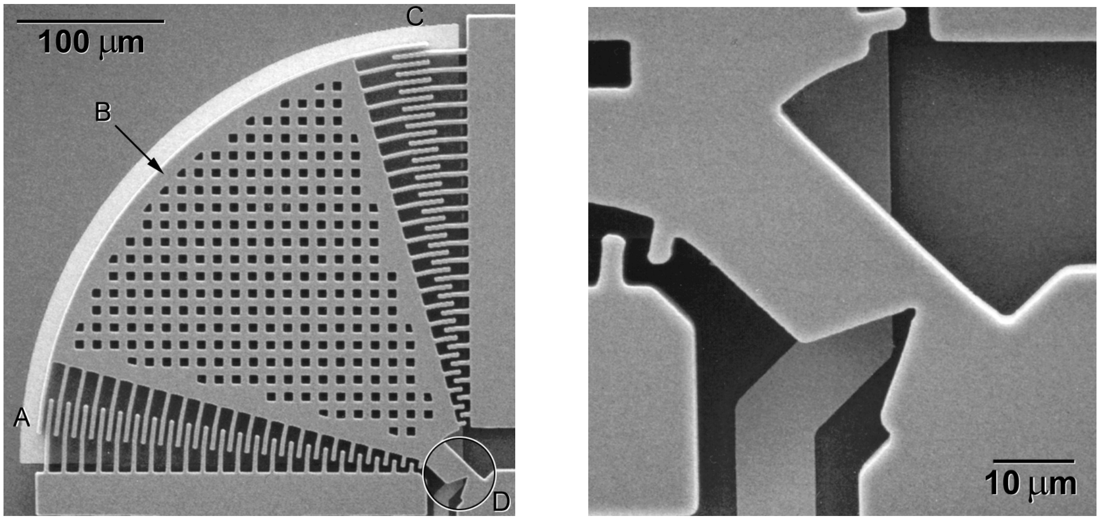

Figure 1 *Scanning electron micrographs of the stress-life fatigue characterization structure. The electrostatic comb drive actuator (A), resonant mass (B), capacitive displacement transducer comb (C), and notched cantilever beam specimen (D) are shown in an overview on the left. A detail of the notched beam is shown on the right.*

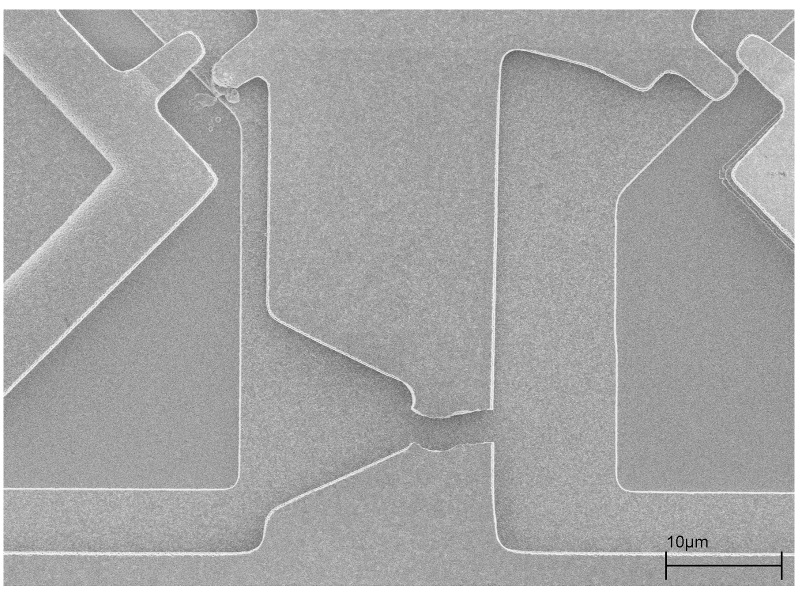

Figure 2 *Notch region of a failed polycrystalline silicon fatigue specimen.*

The peak amplitude is selected by fitting a second-order polynomial to the peak and extracting the maximum. The specimen is then excited at the peak frequency at a defined excitation voltage for a period of time. The frequency response is then again evaluated by sweeping around the excitation frequency. Over time this permits measuring any change in resonant frequency and consequently any change in the mechanical response of the specimen. This scheme is simple yet effective in detecting the compliance changes associated with crack growth and damage accumulation.

Materials and Test Samples

In this study, *n*-type low pressure chemical vapor deposited (LPCVD) polycrystalline silicon samples were micromachined using MCNC/Chronos' MUMPs process. The notched beams in this study have been designed for testing at approximately 40 kHz and are approximately 40 μm long, 19.5 μm wide, and 2 μm thick. The notch was located 9.8 μm from the base and was 13 μm deep. The notch root radius was approximately 1 μm and represents the smallest root radius that could be created given the processing conditions. The samples were prepared for testing by first removing the sacrificial oxide layer in 49% hydrofluoric acid for 3 minutes. After drying at 110ºC in air, the samples were mounted in ceramic electronic packages for testing.

RESULTS AND DISCUSSION

Thirteen thin film (2 μm thick) polycrystalline silicon specimens were tested to failure in laboratory air (nominally 25ºC, 30 to 50% relative humidity (RH)) using the sample geometry and testing procedure described above. Specimens were allowed to resonate for 1 to 5 minutes followed by recharacterization of the resonant frequency. The notched beams were driven in the in-plane bending resonant mode with a sinusoidal wave form with no DC offset. These conditions generated fully reversed stresses at the notch, i.e., a load ratio (minimum load by maximum load) of R = -1. Changes in resonant frequency were monitored by sweeping the drive frequency and recording the amplitude response using the previously described procedure. The "fatigue life" test duration ranged from about 10 seconds to 34 days, or 5×10^5 to 1×10^{11} cycles before failure over stress amplitudes ranging from approximately 1.8 to 3.7 GPa. The stress amplitude was controlled to better than 1% accuracy. Upon completion of the test, the measured lifetimes and the compliance change were evaluated.

The thin film polycrystalline silicon cantilever beams exhibited a time-delayed failure under fully reversed, cyclic stresses in relatively moist, room temperature air. The life of the notched beams as a function of fully reversed stress amplitude is shown in Figure 3. The peak tensile stress in the short life tests is consistent with the single cycle, or ultimate, strength typically measured in polycrystalline silicon.[10] Typically, delayed failures in brittle materials in stress-life tests will be clustered around this stress level. The life of the thin film polycrystalline silicon, however, increased monotonically with decreasing stress amplitude with a 50% reduction in stress amplitude resulting in an increase in life of approximately five orders of magnitude. This trend has not been observed in the fatigue stress-life testing of any bulk brittle materials (e.g., ref [2]); instead, a range of lives close to the strength of the material is observed, with few delayed failures at lower stress amplitudes.

The resonant frequency of the cantilever beam was used to monitor the specimen during cyclic loading and was observed to decrease monotonically before the specimen finally failed from the notch. This strongly suggests that the failure of the thin film polycrystalline silicon occurs after progressive accumulation of damage, e.g., by the stable propagation of a crack. This manifests itself as a progressive decay in the stiffness (resonant frequency) of the cantilever beam. The longer the life of the specimen, the larger the decrease in beam stiffness (Figure 4), which is again consistent with the notion of the accumulation of damage. However, it has proved to be difficult to correlate the growth of flaws observed on the fracture surfaces with such compliance changes.

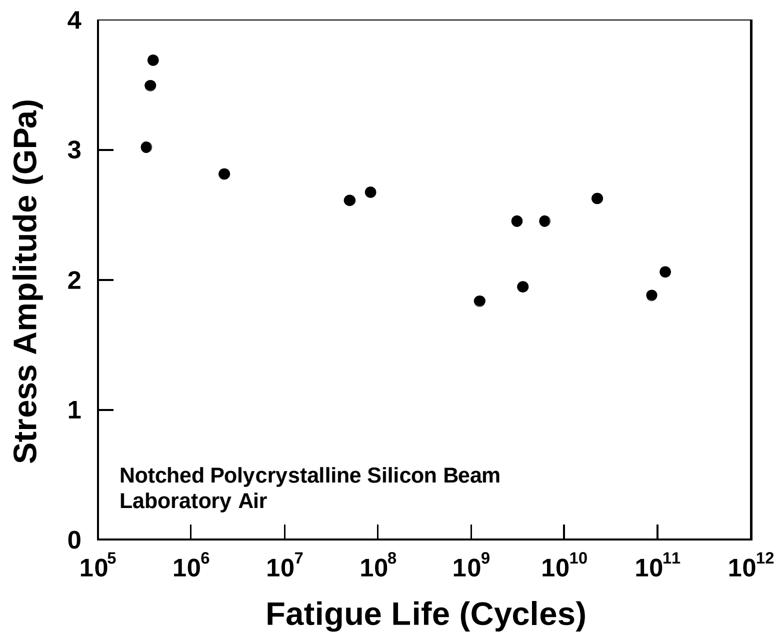

Figure 3 *Stress-life (S/N) fatigue curve of thin film polycrystalline silicon in laboratory air (nominally 25ºC, 30 to 50% relative humidity). The notched cantilever beams were tested at a frequency of nominally 40 kHz.*

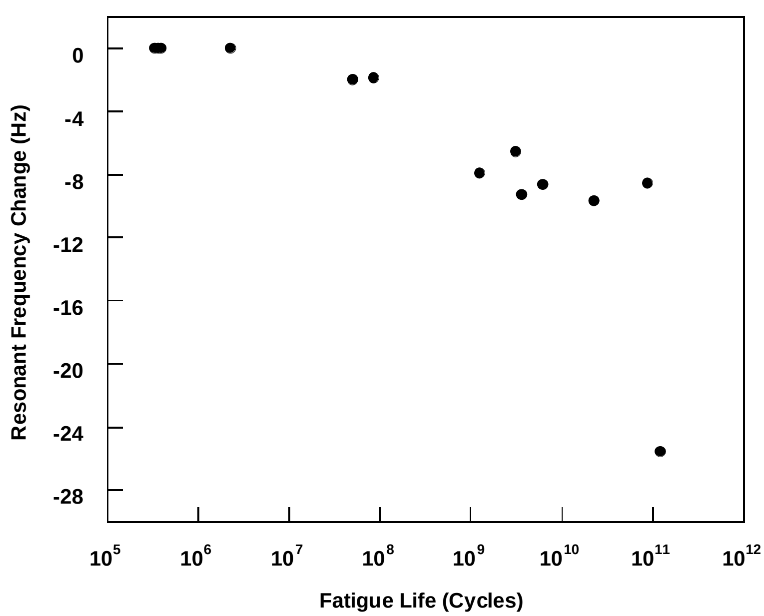

Figure 4 *Resonant frequency change as a function of specimen life for thin polycrystalline silicon films cycled in laboratory air (nominally 25ºC, 30 to 50% relative humidity).*

CONCLUSIONS

Based on this experimental study of the cyclic fatigue of thin (~2 μm) films of polycrystalline silicon at high frequencies (40 kHz) in laboratory air, two conclusions can be made. First, the polysilicon thin films can degrade and fail under cyclic loading conditions in moist ambient air at cyclic stresses some 50% of the single-cycle fracture strength. Second, the delayed failure of thin silicon films at stresses much lower than their fracture strength may be a significant limitation to the long-term frequency stability, reliability and life of polycrystalline silicon micromechanical devices. Even if a structure made from silicon does not fail, the loss of stiffness may lead to a large accumulated error that may detrimentally affect components that rely on a constant stiffness for timing or inertial measurement (i.e. an oscillator or gyroscope).

ACKNOWLEDGEMENTS

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REFERENCES

1. R.O. Ritchie, *Int. J. Fract.*, **100,** p. 55-83, (1999).
2. R.O. Ritchie and R.H. Dauskardt, *J. Ceram. Soc. Jap.*, **99,** p. 1047-62 (1991).
3. J.A. Connally and S.B. Brown, *Science*, **256,** p. 1537-9, 5063 (1992).
4. W.W. Van Arsdell and S.B. Brown, *J. MEMS*, **8,** p. 319-27 (1999).
5. S.B. Brown, W. Van Arsdell, and C.L. Muhlstein in *Proceedings of International Solid State Sensors and Actuators Conference (Transducers '97),* **1**, p. 591-3 (1997).
6. C. Muhlstein and S. Brown in *Tribology issues and opportunities in MEMS : proceedings of the NSF/AFOSR/ASME Workshop on Tribology Issues and Opportunities in MEMS*, p. 80 (1997).
7. H. Kahn, R. Ballarini, R.L. Mullen, and A.H. Heuer, *Proc. Roy. Soc. A*, **455,** p. 3807-23 (1999).
8. L.K. Baxter, *Capacitive sensors: design and applications* (IEEE, New York, 1997).
9. D.M. Freeman, A.J. Aranyosi, M.J. Gordon, and S.S. Hong, in *Solid-State Sensor and Actuator Workshop Technical Digest Solid-State Sensor and Actuator Workshop,* p. 150-5 (1998).
10. W.N.Sharpe, S. Brown, G.C. Johnson, and W. Knauss in *Microelectromechanical Systems for Materials Research*, edited by S. Brown, J. Gilbert, H. Guckel, R. Howe, G. Johnson, P. Krulevitch, and C. Muhlstein, (Mater. Res. Soc. Proc. **518**, 1998) p. 57-65.