Micromechanical Investigation of MEMS-Based Short-Wave Infrared Tunable Fabry-Pérot Filters

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Dean of Engineering  
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Dear Sir,

It is with pleasure that I present to you this dissertation entitled “Micromechanical Investigation of MEMS-Based Short-Wave Infrared Tunable Fabry-Pérot Filters” as required for a Doctor of Philosophy degree.

Yours sincerely,

Byron Alan Walmsley
Abstract

This study investigates the mechanical and physical properties of low-temperature (100-300 °C) plasma enhanced chemical vapour deposited (PECVD) silicon nitride (SiN$_x$H$_y$) thin films for the fabrication of short-wave infrared tunable Fabry-Pérot filters with high fill factor, high cavity finesse and low actuation voltages. It has been the intentions of this work to fabricate a tunable filter that can be monolithically integrated with temperature-sensitive substrates, namely mercury cadmium telluride (Hg$_{(1-x)}$Cd$_x$Te) photoconductors and photodiodes.

A range of methods have been utilised to determine the Young’s modulus ($E$), residual stress ($\sigma_0$), density ($\rho$) and Poisson’s ratio ($\nu$) of PECVD SiN$_x$H$_y$ thin films. In order to understand how $E$, $\sigma_0$, $\rho$ and $\nu$ are affected by process conditions, a range of SiN$_x$H$_y$ thin films deposited with varying chuck temperatures, RF powers and chamber pressures were measured. The resonance method was used to determine $E$ and $\nu$ of SiN$_x$H$_y$ thin films deposited under varying process conditions. The resonance method involves exciting the bending and torsional vibration modes of a microcantilever beam fabricated from a film. The $E$ and $G$ values can be extracted directly from the bending and torsional vibration modes and the $\nu$ value can be determined from the calculated $E$ and $G$ values. The density of the films was determined using the quartz crystal microbalance method. In order to determine the validity of the resonance method, finite element modelling was used to determine its dependence on microcantilever beam dimensions. The microcantilever beam bending method was also utilised to determine $E$ of SiN$_x$H$_y$ thin films deposited at various chuck temperatures. This method uses a nanoindenter to apply a point load to the beams whilst measuring its restoring force. The Young’s modulus of the beam can be determined from the applied load versus displacement curve if the dimensions of the beam are known. The validity of the Young’s modulus measurements was determined from nanoindentation experiments.

In order to determine the validity of the Poisson’s ratio measurements using the resonance method, a modified double-membrane bulge test was used to determine $\nu$ of low-tensile stressed SiN$_x$H$_y$ thin films. This test method utilises a square membrane and a large aspect
ratio rectangular membrane that is fabricated alongside from the same thin film. The Poisson’s ratio is determined from the ratio of the bulge deflections of the two membranes under an applied pressure. The method is suitable for determining ν of either stress-free thin films or those containing low tensile residual stresses. This method also provided E and σ₀ values of the films.

This study reveals that the resonance method, the microcantilever beam bending method and the bulge test provide similar E values compared to nanoindentation for the SiNₓHᵧ thin films. This study has also shown that σ₀ values obtained using the bulge test agrees with the substrate curvature method. The results show that all parameters measured are highly dependent on process conditions. In general, E, ρ and ν of the films all increase with increasing deposition temperature. Increasing the temperature also increases the tensile residual stress of the films. This study also reveals that increasing the RF power and decreasing the chamber pressure increases E and ρ, as well as increasing the compressive residual stress of the films.

The theoretical design and analysis, as well as the fabrication of a new surface micro-machined short-wave infrared tunable Fabry-Pérot filter for adaptive infrared photon detectors is also presented in this study. The proposed structure, termed the suspension filter, uses broad spectral range, high reflectivity distributed Bragg reflector (DBR) mirrors, resulting in very high finesse filters. The device utilises multiple sacrificial layers to define the resonant cavity spacer and the separation of the top mirror from the supporting flexures. The flexures were fabricated from low-temperature (PECVD) SiNₓHᵧ thin films. Separation of the top mirror from the supporting flexures allows for improved fill-factors (up to 79%), as well as increased tuning range. Theoretical optical and electro-mechanical results shows large wavelength tuning ranges (1.5-2.5 µm) at low actuation voltages (<30 V) are possible using the proposed design, whilst still maintaining a high cavity finesse. Optical characterisation of fixed filter micro-cavities on Si substrates show transmissions of ~60% with small line widths (35 nm) are achievable using the proposed mirror system. Mirror displacement versus applied bias voltage curves obtained from a released filter fabricated on Si show a stable mirror displacement of 620 nm was achieved, whilst theoretical results suggested the required 750 nm mirror displacement is possible using the proposed design.
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Chapter 1

Introduction

State-of-the-art infrared (IR) focal plane array (FPA) technologies with broad-band multi-colour capabilities have received increased interest in recent years for military applications, as well as for applications in the food, agricultural, mining and biomedical industries. Examples include high-resolution night vision, food spoilage identification and filtering, gas leak detection and skin cancer detection. For a system to function efficiently in these applications, it would need to be portable, robust and provide quantitative spectral information with high spatial resolution in real-time. A device meeting these criteria would be in high demand and with recent advances in semiconductor technologies, fabrication of such a device is made possible.

Currently, the highest performing photon detectors operating in the 1-12 μm IR spectral range are based on utilising mercury cadmium telluride (Hg$_{1-x}$Cd$_x$Te) photoconductors and photodiodes [1]. Compared to other detector technologies, only Hg$_{1-x}$Cd$_x$Te offers the ability to operate across the short-wave (1-3 μm), mid-wave (3-5 μm) and long-wave (8-12 μm) IR spectral ranges with extremely high detectivity. This is achieved by tuning its bandgap by changing the mole fraction of $x$ [2]. Commercially available Hg$_{1-x}$Cd$_x$Te IR detectors are sensitive to a broad spectral band, without the ability to discriminate between wavelengths of detected photons in real-time [3]. Today’s detectors also tend to operate in scanning mode, and suffer with time delay associated with data processing. A device that is able to detect only narrow spectral bands and could be tuned to a desired wavelength would greatly reduce the amount of data to be processed, thus allowing real-time video display.
At present, IR spectral information is typically obtained using grating or Fourier transform spectrometers, which are bulky and expensive [3]. It has been proposed that a cheaper alternative can be fabricated whilst matching the performance and resolution of the commercially available spectrometers [4, 5]. The device is fabricated using micro-electro-mechanical systems (MEMS)-based technologies, and utilises a tunable Fabry-Pérot (FP) optical filter, which is monolithically integrated with an Hg1-xCd,Te detector. A description of the proposed micro-spectrometer is given in the following sections of this chapter including a short introduction of MEMS concepts and current MEMS devices commercially available.

1.1. Micro-Electro-Mechanical Systems

For years scientists have realised the potential of miniaturisation for the fabrication of sensors and other devices. Successful development of products that are small in size and can be reproduced identically in large volumes will lead to lower production costs whilst improving reliability and function. As the inertia of an object is proportional to its size, decreasing the size of an object decreases its sensitivity to mechanical vibrations. Also, objects that are small require less energy to manipulate their movement. Consequently, there has been growing interest in recent years in systems that integrate electronics with movable parts on a micron scale for the fabrication of portable devices. Systems that utilise both electronics and mechanics on such a scale are known as micro-electro-mechanical systems, or MEMS. In practice, MEMS involve both electronic and non-electronic elements, and can perform functions such as sensing, signal processing, actuation, display and control [6]. In general, MEMS devices are light in weight, portable and efficient in terms of power consumption.

Most MEMS devices and systems involve some form of photolithographic-based microfabrication process [6]. The basic fabrication processes are similar to those used in the microelectronics industry, including batch fabrication methods, which offer the potential to manufacture devices and systems in high volume, and thus greatly reduce production costs [6]. Consequently, it makes sense to combine both MEMS and microelectronic parts together to produce a fully integrated microsystem. An example of a microsystem is the surface-micromachined integrated silicon accelerometer. Figure 1.1 illustrates an accelerometer manufactured by Analog Devices, Inc. The device comprises of a MEMS sensor at the centre of a chip, which is surrounded by extensive
complementary metal–oxide–semiconductor (CMOS) signal processing electronics. Fabricating a monolithic device such as the one shown in Fig. 1.1 is feasible provided the steps involved in fabricating the MEMS are compatible with the microelectronics. Alternatively, the microsystem can be partitioned into subsystems that are fabricated separately and then assembled into a compact system during packaging [6]. This method however can cause complications with the design, which can limit the performance of the device.

MEMS fabrication processes include specialised ‘micromachining’ techniques, which allow for structures to be released from underlying layers. Once released, these structures are free to be moved and manipulated by external forces applied through electromagnetics, surface tension, chemical reactions, pressure differentials and inertial forces. The types of materials used to fabricate MEMS depend on a number of constraints, however metals, ceramics and polymers can all be utilised. In general, ceramics such as silicon nitride and silicon oxide, as well as single crystal Si, amorphous Si and poly-Si are commonly used as structural materials. This is mainly due to their superior machining capabilities, excellent mechanical properties and their chemical stability. Metals are commonly used as conductors, as well as reflective coatings. Polymers are less common as a structural material in MEMS, mainly due to their incompatibility with high temperature processing steps. MEMS are usually fabricated

FIG. 1.1. A surface-micromachined integrated Si accelerometer [7].
from materials in thin-film form and three-dimensional structures are generally constructed from multiple layers of two-dimension features [6].

Figure 1.2 illustrates a close-up view of the inertial sensor used in a MEMS accelerometer, similar to the one shown in Fig. 1.1. The device works by detecting the differential lateral capacitance between the interleaved fingers of the comb structure when the central mass is displaced during periods of acceleration. In order to increase the sensitivity of the device, large efforts are made to increase the mass of the inertial sensor. Due to their small size and low cost however, MEMS accelerometers are commonly used as air-bag sensors in automobiles. The highly planar surfaces are a common feature of MEMS, as well as the small distances between the structures and the substrate. Surface micromachined accelerometers are now a mainstream technology that ship in millions of units for automobile and consumer applications [7].

Micro-opto-electro-mechanical systems (MOEMS), also known as optical-MEMS, are a subset of MEMS which include optical components such as wave guides, diffraction gratings and mirrors [8, 9]. MOEMS commonly use thin-film materials with optical properties that are conducive to reflecting, diffracting or refracting light in one, two or

FIG. 1.2. Scanning electron micrograph (SEM) of a surface-micromachined integrated Si accelerometer [7].
three-dimensions. An example of a MOEMS device is the Texas Instrument’s Digital Micromirror Device (DMD) shown in Figure 1.3 [10]. A DMD chip is fabricated monolithically using CMOS compatible processes and consists of several hundred thousand 16 µm square mirrors arranged in a rectangular array. Each mirror represents a pixel of the image to be displayed, and can be individually rotated ±10°, representing an on or off state. Incident light from a source is focused onto the DMD chip, which reflects light onto a screen via projection optics. In the ‘on’ position, a pixel will appear bright on the screen. In the “off” position, the light is directed elsewhere (usually onto a heat sink),

![Diagram of DMD pixels](image1.png)

(a)

![Scanning electron micrograph of DMD pixels](image2.png)

(b)

FIG. 1.3. (a) Schematic diagram of two DMD pixels, and (b) scanning electron micrograph of an array of DMD pixels [10].
making the pixel appear dark. A grey-scale image is produced by rapidly changing the position of the pixel between the on and off states repeatedly, which in turn will change the intensity of light reflected onto the screen. Colour is added to the image using colour filters. The top mirror surface of each pixel is fabricated from thin film aluminium and each are mounted onto a yoke, which is connected to a compliant torsional hinge via support posts. Electrodes on either side of the hinge control the position of the mirror by electrostatic attraction. Due to their small size, only a few volts are required to manipulate the position of the mirrors with a very high degree of accuracy.

There has been tremendous growth in recent years in the MEMS industry. As shown in Fig. 1.4, worldwide revenue for MEMS is set to double in 2007 compared to revenue in 2002 [11]. Some areas of MEMS which are experiencing major growth over the last 10 years include pressure measurement devices (for use in the automotive, medical and mining industries), optics (for example, optical switches and displays) and RF devices (for example, cell phone component devices and radar) [6]. Though few will argue that there is potential demand for MEMS applications over a range of industries, making the transition from concept to high-volume production is expensive and risky [12]. Many expect that once the industry matures, it will become easier for companies to
grow and develop. There is a need, however for a fundamental core technology, which can be applied over a range of MEMS applications. Once this is achieved, initial investment costs will decrease, whilst increasing a product’s reliability.

The fabrication of MEMS devices traditionally involve high-temperature silicon-based processes (>500°C). Some advantages of high temperature processing include:

- The processes are well known and repeatable;
- The deposited films are of a very high quality and environmentally stable;
- The mechanical and physical properties of high-temperature deposited films are well understood and easily tuned;
- The high temperatures can create a cleaner environment, which results in the removal contaminants and most importantly moisture [13].

For a MEMS-based IR tunable FP filter monolithically integrated with an Hg$_{1-x}$Cd$_x$Te detector to be realised, low-temperature processes need to be utilised. This is mainly due to the degradation of Hg$_{1-x}$Cd$_x$Te at temperatures above 125 °C [3-5]. The maximum process temperature Hg$_{1-x}$Cd$_x$Te can withstand is ~150 °C, however the material can only be held at this temperature for a short period of time before Hg diffuses through the lattice. Low-temperature processing has its drawbacks, namely:

- The processes are not well known and, at times, unrepeatable;
- Unless deposited correctly, the films can be very poor in quality and environmentally unstable [14];
- The mechanical and physical properties of low-temperature deposited films are not well understood;
- Low-temperature processing can lead to contamination and issues with moisture removal.

Though some of the problems regarding low-temperature processing can be overcome (i.e. utilising ‘getter’ films that immobilise stray moisture), the technology is regarded as being unreliable and inferior to high-temperature processes. Therefore, development of reliable low-temperature thin film synthesis and processing techniques is of critical importance for the success of a range of MEMS-based technologies and more research in this area is required.
1.2. Principles of MEMS-based Tunable Infrared Micro-Spectrometer

1.2.1. Working principle of Fabry-Pérot interferometer

A typical Fabry-Pérot (FP) interferometer consists of two plates separated by a spacer to form an optical cavity. The plates are usually made of glass or quartz with planar surfaces [15-17]. Their inner surfaces are coated with partially transparent films that act as highly reflective mirrors, which are parallel to each other, as shown in Fig. 1.5. The cavity in this case has refractive index \( n \) and length \( d \). For the purposes of simplicity of discussion, the absorption of the mirrors is assumed zero. When a plane wave of light of unit amplitude and free-space wavelength \( \lambda_0 \) is incident at an angle \( \theta \) to the surface of one of the two mirrors, multiple reflections \( r \) and transmissions \( t \) occur, with progressively reduced amplitudes with increasing order of reflection and transmission, as indicated in the figure. The subscripts of the amplitudes denote the interface at which they occur (mirror 1 or 2) and the superscripts denote their direction (‘+’ denotes \( r \) and \( t \) from left-to-right, ‘-’ denotes \( r \) and \( t \) from right-to-left). The multiple

![FIG. 1.5. Reflection of a plane wave in an FP interferometer.](image-url)
reflections between the two mirrors create two types of waves; those that are reflected back towards the source \((a,b,c)\) and those that are transmitted through the cavity \((A,B,C)\) \[18\]. Between every pair of successive reflections or transmissions, e.g. between \(a\) and \(b\) or between \(B\) and \(C\), there is an optical delay of a phase difference determined by the cavity length \(d\). This phase lag \((\varphi)\) is given by [15-17]:

\[
\varphi = \frac{4\pi nd \cos \theta}{\lambda_0}
\]

If the first \(m\) transmitted beams are now superimposed, the amplitude of the resultant electric field vector of transmitted light is given by [16]:

\[
E_t(m) = t_1^* t_2 + t_1^* r_2^* r_1^* t_2^* e^{i\varphi} + t_1^* (r_2^* r_1^*)^2 t_2^* e^{2i\varphi} + \ldots
\]

\[
= t_1^* t_2^* [1 + r_1^* r_2^* e^{i\varphi} + (r_1^* r_2^*)^2 e^{2i\varphi} + \ldots + (r_1^* r_2^*)^{(m-1)} e^{i(m-1)\varphi}]
\]

As \(m \to \infty\):

\[
E_t = \frac{t_1^* t_2^*}{(1 - r_1^* r_2^* e^{i\varphi})}
\]

For the case where:

\[
\varphi = k\pi, \quad k = 1, 2, 3, \ldots
\]

all individual transmitted waves are in phase and \(E_t\) in Eqn. (1.3) is maximised. This is known as the resonant condition \[18\]. It is obvious that the satisfaction of the resonant condition is determined by the relative values of \(d\) and \(\lambda_0\) for any given \(\theta\).

When a plane wave consisting of several wavelengths of light is incident to an FP interferometer, only the wavelengths that satisfy Eqn. (1.4) will resonate, with all other wavelengths reflected. Those wavelengths that satisfy Eqn. (1.4) are known as the resonant wavelengths. A change in \(d\) will result in different wavelengths resonating inside the cavity. An FP interferometer can therefore be constructed with a certain cavity length and serve as a filter by selectively reflecting or transmitting particular wavelengths of light. More importantly, if an FP resonator can be constructed with an adjustable cavity length, it may be tuned to different wavelengths in service. By convention, resonator systems with fixed cavity lengths are referred to as Fabry-Pérot etalons, whilst resonator systems with variable cavity lengths are referred to as Fabry-Pérot interferometers [15].

For an FP interferometer to filter IR radiation, the length of the cavity needs to be of the same order as the wavelength. For example, cavity lengths for short wave IR (SWIR) interferometers are usually around 1 µm [18, 19]. In order to vary the cavity
length, FP interferometers were initially designed so that one plate remained fixed, while
the other was mounted on a screw-controlled carriage to allow continuous variation of the
plate separation [15]. However, difficulties in the mechanical construction, their
imprecision in changing the cavity length, as well as their sensitivity to mechanical
vibrations have caused the arrangement to become obsolete. Due to the recent advances
in MEMS technology, new interest has been focussed on the FP interferometer. FP
interferometers constructed using thin film materials are known as *Fabry-Pérot filters*.
The FP filter has many advantages over the older-style interferometers, including:

- Insensitivity to mechanical vibrations;
- The ability to vary the length of the cavity with high precision by electrostatically
  actuating one of the mirrors;
- Fabrication of arrays of identical FP filters.

The last point suggests that if an array of FP filters is integrated with photodetector
material, a micro-spectrometer with imaging capabilities is possible.

1.2.2. Technical challenges of a MEMS-based Fabry-Pérot filter

A generic cross-sectional view of a MEMS-based FP filter that is monolithically
integrated with a photodetector is shown in Fig. 1.6. The upper mirror is positioned on
top of self-supporting flexures and the bottom mirror is fabricated on the detector
material, forming a cavity in between. The cavity length may be adjusted by applying an
electrostatic load between the two mirrors, which deflects the supporting structure
downwards to bring the top mirror closer to the bottom mirror. This is usually achieved
by depositing conducting layers on the substrate as well as on the flexures. These layers
act as electrodes, while the flexures themselves act as the insulating layer. The restoring
force (i.e. the force acting against the electrostatic load) comes from the elasticity of the
supporting flexures. With this mechanism, the cavity length can be adjusted continuously,
thus allowing selective filtering of IR radiation of any wavelength within that range to the
photodetector sitting below. Recent attempts in fabricating such devices have been
carried out using \( \text{Hg}_{1-x}\text{Cd}_x\text{Te} \) IR photodetectors and surface micromachined monolithic
fabrication techniques [4, 5, 19]. The advantages of using this method are that it is
excellent for arrays in terms of pixel alignment and scalability. An alternative to using
monolithic fabrication techniques is the hybrid fabrication method where the detector
material and the FP filter are fabricated separately, and then aligned together afterwards. This has the advantage of separating high temperature MEMS processes from temperature sensitive substrates (e.g. Hg$_1$-x,Cd$_x$Te). Difficulties arise, however, with fabricating arrays in this manner due to issues with $x$-$y$ alignment and planarity.

For a surface micromachined tunable FP filter to operate effectively, the following main technical challenges exist:

1. Parallelism
2. Tuning range
3. Actuation voltage and tuning frequency
4. Fill factor

Parallelism

As introduced above, the wavelength of IR radiation filtered is directly determined by the cavity length. Therefore, flatness and parallelism of the two mirrors are of critical concern for the performance of the FP filter. Any curvature or tilt of the actuated mirror will adversely affect the uniformity of the cavity length [20, 21], thus the bandwidth and accuracy of the detector. Mirror distortion can be introduced during filter fabrication, as well as during actuation. Filter designs that incorporate the top mirror being deposited directly onto the supporting flexures are prone to mirror distortion during actuation. As the flexures themselves will inevitably deform during actuation, so too will the top

---

**FIG. 1.6. Cross-sectional view of an IR micro-spectrometer.**
mirror. Increasing the length of the support flexures helps to improve parallelism as the amount of curvature in the mirror decreases with increasing flexure length [18-23].

**Tuning range**

The tuning range of the filter needs to be maximised in order to increase the range of wavelengths that can be detected. When using electrostatic actuation, the largest theoretical top mirror displacement achievable is 1/3 the cavity length when cantilevers are employed as the supporting flexures [20-22]. Larger tuning ranges (theoretically, up to 2/3 the cavity length) are achievable if fixed-fixed beams are used to support the top mirror instead of cantilever beams. Ideally, stable top mirror displacement across the entire cavity length is desirable. Using electrostatic forces however, stable mirror control is difficult to achieve at large mirror displacements due to its non-linear nature and pull-in phenomenon [22].

**Actuation voltage and tuning frequency**

As mentioned above, electrostatic forces are commonly used to actuate the filter. This is mainly due to its ability to control the position of the top mirror with a high degree of accuracy at small mirror displacements, as well as its ability to actuate the filter at high frequencies. High frequencies are required to achieve a real-time imaging system. Ideally, low actuation voltages are required to allow the integration of this technology with application-specific integrated circuits (ASICs) needed for filter control and detector signal acquisition. Actuation voltages can be minimised by decreasing the stiffness of flexures. Current methods used to decrease the stiffness of the flexures include:

- Changing the flexure’s dimensions;
- Changing the flexure’s constraints;
- Altering the flexure’s material properties.

Increasing the length of the flexures decreases the voltage required to displace the mirror. Cantilever beams are less constrained compared to fixed-fixed beams and thus require smaller voltages for a given displacement. Fixed-fixed beams however are less constrained compared to membranes that are constrained completely around their perimeter. The stiffness of a flexure is directly proportional to its material properties. A decrease in the flexure’s Young’s modulus \((E)\) will result in a decrease in the flexure’s
stiffness. Depending on the constraint of the flexure, altering the flexure’s residual stress ($\sigma_0$) can also lead to a decrease in the flexure’s stiffness.

The material properties of a thin film material are directly dependent on the conditions of which it was processed. Thus in order to design, optimise and improve the performance of a MEMS-based tunable SWIR filter, it is of critical importance that the mechanical and physical properties of the material used to fabricate the flexures are well known. It is also important to gain an understanding of how these properties are affected by changes in process conditions. Other material properties that are of interest to the filter designer are the flexure density ($\rho$) and Poisson’s ratio ($\nu$). The density of a material is related to its mass and thus important vibration information can be obtained with its knowledge. The Poisson’s ratio relates a material’s Young’s modulus to its Shear modulus ($G$), and hence provides important information regarding how a material deforms elastically when subjected to internal and external loads.

**Fill factor**

For single-element tunable detectors, the space occupied by the MEMS supports and flexures is not a significant issue, however, for real-time hyperspectral imaging applications, in which each element is part of a large two-dimensional focal plane array, the space occupied by the supports and flexures can be a significant proportion of the total area. This space does not contribute to sensing and detecting. This is particularly true for small pixel sizes. High fill factor, defined as the ratio of optically active area to total pixel area, is critical and is required for high resolution systems, while small pixel sizes are required to reduce the cost of ancillary optics needed for imaging. The requirement for high fill factor is incompatible with traditional methods used to improve parallelism, tuning range and minimisation of actuation voltage (i.e. increasing flexure length).

### 1.2.3. Recent MEMS-based Fabry-Pérot filter designs

Figure 1.7 shows a SWIR tunable FP filter monolithically integrated with a Hg$_{1-x}$Cd$_x$Te detector, as fabricated by Dell et al. [23]. Figure 1.7(a) shows a scanning electron microscope (SEM) image of the filter and Fig 1.7(b) schematically illustrates a cross-section of the device. Distributed Bragg reflector (DBR) mirrors were utilised consisting of thermally evaporated germanium and silicon monoxide dielectric layers. The supporting flexures were fabricated in an X-style using a plasma-enhanced chemical
vapour deposited (PECVD) silicon nitride thin film deposited at 125 °C. The top and bottom electrodes were fabricated from thermally evaporated gold/chromium thin films. Partially cured polyimide (PI) was used as the sacrificial layer for the formation of the cavity. All fabrication steps involved processing temperatures at 125 °C or below, except for the polyimide thin film, which was partially cured at 130 °C for 20 minutes. The size of the top mirror is 100 µm and the length of the supporting flexures is also 100 µm. The target thickness for the PI sacrificial layer was ~1.4 µm and the filter was designed to tune between IR wavelengths 1.6-2.5 µm. Actuation of the filter using an applied bias voltage of up to 7.6 V resulted in an IR tuning range of 1.8-2.2 µm. Whereas successfully demonstrating the working principles of a tunable FP IR detector, the X-style flexure design is not appropriate for array applications as the fill factor is ~17%. For real-time hyperspectral imaging applications, fill factors of at least 56% are required.

The filter shown in Fig. 1.7 is a common design employed by FP filter designers [3-5, 18-20, 23-27]. This type of filter is generally designed so that most of the deformation will occur in the flexures during actuation. Consequently it is imperative that the mechanical and physical properties of the flexure material is characterised. It is
known that $E$ and $\sigma_0$ of low-temperature PECVD silicon nitride thin films are dependent on the deposition conditions [3, 4, 18, 28]. It is not as clear however how $\rho$ and $\nu$ of these films are affected by deposition conditions. The Young’s modulus and $\sigma_0$ have been determined previously for low-temperature PECVD silicon nitride thin films deposited under certain deposition conditions using nanoindentation and substrate curvature methods, respectively [3, 18]. The accuracy of these methods is however subject to criticism [3, 28-30], and more work in this area is required.

1.3. Thesis Objectives

The main objective of this work is to improve the performance of the surface micromachined monolithically fabricated SWIR tunable FP filter proposed by Dell et al. [23], both optically and mechanically. To achieve this objective, the following specific issues will be investigated:

1. To quantitatively characterise the mechanical and physical properties of thin film materials used to fabricate the filter, and to characterise the effects of deposition conditions on these properties. Specific attention will be paid to determining $E$, $\rho$, $\nu$ and $\sigma_0$ of PECVD silicon nitride thin films using a variety of methods in a bid to gain further understanding of how process conditions affect these properties.

2. Mechanical and optical design and simulation of a new generation SWIR tunable FP filter that can be monolithically integrated with Hg$_{1-x}$Cd$_x$Te photodetector material. The proposed device will have a fill factor of greater than 56% and be able to tune across the IR spectrum from 1.5-2.5 µm, whilst maintaining parallelism to within a certain tolerance. For the purposes of this study, this tolerance is defined as the variation in height across the moveable mirror, and must not exceed 10 nm (i.e. within the same order of magnitude as the surface roughness of the mirror materials).

3. To fabricate a prototype device to demonstrate the concept and the performance. The device must be able to achieve these requirements under electrostatic actuation of less than 30 V.
Chapter 1 - Introduction

1.4. List of Publications

1.4.1. Publications Resulting From This Work


1.4.2. Publications Related to This Work


Chapter 1 - Introduction

References


11. MEMS Industry group, [http://www.memsindustrygroup.org](http://www.memsindustrygroup.org) (Figure re-drawn).


Chapter 2

MEMS Fabrication Principles

In this chapter, the fundamentals of microfabrication are discussed with emphasis on processes and materials which are directly related to this work. In particular, processes used for the development of test structures needed to assess the material properties of low-temperature deposited materials, and processes specifically related to the integration of MEMS FP filter devices with Hg$_{1-x}$Cd$_x$Te detector structures. A brief introduction to lithography is provided along with pattern transfer processes that concentrate on physical and chemical vapour deposition, as well as wet and dry etching of thin film materials. The principles behind bulk and surface micromachining are discussed in detail, along with important issues associated with the fabrication of MEMS devices.

2.1. Lithography

Lithography is defined as “the technique used to transfer copies of a master pattern onto the surface of a solid material” [1]. The most common form of lithography is photolithography, which is used in the integrated circuit (IC) industry to transfer patterns from masks onto thin films. Photolithography utilises photo-sensitive materials, known as photoresists to create patterns on solid surfaces. Photoresists usually consist of a polymer base resin, a sensitizer and a casting solvent [1]. They are generally spun onto the surface of a sample and soft baked in order to drive off the solvent to form a thin film polymer material on the surface.

When certain soft baked photoresists are exposed to UV radiation, the polymer changes its structure. Positive photoresists contain diazoquinone ester as a sensitisr, and the photochemical reaction during exposure transforms the ester into a carboxylic acid,
which is soluble in basic developing solutions [1, 2]. In contrast, exposure of negative photoresists to UV light creates cross-linking in the polymer, which is less soluble in a developing solution. Higher resolution lithography systems utilise electron beam, ion beam or an x-ray source to pattern photoresists.

Contact printing is a common exposure method utilising UV as the exposure source and involves direct contact of a mask onto the surface of the baked photoresist. Examples of this method using both positive and negative photoresists are shown in Fig. 2.1. In this particular example, the photoresists are used to define the shape of a thin film, which is deposited onto a substrate. As shown in Figs. 2.1 (a) and (b), a thin film is first deposited onto the substrate (step (i)). Photoresist is then spun onto the film and soft baked to drive off solvents (step (ii)). In step (iii), a mask is positioned so that it is in contact with the sample to define the pattern. The masked sample is then exposed to UV, which starts the photochemical reaction in the photoresist. Depending on the photoresist used, a post-bake as well as a flood exposure of the resist layer may be required after initial exposure to either drive exposure reactions to completion or to initiate new reactions. Developing of the photoresists in step (iv) results in the more soluble areas of the photoresists to be washed away, leaving the desired pattern. Areas of the thin film underneath the photoresists are now protected from other fabrication processes, such as wet or dry etching (step (v)). Stripping of photoresists (step (vi)) using simple solvents (for example, acetone) or acids reveals the patterned thin film structure.

2.2. Pattern Transfer Processes

Pattern transfer processes in microfabrication are classified as being either additive or subtractive. Examples of additive and subtractive processes which are related to this work are discussed below.

2.2.1. Physical Vapour Deposition

Physical vapour deposition (PVD) is a thin film process that involves either evaporation or sputtering of a material onto a substrate. Examples of PVD systems include thermal evaporators, sputtering systems, molecular beam epitaxy (MBE), laser ablation, ion plating and cluster deposition systems. For the purposes of this work, only thermal evaporation systems will be discussed. Low-pressure reactors are employed in
FIG. 2.1. Photolithography and pattern transfer processes using (a) positive photoresist, and (b) negative photoresist.
these systems to minimise intermolecular collisions, which would otherwise affect the density and uniformity of the resulting deposited film.

Thermal evaporation systems are based on boiling or sublimating a source material onto a substrate [1]. Figure 2.2 illustrates a schematic of the thermal evaporation system used in this work. The source material is loaded into a boat or filament, which is made from a refractory metal (e.g. tungsten). The boat is heated via resistive heating and the resulting evaporant onto the substrate, which can either be heated or cooled via a temperature control stage. All surfaces that have a line-of-sight with the source material are coated. The process is highly directional, thus the films tend to be non-conformal in nature. The thickness of the deposited film is measured using a water-cooled quartz crystal thickness monitor. Materials such as Al, Cr, Au, Ge and SiO can be easily deposited in this manner.

FIG. 2.2. Schematic diagram of a thermal evaporation system.
Chapter 2 - MEMS Fabrication Principles

The boats or filaments used tend to be small in size, which can limit the thickness of the deposited film. Impurities in the boat or filament can easily contaminate the source material, which of course is undesirable. Consequently, thermal evaporation systems such as the one illustrated in Fig. 2.2 tend to only be found in laboratories [1]. For industrial applications, electron beam (e-beam) PVD systems are more common as there is less contamination of the source material. An e-beam system utilises a focussed, high-intensity electron beam gun, which locally melts the source material. The source material forms its own crucible, which is positioned in a water-cooled hearth. The temperature at the interface between the source material and the hearth is too low for chemical reactions to occur, which results in less contamination. Higher quality films and higher deposition rates are achieved using this method.

2.2.2. Chemical Vapour Deposition

As the name suggests, chemical vapour deposition (CVD) involves the chemical reaction between reactant gases in their vapour phase. The reactant gases are generally diluted with an inert carrier gas (e.g. nitrogen) and react and diffuse onto a substrate [1]. Two common examples of CVD systems are low-pressure chemical vapour deposition (LPCVD) and plasma-enhanced chemical vapour deposition (PECVD). With regards to this work, the major difference between the two deposition methods is the appropriate temperature required for film deposition. LPCVD systems require substrate temperatures typically above 500 °C to promote surface reactions. In contrast, PECVD systems utilise a radio frequency (RF) induced plasma to transfer energy into the reactant gases and thus promote surface reactions. Consequently, films can be deposited on substrates with temperatures below 300 °C.

Some advantages of using LPCVD systems to deposit thin films include excellent purity and film uniformity, conformal step coverage and large wafer capacity [1]. LPCVD can be used to deposit α-Si, poly-Si, Si₃N₄, phosphor-silicate glasses (PSG) and boro-silicate glass (BSG) thin films [3]. Silicon nitride films deposited using LPCVD processes have higher $E$ and $\rho$ values [4, 5] compared to PECVD silicon nitride films, and eventually become stoichiometrically equivalent to Si₃N₄ with increasing deposition temperature [6]. In contrast, PECVD films contain high percentages of hydrogen, and are more properly described as Si₃NₓHₙ [7]. The hydrogen content is process condition
dependent, with more hydrogen being incorporated as deposition temperature is decreased, which affects the chemical structure and properties of the film [8].

A schematic illustration of a parallel-plate PECVD system used to deposit silicon nitride is shown in Fig. 2.3. Wafer samples are placed on the anode, which is heated by a simple resistive heater element [3]. The reactant gases, silane (SiH₄), ammonia (NH₃) and nitrogen (N₂) are fed into the chamber through the cathode, where they are excited by electrons that are accelerated by the applied electric field. This results in complex plasma-phase reactions where both charged species (ions and electrons) and neutral species (radicals, atoms and molecules) are produced [3]. The charged species are transported to the substrate by ion acceleration whereas the neutral species are transported by diffusion [9]. The heated anode provides the necessary energy to initiate condensation and film growth from the particles created in the plasma [3]. By-products are removed from the chamber under the anode to enhance the laminar flow of the reactant gases.

![FIG. 2.3. Schematic diagram of a parallel plate PECVD reactor.](image)
Reactor parameters, such as chamber pressure, RF power, frequency of the RF excitation and deposition temperature, all influence the properties and induced stresses of a deposited film [1]. In general, a low chamber pressure, high RF power, low RF excitation frequency and high deposition temperature will improve the film’s physical quality, including thin film adhesion to the substrate, step coverage and lowered porosity, as well as the film’s mechanical properties. The stresses of the deposited film are also sensitively affected by changes in reactor parameters.

For the case where N\textsubscript{2} is used as the diluting gas, Claassen et al. [10] provides in detail the reactions which occur between SiH\textsubscript{4}, NH\textsubscript{3} and N\textsubscript{2} in a PECVD chamber. The reactions include (i) gas phase ionization, (ii) gas phase dissociation, (iii) gas phase reaction, (iv) condensation and film growth, and finally (v) hydrogen elimination by cross-linking:

(i) Gas phase ionization

\[
\text{SiH}_4 \xrightarrow{\text{energy}} \text{SiH}_n^+ + (4-n)\text{H} + e^- , \quad n = 1,2,3 \quad (2.1a)
\]

\[
\text{NH}_3 \xrightarrow{\text{energy}} \text{NH}_m^+ + (3-m)\text{H} + e^- , \quad m = 1,2 \quad (2.1b)
\]

\[
\text{N}_2 \xrightarrow{\text{energy}} \text{N}_2^+ + e^- \quad (2.1c)
\]

(ii) Gas phase dissociation

\[
\text{SiH}_4 \xrightarrow{\text{energy}} \text{SiH}_n^+ + (4-n)\text{H} , \quad n = 1,2,3 \quad (2.2a)
\]

\[
\text{NH}_3 \xrightarrow{\text{energy}} \text{NH}_m^+ + (3-m)\text{H} , \quad m = 1,2 \quad (2.2b)
\]

\[
\text{N}_2 \xrightarrow{\text{energy}} 2\text{N} \quad (2.2c)
\]

(iii) Gas phase reaction

It is assumed that the SiH\textsubscript{2} and NH species are the most important intermediates in the gas phase reactions [10]:

\[
\text{SiH}_2 + \text{SiH}_4 \longrightarrow \text{Si}_2\text{H}_6 \quad (2.3a)
\]

\[
\text{NH} + \text{SiH}_2 \longrightarrow \text{SiNH}_3 \quad (2.3b)
\]

(iv) Condensation and film growth

\[
\text{SiH}_2 + \text{NH} \quad (2.4)
\]
Hydrogen elimination by cross-linking

\[
\begin{align*}
\text{Si} & \quad \text{Si} \\
\text{H} & \quad \text{H} & \quad \text{Si} \\
\text{N} & \quad \text{Si} \\
\text{Si} & \quad \text{Si} \\
\end{align*}
\quad \rightarrow
\begin{align*}
\text{Si} & \quad \text{Si} \\
\text{N} & \quad \text{Si} \\
\text{Si} & \quad \text{Si} \\
\text{Si} & \quad \text{Si} \\
\text{H} & \quad \text{H} \\
\text{H} & \quad \text{H} \\
\text{N} & \quad \text{Si} \\
\text{Si} & \quad \text{Si} \\
\text{Si} & \quad \text{Si} \\
\end{align*}
\quad + \quad \text{H}_2
\]

The breaking of the Si-H and N-H bonds in Eqn. (2.5) is directly dependent on deposition temperature and on the extent of ion bombardment [10].

### 2.2.3. Wet and Dry Etching

Etching of materials is a subtractive process that is used to define a pattern in a thin film or substrate. The process involves the removal of a solid surface of a material by physical and/or chemical means. Wet etching is a chemical process that occurs in a liquid phase. An example of wet etching is the removal of the native oxide layer from a Si substrate using HF, via the following chemical reaction:

\[
\text{SiO}_2 + 6\text{HF} \rightarrow \text{H}_2\text{SiF}_6 + 2\text{H}_2\text{O}
\]  

(2.6)

The reaction is fast, however HF does not react with Si, so the substrate is not damaged during the etching process. HF can also be used to etch Si$_3$N$_4$ [11]:

\[
\text{Si}_3\text{N}_4 + 18\text{HF} \rightarrow 3\text{H}_2\text{SiF}_6 + 2(\text{NH}_4)_2\text{SiF}_6
\]  

(2.7)

Wet etching of photoresists using a solvent (e.g. acetone) is a fast process that can be exploited to define thermally evaporated thin film materials. Due to the non-conformal nature of a thermally evaporated material (i.e. Au), an image-reversal photoresist with sidewall profiles as shown in Fig. 2.4 can be etched away after the material has been evaporated onto its top surface. The sidewall profile of the resist creates an opening for the solvent, which dissolves the resist by undercutting the thermally evaporated layer. This is known as a lift-off process and requires only solvents to define the pattern of the film.

Dry etching of a surface can occur through ion bombardment, chemical reaction between reactive species, or by combined physical and chemical mechanisms. An example of dry etching is reactive ion etching (RIE). The process involves the introduction of a reactant gas or gases, such as CF$_4$ into a chamber. The gas is then
dissociated by energetic particles, i.e. plasma electrons. The radicals, such as CF$_3^+$, CF$_3$ and F that are produced during the dissociation step migrate towards the substrate sample surface, where they become adsorbed, diffuse over the surface and react with surface species [1]. The reaction products are then desorbed and migrate back into the bulk gas. The plasma can be either RF-generated and/or inductively coupled plasma (ICP)-generated. ICP sources create high-density, low-pressure, low-energy plasma, which increase the etch rate of the dry etch process. As dry etching utilises both physical and chemical processes, a wet etch process tends to have a higher selectivity due to its reliance on chemical reactions only. This is advantageous when trying to etch only a single layer of material from a multilayered structure made from different materials.

2.3. **Bulk Micromachining**

Bulk micromachining is a special case of wet etching and involves the etching of the substrate. Bulk micromachining is used to etch materials such as Si, SiC, GaAs, InP and Ge [1]. With excellent machinability and low cost, single crystal Si is the most common bulk micromachined material in industry and has been used to fabricate a range of MEMS devices including pressure sensors and accelerometers [3, 12].

---

**FIG. 2.4.** Wet etching of photoresist in a lift-off process.
The atoms in a Si crystal are arranged in a diamond lattice structure, with tetrahedral atomic coordination, as shown in Fig. 2.5. The structure is face-centred cubic, with four extra atoms inside the unit cell filling the tetrahedral positions [13]. The \{111\} planes present the highest packing density. Consequently, these planes etch at the slowest rate in an anisotropic etchant among all planes in the crystal structure [1]. Table 2.1 summarises three common etchants and their associated etch rate ratios relative to the (100) and (111) planes for single crystal Si [14, 15]. A solution of hydrofluoric, nitric and acetic acids (HNA) etch both the (100) and (111) planes at the same rate. This type of etchants is known as an isotropic etchant. Tetramethyl-ammonium hydroxide (TMAH) and KOH show drastic differences in etching rate between the two planes, and are anisotropic. KOH exhibits extreme selectivity for (100) planes relative to (111), which is a highly desirable property for certain fabrication processes. The actual etch rate of crystal planes depend largely on agitation, concentration of the reactants and temperature of the solution [14, 16, 17].

Si wafers are commonly oriented so that their planar faces are parallel to the (100), (110) or (111) planes. Using an isotropic etchant however, it is irrelevant what the orientation of the wafer is. Figure 2.6 (a) and (b) illustrate schematically cross-sectional and isometric views, respectively of a (100)-oriented Si substrate etched anisotropically in KOH solution. The top surface of the substrate is initially masked with a thin film material and patterned either by wet or dry etching to expose the underlying silicon.

![FIG. 2.5. Unit cell of crystalline Si.](image-url)
Due to the anisotropic nature of the etchant, the (111) planes are etched much slower than the (100) and (110) planes, resulting in sloped surfaces at an angle 54.74° from the horizontal. Upon further etching, these sloped surfaces will meet at either a single point or line, depending on the shape of the mask. If the thin film mask is patterned to expose the Si substrate in the shape of a square, the etched silicon will be in the shape of a pyramid. In comparison, a rectangular mask opening will form V-grooves in the substrate.

Figure 2.7 illustrates the process steps necessary to release a thin film microcantilever from a (100)-oriented Si substrate using bulk micromachining techniques. The film is first deposited uniformly across the top surface of the wafer (Fig. 2.7(a)). The film is then patterned in the shape of microcantilevers (Fig. 2.7(b)). After initial wet etching of the Si substrate in KOH solution, a break in the crystallographic planes underneath the corners of the beam occurs. This results in further etching of the Si substrate underneath the beam along its length, eventually working its way to the beam’s base (Fig. 2.7(c)). A schematic illustration of a top view of a released microcantilever is shown in Fig. 2.8 (a). For such a process to work successfully, the sides of the beam must be oriented so that they lie along the <110> directions. However,

### Table 2.1

<table>
<thead>
<tr>
<th>Etchant</th>
<th>(100)/(111) etch rate ratio</th>
</tr>
</thead>
<tbody>
<tr>
<td>HF, HNO₃, CH₃COOH (HNA)</td>
<td>1:1</td>
</tr>
<tr>
<td>TMAH</td>
<td>up to 35:1</td>
</tr>
<tr>
<td>KOH</td>
<td>up to 400:1</td>
</tr>
</tbody>
</table>

FIG. 2.6. Etching of (100)-oriented Si showing (a) cross-sectional view, and (b) isometric view.
this method will not work for fixed-fixed beams, as shown in Fig. 2.8(b). In this case, two separate V-grooves will be etched into the substrate on either side of the beam, without under cutting to release the beam. To release a fixed-fixed beam using (100)-oriented Si, the structure must be aligned away from the $<110>$ directions, as shown in Fig. 2.8(c). An optical image of an actual released SiN$_x$H$_y$ fixed-fixed beam using this method is shown in Fig. 2.9.

The supports of the fixed-fixed beam in Fig. 2.9 are obviously non-ideal. Fixed-fixed beams with ideal supports can be successfully bulk micromachined using (111)-oriented Si wafers [18]. Figure 2.10 shows an optical micrograph of an actual released bulk micromachined SiN$_x$H$_y$ fixed-fixed beam, deposited on an (111)-oriented Si wafer. The steps required to fabricate a beam in this manner are shown in Fig. 2.11. The mask

---

**FIG. 2.7.** Schematic illustrations of the process steps used to release a microcantilever from (100)-oriented Si using bulk micromachining techniques.
layer is patterned in the shape of a hexagon, with a fixed-fixed beam suspended across its centre. The sides of the hexagon represent \(<110>\) directions in the (111)-oriented Si wafer \([17]\). The process involves a dry etch step of the Si-substrate to define the cavity size underneath the beam. If this step is not carried out, long etch times will be necessary to release the beam. As seen in Fig. 2.10, the Si substrate is in focus, which suggests the cavity between the beam and the substrate is very small. In this case, the cavity size was \(\sim2\ \mu\text{m}\).

Circular, square and rectangular thin film membranes that are completely constrained around their perimeter are common MEMS structures have been used to sense changes in pressure \([12]\). MEMS-based pressure sensors are often fabricated using bulk micromachining techniques. Circular-shaped membranes utilise an isotropic etchant of the substrate, whilst square or rectangular-shaped membranes utilise an anisotropic etchant. Figure 2.12 illustrates the steps required to fabricate a square membrane made of thin film material using (100)-oriented single-crystal Si as the substrate and KOH as the anisotropic etchant. Thin film material is initially deposited on both sides of the wafer (Fig. 2.12(a)). Patterning of one of the films into the shape of a square is then carried out (Fig. 2.12(b)), followed by an anisotropic etch of the substrate (Fig. 2.10(c)). Figure 2.12(d) illustrates the bottom view of the released membrane. Due to the angle of the (111) planes, the sides of the released membrane are smaller in dimension than the etched pattern. The released membrane dimensions however can be easily calculated if the thickness of the substrate is known.
FIG. 2.9. Optical micrograph of a released SiN$_x$H$_y$ fixed-fixed beam. The beam was released by etching the (100)-oriented Si substrate with KOH solution.

FIG. 2.10. Optical micrograph of a released SiN$_x$H$_y$ fixed-fixed beam. The beam was released by etching the (111)-oriented Si substrate with KOH solution.
FIG. 2.11. Process steps for fabricating a bulk micromachined fixed-fixed beam with ideal supports using (111)-oriented Si wafer.
2.4. Surface Micromachining

Surface micromachining utilises a sacrificial layer, which is either wet or dry etched, to release a structure. In comparison to bulk micromachining which involves the etching of bulk materials fabricate three-dimensional structures, surface micromachined features are built up layer by layer on the surface of the substrate [1].

Figure 2.13 illustrates the steps involved for the fabrication of a surface micromachined thin film microcantilever. Firstly, sacrificial layer material is deposited onto the substrate (Fig. 2.13(a) and (b)). This layer is then patterned using either a wet or dry etch process to define the support of the microcantilever (Fig. 2.13(c)). A thin film is then deposited onto the sacrificial layer (Fig. 2.13(d)). Patterning of this layer into the shape of a microcantilever then carried out (Fig. 2.13(e)). The sacrificial layer is then removed using either a wet or dry etching process to release the beam (Fig. 2.13(f)).
FIG. 2.13. Process steps necessary for the fabrication of a surface micromachined microcantilever.
Chapter 2 - MEMS Fabrication Principles

The choice of sacrificial layer material to use is dependent on the selectivity of its removal process. Polyimide has previously been used as a sacrificial layer for defining the cavity in surface micromachined FP filters [19]. This is because it can be easily removed in a weak base solution at room temperature. The solution selectively etches polyimide whilst having no affect on the other materials in the filter. Polyimide also has the advantage of being a low-cost material that is easily spun onto a substrate, and can be partially cured at relatively low temperatures (130 °C). This makes it compatible with temperature sensitive substrates (e.g. Hg$_{1-x}$Cd$_x$Te).

2.5. Fabrication Issues

2.5.1. Step Coverage

Step coverage is a major issue when fabricating MEMS structures using surface micromachining techniques. PVD thin films are especially prone to poor step coverage due to the highly directional nature of the deposition process. As illustrated in Fig. 2.14, if the sample is positioned perpendicular to the direction of path of the evaporant, only the horizontal surfaces will be coated with the evaporated material while vertical side walls will receive little evaporated material.

To improve the continuity of a PVD film, the side wall angle of the underlying thin film needs to be altered. This can be achieved using a photoresist with an overcut profile, which is used to define the pattern of the underlying thin film, followed by a dry etch. An example of the procedure is shown in Fig. 2.15. In this case, the thin film underlying the

![FIG. 2.14. Step-coverage of thermally evaporated Au.](image)
FIG. 2.15. Process steps necessary to improve step-coverage of PVD films.
PVD film (e.g. Au) is SiN$_x$H$_y$ (Fig. 2.15 (a)). Photoresist is spun on and baked onto the SiN$_x$H$_y$ surface (Fig. 2.15 (b)) and patterned (Fig. 2.15 (c)). A low exposure dose of UV creates an overcut profile of the resist edges. Dry etching using CF$_4$ and O$_2$ not only etches the exposed SiN$_x$H$_y$ layers, but also etches to a large degree the resist layer. The pattern of the resist layer is consequently transferred to the underlying SiN$_x$H$_y$ layer which will aid in the continuity of the following PVD deposited film (Fig. 2.15(e), (f)).

2.5.2. Thin Film Stress

Nearly all thin film materials exist in a state of stress due to a mismatch in thermal expansion coefficient, lattice mismatch, impurities, growth process and non-uniform plastic deformation [1]. Residual stress in thin films can be divided into two categories: growth stresses and extrinsic stresses. Growth stress, also known as intrinsic stress, is developed during film nucleation [1, 20]. Intrinsic stress, in thin films is usually a consequence of the non-equilibrium nature of thin film deposition processes [1]. Changes in the growth of a film during deposition can lead to changes in stress through the film’s thickness. Such stresses are known as stress gradients [21-29]. Changes in the growth of a film can be caused by lattice mismatch, impurities, substrate heating during the deposition process and phase transformations [1, 24, 25]. The resulting stress gradient may be induced through dislocations or changes in the crystal lattice, grain size or packing density through the film’s thickness.

Extrinsic stresses refer to stresses that arise from external factors such as temperature gradients during processing and packaging [21]. Sources of extrinsic stresses include differences in thermal expansion coefficients between bonded layers due to temperature changes, chemical reactions of the film with the environment, and redistribution of intrinsic stresses caused by micromachining of structures of the films, which interrupt the continuity of the films [26, 27,]. Intrinsic and extrinsic stress can occur in PVD and CVD thin films [21-27] and can be compressive or tensile in nature.

A film that contains a residual stress will ultimately transfer this stress to the substrate, resulting in substrate deformation. This is undesirable, especially if further processing is required. A flat substrate is desirable especially for lithography processes, and large substrate curvatures can disrupt heat transfer in PVD and CVD processes. If the adhesion of the film to the substrate is poor, large residual stresses can result in film
delamination [32-40]. In certain circumstances, a compressively stressed thin film can delaminate from the substrate in the form of ‘telephone cord’ shaped patterns. An example of this morphology is shown in Fig. 2.16 [40]. In this instance a 300 nm thick tungsten film has delaminated from its substrate in regular, sinusoidal patterns in the plane of the film. The buckled film adopts this morphology in order to balance the driving force for delamination and the interfacial fracture energy [40]. In contrary to compressively stressed films, large tensile stressed films can crack, which is also undesirable [41-43].

Structures fabricated from thin films that contain residual stress can behave differently to thin film structures containing zero stress. The difference in behaviour is not only dependent on the type and magnitude of the stress it is also dependent on the type of structure, as well as its dimensions and constraints. Figure 2.17 provides examples of how stress can affect simply supported surface micromachined cantilever and fixed-fixed beams. For the case of a microcantilever beam fabricated from a compressively stressed film, such that the stress is homogenous through its thickness, the beam will increase in length once the sacrificial layer is removed. This increase in length relaxes the stress from the beam, resulting in a stress-free structure. The same effect will also occur for a microcantilever beam that is fabricated from a homogenous tensile stressed film, which will shorten its length once the sacrificial layer is removed. A microcantilever
beam that is fabricated from a film with a non-homogenous stress through its thickness (i.e. a stress gradient) will bow either up or down once the sacrificial layer is removed. In this particular example, the top surface of the microcantilever is more tensile (or less compressive) than the bottom surface, resulting in the beam curving upwards. The opposite curvature will form if the sign of the stress gradient is reversed.

A fixed-fixed beam (Fig. 2.17(b)) that is fabricated from a homogenous compressively stressed thin film will buckle, provided that the compressive stress is greater than the Euler buckling limit, which is also a function of the Young’s modulus of the film and beam geometry [12]. The shape of the bulking profile is dependent on the type of constraints of the beam, as well as the beam’s dimensions [44]. Thin film fixed-fixed beams generally only experience first-order buckling profiles, however higher order profiles are possible if the long beams are fabricated [44]. For a fixed-fixed beam
fabricated from a homogenous tensile stressed film, its shape will not change after the removal of the sacrificial layer. Consequently, the stress in the original film is conserved in the beam so long as the stress in the beam is less than its failure strength. Fixed-fixed beams that experience a stress gradient will maintain their original state after release provided the net stress in the beam is tensile, or less than Euler buckling limit if it is compressive.

Circular, square and rectangular membranes that are constrained around their perimeter behave similarly to fixed-fixed beams when subjected to residual stress. As their dimensions and boundary conditions are different compared to fixed-fixed beams, it is expected the amount of stress required fail or buckle a membrane will be different. Due to their higher constraints, membranes can achieve higher orders of buckling compared to fixed-fixed beams. Figure 2.18 illustrates buckled PECVD silicon nitride thin film square membranes of varying lateral dimensions [45]. It is obvious that the magnitude of buckling the membrane experiences is dependent on its dimensions.

FIG. 2.18. Optical micrographs of square PECVD silicon nitride membranes with lateral dimensions (a) 601 µm, (b) 1014 µm, (c) 1425 µm, and (d) 1980 µm [45].
2.5.3. Stiction

Utilising a wet etch process to remove a sacrificial layer can be advantageous with regards to selectivity, cost and time. However, once the sample is wet released, it will need to be dried. Drying of MEMS structures and devices is not a trivial process, especially for surface micromachined structures where distances between suspended structures and the underlying substrate can be less than a few microns. A drying process that removes unwanted liquid from a sample by evaporation will cause any liquid trapped underneath a suspended structure to decrease in volume. The forces due to surface tension will pull the suspended structure towards the substrate. Once the structure comes into contact with the substrate, van der Waals’ dispersion forces between the two bodies will bond the two surfaces together. If these forces are greater than the restoring force of the suspended structure, the structure will be permanently bonded to the substrate. This is known as stiction, and can severely limit device yield and greatly increase fabrication costs [3].

Stiction is a common problem when drying surface micromachined structures. Critical point drying (CPD) is a common and successful procedure which helps to avoid stiction under these circumstances [3]. In this procedure, samples are initially dehydrated in a fluid that is miscible with liquid CO₂ (e.g. methanol). Samples are loaded into a CPD chamber (still submerged in the dehydrating fluid), which is then sealed. The fluid is then displaced with liquid CO₂ at room temperature and 10 atm pressure. The chamber is then heated, and due to the fixed volume of the chamber, the pressure also increases. The process is designed to increase the pressure and temperature of the liquid CO₂ past its critical point (31.1 °C, 73 atm) and in doing so the phase transition from liquid CO₂ to gaseous CO₂ is avoided. The chamber, once filled with this supercritical gas, is slowly released from its pressure back to room pressure, whilst maintaining the elevated temperature. The result is a dried sample with structures feely suspended above the substrate.

2.6. Summary

This chapter has reviewed the techniques and principles for the synthesis of thin film materials and fabrication of MEMS structures using thin film materials. The methods discussed have been utilised in this study to fabricate structures for the determination of
material properties of thin film materials, as well as for the fabrication of FP filters. Details on this work can be found in Chapters four and five of this thesis, respectively.

References


Chapter 3

Thin Film Mechanics and Actuation Methods

As essentially mechanical structures, thin-film components of MEMS devices are designed to experience mechanical forces. For example, the FP filter illustrated in Fig. 1.7 utilises electrostatics to provide a distributed load over an electrode’s surface, which results in movement of the top mirror towards the bottom mirror. In order to understand how a thin film structure behaves under load requires the knowledge of basic solid mechanics and actuation principles. This chapter outlines the mechanics of thin films deposited on substrates, thin-film microcantilever beams, fixed-fixed beams and membranes under various loads, such as point loads, distributed loads, and internal stresses. Common MEMS actuation principles are also presented with particular emphasis to those that are relevant to this work. Finally, the vibration behaviour of thin film structures subjected to different environments is also discussed. The structures described in this chapter are those directly related to determining the properties of thin films presented in Chapter 4 and FP filter designs presented in Chapter 5 of this thesis.

3.1. Mechanics of Materials

As a brief introduction to mechanics, the elastic prismatic bar of initial length ($L$) and cross sectional area ($A$) is presented in Fig. 3.1. When a force ($F$) is applied axially to the bar, it will elongate to $L_1$ in the direction of the force. The axial stress ($\sigma_a$) experienced by the bar due to $F$ can be described as:

$$\sigma_a = \frac{F}{A}$$  \hspace{1cm} (3.1)

The axial strain ($\varepsilon_a$) experienced by the bar can be described using:
where $\Delta L = L_1 - L$. When a material behaves elastically and exhibits a linear relationship between stress and strain, it is said to be linear elastic [1]. This linear relationship can be described using Hooke’s law:

$$\sigma_a = E\varepsilon_a$$  \hspace{1cm} (3.3)$$

where $E$ is a constant of proportionality known as the Young’s modulus. For a material that is linear elastic, the axial elongation will be accompanied by a lateral contraction and the strains in the axial and lateral directions are proportional. This will only be true if the material is homogenous throughout the bar and its elastic properties are the same in all directions perpendicular to the longitudinal axis. The relationship between $\varepsilon_a$ and the lateral strain ($\varepsilon_l$) can be described using:

$$\nu = -\frac{\varepsilon_l}{\varepsilon_a}$$  \hspace{1cm} (3.4)$$

where $\nu$ is known as the Poisson’s ratio. For a bar in tension (as is the case in Fig. 3.1) the axial strain is positive and the lateral strain is negative. If the bar experiences a compressive stress, the lateral strain would be positive and the axial strain would be negative. For most purposes, $\nu$ is assumed to be the same in both tension and compression [1].

Similar to axial stress, shear stresses ($\tau$) are accompanied by shear strains ($\gamma$) and for linear elastic materials, are related by Hooke’s law in shear:

$$\tau = G\gamma$$  \hspace{1cm} (3.5)$$

where $G$ is known as the shear modulus and is related to $E$ by the following equation [1]:

$$G = \frac{E}{2(1+\nu)}$$  \hspace{1cm} (3.6)$$
Equation (3.6) shows that $E$, $G$ and $\nu$ are not independent elastic properties of a material. These properties describe how a material will deform when subjected to external and internal forces, so long as the forces do not induce permanent deformation or fracture of the material. Generally, structures are designed to function in the linear elastic region.

### 3.2. Plane Stress and Plane Strain

MEMS structures and devices are usually fabricated from thin-film materials with thicknesses that can be orders of magnitude smaller in dimension compared to the length and width of the structure. In these circumstances, plane stress and plane strain behaviour can exist and are affected by boundary conditions. For example, the internal stress in an isotropic thin film deposited onto a substrate can be described as biaxial. Biaxial stress is a special condition of plane stress which is depicted using the element shown in Fig. 3.2(a). The element has dimensions $a$, $b$ and $c$, in the $x$, $y$, and $z$-directions, respectively and is subjected to a plane stress condition where it only experiences normal stresses in two of the three directions, i.e. $\sigma_x$ and $\sigma_y$, and a shear stress $\tau_{xy}$. The normal strains $\varepsilon_x$, $\varepsilon_y$, and $\varepsilon_z$ can be expressed in terms of stresses $\sigma_x$ and $\sigma_y$ by superimposing the effects of the individual stresses [1]:

$$\varepsilon_x = \frac{1}{E}(\sigma_x - \nu\sigma_y)$$  \hspace{1cm} (3.7)

$$\varepsilon_y = \frac{1}{E}(\sigma_y - \nu\sigma_x)$$  \hspace{1cm} (3.8)

$$\varepsilon_z = -\frac{\nu}{E}(\sigma_x + \sigma_y)$$  \hspace{1cm} (3.9)

For the biaxial stress condition of the thin film shown in Fig. 3.2(b), $\tau_{xy} = 0$ and $\sigma_z = \sigma_y$ and consequently Eqns. (3.7), (3.8) and (3.9) simplify to:

$$\varepsilon_x = \frac{\sigma_x}{E}(1-\nu) = \varepsilon_y$$  \hspace{1cm} (3.10)

$$\varepsilon_z = -\frac{2\sigma_y\nu}{E}$$  \hspace{1cm} (3.11)

Rearrangement of Eqn. (3.10) yields:

$$\sigma_x = \frac{E}{(1-\nu)}\varepsilon_x$$  \hspace{1cm} (3.12)
Denote $E' = E/(1-\nu)$ as the biaxial modulus, and Eqn. (3.12) becomes:

$$\sigma_x = E' \varepsilon_x$$

(3.13)

Hence any uniform stresses that occur parallel to the surface of the thin film on a substrate can be described using the biaxial modulus.

In order to understand plane strain behaviour, consider the small element of dimensions $a$, $b$ and $c$, in the $x$, $y$, and $z$-directions, respectively shown in Fig. 3.3 [1]. The element is subjected to a plane strain condition when it only experiences normal strains in two of the three directions, i.e. $\varepsilon_x$ and $\varepsilon_z$, and a shear strain $\gamma_{xz}$. The element in plane strain condition has no normal strain $\varepsilon_y$ and no shear strains $\gamma_{xy}$ and $\gamma_{yz}$, which results in the faces of the element parallel to the $x$-axis (in Fig. 3.3) being fully restrained against displacement.
3.3. Examples of Plane Stress and Plane Strain Behaviour in MEMS

3.3.1. Thin Film Beams

Plane strain behaviour is common in thin film microcantilever and fixed-fixed beams whose thicknesses are much smaller in dimension compared to their lengths and widths. Consider the microcantilever beam that is subjected to an external point load \( F \) shown in Fig. 3.4. The microcantilever is constrained along the \( y \)-axis so that all displacements and rotations are zero. Under the applied load, the beam will deflect and the curvature \( (κ) \) of the deflection curve is equal to:

\[
κ = \frac{d^2w}{dx^2}
\]  

where \( w \) is the deflection of the beam in the \( z \)-direction. Assuming \( w \) is small relative to beam length \( (L) \) and that the thickness of the beam \( (t) \) is small relative to the beam’s width \( (b) \), \( ε_x \) and \( ε_y \) can be described in terms of the normal stresses \( σ_x \) and \( σ_y \) using Hooke’s law [2]:

\[
E E ε_x = σ_x - νσ_y
\]  

\[
E E ε_y = σ_y - νσ_x = 0
\]  

The lateral strain in the \( y \)-direction must be zero in order to maintain continuity in the plate during bending [2]. Rearranging Eqn. (3.16) shows us that \( σ_y = νσ_x \) and substituting this result into Eqn. (3.15) yields:

\[
σ_x = \frac{E}{(1-ν^2)} ε_x
\]
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Denote $E^* = E/(1 - \nu^2)$ as the plate modulus, and Eqn. (3.17) becomes:

$$\sigma_x = E^* \varepsilon_x$$

(3.18)

Hence the stress distribution in a beam that experiences plane strain conditions can be described using the plate modulus, and is applicable for beams which are designed so that $b \geq 5t$ [2, 3].

For a beam that undergoes a deflection due to an external force (such as the beam illustrated in Fig. 3.4), an element of area $dA$ at distance $z$ from the neutral axis will undergo elongation $\varepsilon_x$ equal to:

$$\varepsilon_x = -\kappa z$$

(3.19)

Substituting Eqn. (3.19) into Eqn. (3.18) yields:

$$\sigma_x = -E^* \kappa z$$

(3.20)

The normal stress $\sigma_x$ acting on the element $dA$ in Fig. 3.4 produces an element of moment $(dM)$ equal to $-\sigma_x zdA$. The integral of all such elemental moments over the entire cross-sectional area $(A)$ must equal the bending moment:

$$M = -\int_A \sigma_x z \, dA = \int_A E^* \kappa z^2 \, dA = E^* \kappa I$$

(3.21)

where:

$$I = \int_A z^2 \, dA$$

(3.22)

is the moment of inertia of the beam. For a beam of rectangular cross-section profile, $I$ is expressed as [1]:

...
Substituting Eqn. (3.14) into Eqn. (3.21) yields:

\[ \frac{d^2w}{dx^2} = \frac{M}{E^*I} \]  

(3.24)

This equation can be integrated to find the deflection at any point along the length of the beam. For the microcantilever shown in Fig. 3.4, it can be shown that the deflection at the end of the beam is equal to [1]:

\[ w = \frac{FL^3}{3E^*I} \]  

(3.25)

In certain situations, the thickness of a microcantilever beam fabricated using bulk micromachining processes will be tapered. As discussed in Chapter 2, the release of a microcantilever on (100)-oriented Si in KOH solution occurs across the length of the beam as the substrate is removed. If some etching of the beam occurs during this process, the beam’s cross section will look similar to that shown in Fig. 3.5 below:

For the case where a microcantilever beam’s thickness decreases linearly along its length towards the tip, i.e. its thickness at its base is greater than at its end, Eqns. (3.23) and (3.25) require further development as the beam’s bending moment varies along its length. By integrating the bending moment of a cantilever beam of varying thickness, the beam’s deflection \( w \) when subjected to a point load \( F \) can be calculated using:

\[ w = \left( \frac{F}{E^*} \right) \left( \frac{6}{bm^2} \right) \left( 2\ln(1 + \frac{mL}{t_t}) - \frac{mL(3mL + 2t_t)}{(t_t + mL)^2} \right) \]  

(3.26)

FIG. 3.5. Tapered microcantilever beam due to bulk micromachining processes.
where \( t \) is the thickness of the beam at the point where the load \( F \) is applied and \( m \) is the gradient of the beam thickness. The derivation of Eqn. (3.26) is shown in Appendix A of this thesis.

### 3.3.2. Thin Film Membranes

In addition to slender beam structures, membranes are another common form of mechanical structures in MEMS devices. Thin film membrane structures are commonly subjected to distributed loads, such as pressure loads or electrostatic loads. Piezoresistive pressure sensors are typical examples of MEMS devices that utilise thin film square or rectangular membranes. These membranes are released from their substrate, but are still constrained around their perimeter. A pressure difference across the membrane results in bulging of the membrane. Piezoresistive sensors deposited on one side of the membrane allow for accurate measurement of the induced bending strain, which can then be correlated to the pressure difference [4]. For a square thin-film membrane with zero residual stress that is constrained around the perimeter, the bulge height at the central point of the square is related to the applied pressure by [2, 5-7]:

\[
P = \frac{f(\nu)E'tw_s^3}{d_s^4}
\]

where \( P \) is the uniform applied pressure on the membrane, \( t \) is the membrane thickness, \( w_s \) is the bulge height at the centre, \( a_s \) is the half length of the side of the square, and \( f(\nu) \) is a function of Poisson’s ratio, which is material-independent [5]. Table 3.1 summarizes recently reported expressions for \( f(\nu) \) for square membranes [8].

One would assume the inclusion of \( E' \) is a result of the square membrane experiencing biaxial stress conditions. However, the inclusion of \( f(\nu) \) suggests that this is not entirely true [5]. The strain experienced by the membrane varies from biaxial in the centre of the membrane to plane strain at the edges and consequently \( E' \) is insufficient to characterise the behaviour of the membrane under an applied pressure.

<table>
<thead>
<tr>
<th>Authors</th>
<th>( f(\nu) ) expression for square thin-film membranes.</th>
</tr>
</thead>
<tbody>
<tr>
<td>Pan et al. (1990) [6]</td>
<td>( 1.37(1.446 - 0.427 \nu) )</td>
</tr>
<tr>
<td>Vlassak and Nix (1992) [5]</td>
<td>( (0.800 + 0.062 \nu)^3 )</td>
</tr>
<tr>
<td>Maier-Schneider et al. (1995) [7]</td>
<td>( 1.994(1 - 0.271 \nu) )</td>
</tr>
</tbody>
</table>
For a rectangular membrane with a sufficiently large length-to-width aspect ratio that is constrained around the perimeter, the bulge deformation of the membrane under uniform applied pressure conforms to the plane strain condition, and the bulge contours are parallel to the long sides of the rectangle, except at the ends. For a rectangular membrane of side dimensions $2a_r$ by $2b_r$, the plane strain condition has been empirically determined to be met when $b_r \geq 4a_r$ [5]. Structures that meet this condition are referred to as infinite rectangles in this discussion. Under the plane strain condition, the central line bulge height ($w_r$) and uniform applied pressure relationship is expressed as [5]:

$$P = \frac{4E_t w_r^3}{3a_r^4}$$  \hspace{1cm} (3.28)

Equations (3.27) and (3.28) have been derived using an energy minimisation method, under the condition that the total elastic strain energy of the system is at a minimum with respect to the unknown parameters of the displacement field [5]. The potential energy of a membrane under a uniform applied pressure is represented by the strain energy of the membrane due to stretching, as well as the potential energy of the pressure applied to the membrane. Equations (3.27) and (3.28) do not include the effects of bending as part of the total strain energy of the membrane, which is a valid approximation when the membrane thickness is much less than the membrane side dimensions. Consequently, membranes are usually designed so that the $a_r/t$ and $a_t/t$ ratios are greater than 1000 [5, 6, 9-11].

### 3.4. Effects of Stress on Thin Film Structures

#### 3.4.1. Microcantilever Beams

In Chapter 2 it was shown that a microcantilever beam fabricated from a thin film material with in-plane homogenous residual stress ($\sigma_0$) will change its length when released from its substrate. The result is a stress-free microcantilever beam that is planar and parallel to the substrate. The removal of stress during the release process will also occur for microcantilever beams fabricated from thin film materials that exhibit a stress gradient ($\delta\sigma/\delta t$) through the thickness. In this case however, the beam will bow with a radius of curvature ($R$) depending on the sign of $\delta\sigma/\delta t$. The stress gradient may be determined from the strain gradient ($\delta\varepsilon/\delta t$) of the film using [12, 13]:

$$\frac{\delta\sigma}{\delta t} = \frac{\delta\varepsilon}{\delta t} E$$  \hspace{1cm} (3.29)
where:
\[
\frac{\delta e}{\delta t} = \frac{w_g}{L^2} = \frac{1}{R}
\]  
(3.30)

where \(w_g\) is the deflection of the beam at its end due to \(\delta e/\delta t\). Large stress gradients will affect the restoring force of a microcantilever when subjected to a point or distributed load.

Consider Fig. 3.6, which illustrates a cross-sectional view of a microcantilever beam with an initial stress gradient, as well as being subjected to a point load \((F)\) parallel to the \(z\)-axis. The deflection at the end of the beam is equal to the deflection due to the stress gradient \((w_g)\) plus the deflection due to the applied force \((w_F)\). For large stress gradients, the restoring force of the beam due to \(F\) is proportional to the beam’s equivalent length \((L_e)\). The applied force \(F\) is no longer perpendicular to the cantilever, resulting in an equivalent force \((F_e)\) deflecting the beam. If \(w_g\) is of the same order as \(L\), Eqn. (3.25) will not accurately determine the deflection of a beam due to load \(F\), and is only applicable for flat beams that experience small deflections [1].

In the case of a multilayer thin film structure, a stress gradient may be created post film deposition by temperature variations due to a difference in thermal expansion coefficient between the layers. Consider the two-layered cantilever beam shown in Fig. 3.7(a) at Temperature, \(T_1\). The bottom and top layers of the microcantilever are fabricated from films denoted \(film 1\) and \(film 2\), respectively such that their thermal expansion coefficients are not equal. If the beam experiences a temperature change, \(T_2\) (Fig. 3.7(b)) the expansion (or contraction) of \(film 1\) relative to \(film 2\) is less (or more), resulting in an induced curvature of the beam. The radius of curvature can be calculated using Eqn. 3.31 below [14]:

\[ \text{FIG. 3.6. A microcantilever beam with a stress gradient subjected to a point load.} \]
Chapter 3 - Thin Film Mechanics and Actuation Methods

![Diagram of multilayer microcantilevers](image)

**Fig. 3.7.** Multilayer microcantilevers showing, (a) bi-layer before release, (b) bi-layer after release, (d) tri-layer before release, and (d) tri-layer after release.

\[
R = \frac{2E_1t_1t_2b_1b_2(2t_1^2 + 3t_1t_2 + 2t_2^2) + (E_1t_1^2b_1)^2 + (E_2t_2^2b_2)^2}{6(e_2 - e_1)(t_1 + t_2)E_1E_2t_1t_2b_1b_2}
\]  

(3.31)

where \(E_1, t_1, b_1\) and \(e_1\) represent the Young’s modulus, thickness, width and strain of *film 1*, respectively, and \(E_2, t_2, b_2\) and \(e_2\) represent the Young’s modulus, thickness, width and strain of *film 2*, respectively. Induced stress gradients due to thermal expansion mismatch of bi-layers is common when thin films are deposited at higher temperatures than the operating temperature of the structure. At the deposition temperature, the two films may be formed in harmony (i.e. zero stress gradient), however when cooled down to the ambient temperature, stresses are generated due to the thermal expansion mismatch. In order to remove induced stress gradient variations due to temperature variations, thin film structures can be designed with three layers instead of two. The tri-layer microcantilever beam shown in Fig. 3.7(c) consists of *film 1*, *film 2* and *film 1* alternate layers. This structure will remain flat even if the temperature is varied from \(T_1\) to \(T_2\) (Fig. 3.7(d)) as the induced bending moment created between the bottom and centre layers (due to differences in thermal expansion coefficient) cancels out the induced bending moment created between the centre and top layers. However the net bending moment will only equal zero if the thickness of the top and bottom films are the same.

### 3.4.2. Fixed-Fixed Beams

For a fixed-fixed beam constrained so that all displacements and rotations are zero at both ends, a point load \(F\) applied at the centre of the beam will result in a displacement, as shown in Fig. 3.8. The load can be approximated by [4, 12]:

- 70 -
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\[ F = \frac{E^* b \pi^4}{6} \left( \frac{t}{L} \right)^3 w + \frac{b \sigma_0 \pi^2}{2} \left( \frac{t}{L} \right) w + \frac{E^* b \pi^4}{8} \left( \frac{t}{L^3} \right) w^3 \]  

(3.32)

where \( w \) is the central beam deflection, and \( \sigma_0 \) is the residual stress in the beam. The first, second and third terms in Eqn. (3.32) represent the beam’s resorting force due to its bending, residual stress and stretching components, respectively. At small deflections, the restoring force is dominated by bending and residual stress components, while at large deflections it is dominated by the stretching of the beam. Equation (3.32) is only an approximation of the exact solution which can be determined using computational simulations [4], however it has been shown it is accurate to within 1.5% for small deflections [12, 15].

If a fixed-fixed beam undergoes enough compressive stress that it exceeds the Euler buckling limit, it will buckle [4]. The compressive strain (\( \varepsilon_0 \)) experienced by the buckled beam is equal to the Euler strain (\( \varepsilon_{\text{Euler}} \)) plus the axial strain (\( \varepsilon_a \)) [4, 12]:

\[ \varepsilon_0 = \varepsilon_a + \varepsilon_{\text{Euler}} = -\frac{\pi^2}{L^2} \left( \frac{w^2}{4} + \frac{L^2}{3} \right) \]  

(3.33)

where \( w \) is the deflection of the beam at the centre. The residual stress (\( \sigma_0 \)) can be calculated using \( \sigma_0 = \varepsilon_0 E^* \).

3.4.3. Thin Film Membranes

For the case where a membrane contains a tensile residual stress \( \sigma_0 \), the relationship between the applied pressure (\( P \)) and the central point bulge height (\( w_s \)) of a square membrane is given by [5-7]:

\[ P = c_s \frac{\sigma_0 t}{a_s^2} w_s + \frac{f(\nu) E^* t}{a_s^4} w_s^3 \]  

(3.34)

where \( c_s \) is a geometry-dependent constant equal to 3.393 for a square membrane [5]. The linear \( w_s \) term in Eqn. (3.34) represents the effect of the initial tensile residual stress in the membrane. The cubic \( w_s \) term expresses the bulge resistance of the membrane to the applied pressure due to stretching, which is equivalent to Eqn. (3.27). Similarly, for an infinite rectangular membrane, the pressure (\( P \)) and the central line bulge height (\( w_r \)) are related by [5]:

\[ P = c_r \frac{\sigma_0 t}{a_r^2} w_r + \frac{4 E^* t}{3 a_r^4} w_r^3 \]  

(3.35)
where $c_r$ is a geometry-dependent constant and is equal to 2 for an infinite rectangular membrane [5]. It is evident that for large bulge heights, Eqns. (3.34) and (3.35) are reduced to Eqns. (3.27) and (3.28), respectively, since the linear terms become negligible.

Similar to a fixed-fixed beam, a membrane that is constrained around its perimeter will buckle if it experiences a compressive residual stress greater than the Euler buckling limit. Ziebart et al. [16, 17] has shown that the respective stress required to induce buckling in square and rectangular membranes is equal to:

$$\sigma_0 = -\frac{1.091E^*I^2}{a_x^2} \quad (3.36)$$

$$\sigma_0 = -\frac{\pi^2E^*I^2}{3a_x^2} \quad (3.37)$$

Ziebart et al. [16, 17] has also shown that Eqns. (3.34) and (3.35) are unsuitable for describing the bulge height of compressively stressed thin film membranes subjected to a distributed load, even if the compressive stress is less than the Euler buckling limit.

### 3.5. Introduction to Finite Element Modelling

So far, the behaviour of structures fabricated from thin film materials has been described using analytical solutions. When a system becomes complex however, it is often more convenient to describe thin film behaviour using computational methods, such as finite element modelling (FEM). These methods are advantageous as they often provide solutions in graphical form, thus providing more information compared to analytical methods. A FEM simulation involves the subdivision of a complex domain into finite elements and then assigning sets of equations to each element. Elements interact with other elements via nodes and a collection of many elements is called a mesh. Figure 3.8 illustrates two different types of meshes used in this work. The commercial FEM suit CoventorWare version 2006.1 was used to mesh a beam with brick elements (Fig. 3.8(a)) and tetrahedral elements (Fig. 3.8(b)). Brick elements are hexahedral in shape and are commonly used to mesh regular shaped structures. Tetrahedral elements are tetrahedral in shape and are often used to mesh irregular shaped structures and curved surfaces. In this work, the accuracy of models was increased by using parabolic elements. Parabolic elements are more computer-intensive compared to linear elements, however are more accurate [18].
3.6. MEMS Actuation Principles

There are four main actuator principles used to drive MEMS: Electrothermal, Piezoelectric, Electromagnetic and Electrostatic actuation. Electrothermal actuation is commonly used in MEMS switches [19, 20] and relies on the induced bending moment developed between two materials of different coefficients of thermal expansion. For example, the heat generated from resisters can be utilised to induce curvature in bilayer cantilevers. High thermal power densities are necessary, however high potentials are not necessarily required. For example, Schmid et al. [19] have shown that an 800 µm long bilayer cantilever can be deflected by 78 µm with a voltage of only 1.3 V. To achieve repeated actuation, the temperature needs to be altered to change the deflection of the beam. Rapid generation and dissipation of heat energy from resistors is not trivial, and can limit the actuation speeds of the device.

Piezoelectric actuation relies on the bending moment induced by a piezoelectric material. Typical piezoelectric materials used are perovskite compounds, for example BaTiO₃, KNbO₃, KTaO₃, SrTiO₃ and PbTiO₃, and can be deposited in thin film form. High temperature processing is usually required for the fabrication of these films [21]. Electromagnetic actuation is the least common of the four actuation principles and is mainly due to its poor scalability and high power consumption [22].

Electrostatic actuation of MEMS relies on an applied bias voltage to provide a distributed load across a surface in order to create motion. The application of electrostatics in MEMS is common practice due to its ability to generate high frequency precise motion, which is proportional to the applied bias voltage. The main disadvantage of using electrostatics is the non-linear nature of motion that occurs with increasing bias.

![Finite element modelling of a beam showing (a) brick element mesh, and (b) tetrahedral element mesh.](image)
voltage, leading to unstable motion at large displacements and consequent failure of the device. Consider the ideal one-dimensional parallel-plate actuator illustrated in Fig. 3.9. When a bias \( V \) of 0 V is applied between the movable and stationary plates, the distance between them (known as the cavity length) is \( z \). When \( V \) is increased, the induced electrostatic force \( F_e \) displaces the movable plate towards the stationary plate by a distance \( \Delta z \). If fringing fields are neglected, \( F_e \) can be described using:

\[
F_e = \frac{\varepsilon_0 AV^2}{2d^2}
\]  
(3.38)

where \( \varepsilon_0 \) is the permittivity of free space, \( A \) is the surface area of the movable plates and \( d \) is the distance between the two plates. The spring in Fig. 3.9 represents the restoring force \( F_r \), which can be described using:

\[
F_r = -k(z - d)
\]  
(3.39)

where \( k \) is the spring constant. At equilibrium, the sum of the electrostatic and restoring forces will equal zero, i.e. \( F_e = -F_r \):

\[
\frac{\varepsilon_0 AV^2}{2d^2} = k(z - d)
\]  
(3.40)

Dividing both sides of Eqn. (3.40) by \( 1/kz \) yields:

\[
\frac{\varepsilon_0 AV^2}{2kd^2} = 1 - \frac{d}{z}
\]  
(3.41)

Normalising the left hand side of Eqn. (3.41) gives:

\[
\frac{\varepsilon_0 AV^2}{2kd^2} \frac{z^2}{z^2} = \frac{\varepsilon_0 AV^2}{2kz^3} \frac{z^2}{d^3} = 1 - \frac{d}{z}
\]  
(3.42)

Let \( \eta = d/z \) and \( \gamma^2 = 2kz^3/\varepsilon_0 A \). Rearranging Eqn. (3.42) yields:

\[
V = \gamma \eta \sqrt{1 - \eta}
\]  
(3.43)
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The variable $\eta$ represents the normalised displacement of the movable plate. Unstable displacement of the plate will occur when $dV/d\eta = 0$, i.e.:

$$\gamma\sqrt{1-\eta} - \frac{\gamma\eta}{2\sqrt{1-\eta}} = 0$$

(3.44)

Solving for $\eta$ yields:

$$\eta = \frac{2}{3} \Rightarrow d = \frac{2}{3} z$$

(3.45)

Equation (3.45) shows that once the movable plate displaces to 2/3 of the initial cavity length (i.e. $\Delta z = 1/3 z$), unstable plate motion will occur and will snap down onto the substrate. This is known as pull-in and is undesirable as it can lead to failure of the device. Equation (3.45) also shows that for linear spring constants, the pull-in displacement is independent of $k$. For non-linear spring constants, the pull-in displacement can be larger, however is dependent on $k$ [23, 24].

For MEMS applications, cantilevers and fixed-fixed beams are commonly used as electrostatic actuators. The beam itself acts as the top electrode, while the bottom electrode (which is stationary) is deposited onto the substrate. The beams are generally fabricated using surface micromachining processes as smaller cavity sizes are achievable compared to bulk micromachining processes. According to Eqn. (3.38), a smaller cavity size means a smaller voltage is required for actuation. The restoring force (i.e. spring constant) of a cantilever is linear since it relies on pure bending. Consequently, the largest displacement the cantilever can achieve before pull-in is 1/3 the cavity length. For a fixed-fixed beam however, the restoring force is non-linear since both bending and stretching forces exist. Strain stiffening in the beam occurs [23, 24] and results in a more linear relationship of beam displacement with applied voltage compared to a cantilever beam, as well as a larger displacement before pull-in occurs. It has been shown that the maximum theoretical displacement of a fixed-fixed beam is 2/3 the cavity length and is dependent on beam dimensions as well as the cavity length [24].

Pull-in analysis of continuous systems, such as cantilevers or fixed-fixed beams is not trivial [24]. Under an electrostatic load, a beam will deflect towards the substrate, resulting in deformation of the beam. Consequently, the spacing between the top and bottom electrode is not uniform along the length of the beam. The inclusion of fringing fields also complicates matters and cannot be neglected. There have been many attempts to accurately determine analytically the pull-in voltages of microcantilever beams and
fixed-fixed beams, as well as membranes [3, 23-26]. The easiest method is to utilise FEM. CoventorWare uses an iterative procedure to achieve convergence of the stored electrostatic energy to the structural displacement energy. For a given bias, an initial geometry is provided to the electrostatics solver to determine the electrostatic force. This force is provided to the mechanical solver and then simulated to find the updated geometry. This process is repeated until equilibrium is achieved within a certain tolerance [18].

Figure 3.10 plots the simulated maximum displacement versus applied bias of a microcantilever beam and a fixed-fixed beam, as well as a fixed-fixed beam that experiences 50 MPa tensile residual stress. For a fixed-fixed beam, the maximum displacement occurs at its centre, and for a microcantilever beam it occurs at its free end. The forces on the beams were simulated using CoventorWare. The dimensions of the beams were 200 µm long, 10 µm wide and 1 µm thick and the cavity size was 2 µm. The $E$ and $\nu$ values of the beams were set to 100 GPa and 0.25, respectively. The permittivity of the medium between the beams and the substrate was to set to free space. A 5 µm brick mesh with parabolic elements was used to mesh the beams. The boundary conditions

![Figure 3.10. FEM simulations showing maximum beam displacement versus applied bias voltage for a microcantilever beam and fixed-fixed beams.](image-url)
were defined such that all displacements and rotations at both ends of the fixed-fixed beams were set to zero. The same boundary conditions were used for the cantilever, except they were only applied to one end. Figure 3.10 shows far more control relative to applied bias and larger displacements are achievable for a fixed-fixed beam compared to a microcantilever beam. However, larger voltages are required for a fixed-fixed beam compared to a microcantilever beam for the same amount of displacement. Residual stress also has a significant impact on the displacement of a fixed-fixed beam. Hence there is a very strong dependence of boundary conditions, as well as material properties on the actuation voltage of a structure.

3.7. Vibration of Thin Film Structures

One of the main advantages MEMS have compared to mechanical structures on the macro scale is their ability to be actuated at very high frequencies. When scaled, the inertial forces generated by MEMS are usually negligible compared to electrostatic, piezoelectric or thermoelectric forces used to actuate the structure. Consequently, high actuation frequencies are possible and are utilised in many MEMS devices such as switches, projection displays and sensors. Vibration engineering on the macro scale will often involve designing systems to operate at frequencies far removed from their resonant frequencies. In doing so, the performance of the system as well as its lifetime is increased. In MEMS however, microstructures are sometimes designed to operate at or very close to their resonant frequencies in order to increase their displacements.

3.7.1. Natural Frequencies of Thin Film Structures

The vibration of a homogenous, isotropic beam can be described as being continuous, i.e. it has a continuous distribution of mass and elasticity along its length and width and through its thickness. The natural angular frequency ($\omega_n$) of a continuous flat beam can be determined via the Euler-Bernoulli bending vibration equation [27]:

$$\omega_n = (\beta_n L)^2 \frac{E I}{m L^3}$$  \hspace{1cm} (3.46)

where $m$ is the mass per unit length of the beam, $L$ is the beam’s length, and $E$ and $I$ are the Young’s modulus and moment of inertia of the beam. ($\beta_n L$) is the $n^{th}$ eigenvalue of the beam and is dependent of boundary conditions. Table 3.2 provides values of $(\beta_n L)^2$ for the first three bending modes of a cantilever and a fixed-fixed beam.
Equation (3.46) can be rearranged to describe the \( n^{th} \) natural bending vibration frequency \( (f_b)_n \) of a beam with rectangular cross section:

\[
(f_b)_n = \left( \frac{\beta_n L}{\pi} \right)^2 t \sqrt{\frac{E}{48 \rho}}
\]

(3.47)

where \( t \) and \( \rho \) are the thickness and density of the beam, respectively. There is inconsistency in the literature to describe the natural bending frequencies of a thin film microcantilever beam. Some sources suggest that Eqn. (3.47) is a good representation [28-31], however other sources suggest \( E \) in Eqn. (3.47) should be replaced with \( E^* \) as a microcantilever is more representative of a plate [32, 33]. Others also suggest that a shape factor should be included, which takes into account the beam’s length, width \( (b) \) and thickness, as well as \( \nu \)[34].

To clarify this uncertainty, finite element modelling (FEM) simulations were conducted using CoventorWare version 2006.1 to determine the first-order bending resonant frequencies \( f_b \) of microcantelever beams with various length-to-width ratios \( (L/b) \). The frequencies were then used to calculate the Young’s moduli \( (E_{calc}) \) of the beams using Eqn. (3.47). In the simulation, \( E, \rho \) and \( \nu \) were set to 100 GPa, 2.0 g/cm\(^3\) and 0.25, respectively. The parameter space was limited to microcantilever beams with widths equal to 20 \( \mu \)m and thicknesses equal to or less than 1 \( \mu \)m. The lengths of the beams ranged from 5 \( \mu \)m to 600 \( \mu \)m. A 1 \( \mu \)m parabolic brick mesh was used to mesh each beam. Figure 3.11(a) plots the effect of \( L/b \) ratio and changes in \( t \) on \( E_{calc} \). It is clear that as the \( L/b \) ratio increases, \( E_{calc} \) approaches \( E \), and is independent of \( t \) over the simulated thickness range for large \( L/b \) ratios. For beams of small \( L/b \) ratios, \( E_{calc} \) varies with \( t \), and for small values of \( t \), \( E_{calc} \) approaches \( E^* \), while for large values of \( t \) approaches \( E \). This is expected since the length of the beam becomes comparable to its thickness, resulting in a beam behaving less like a plate.

<table>
<thead>
<tr>
<th>Beam</th>
<th>((\beta_1 L)^2)</th>
<th>((\beta_2 L)^2)</th>
<th>((\beta_3 L)^2)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Cantilever</td>
<td>3.5160</td>
<td>22.035</td>
<td>61.697</td>
</tr>
<tr>
<td>Fixed-Fixed</td>
<td>22.373</td>
<td>61.673</td>
<td>120.9034</td>
</tr>
</tbody>
</table>

Table 3.2
Numerical values for \((\beta L)^2\) for typical beam end conditions.
FIG. 3.11. (a) Calculated Young’s modulus, and (b) calculated shear modulus using Eqns. (3.41) and (3.42), respectively from simulated frequency response of microcantilever beams with varying $L/b$ ratios and $t$ values. $E$ and $E^*$ represent the Young’s modulus and the reduced Young’s modulus used in the FEM simulation. $G$ was obtained from Eqn. (3.6).
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The torsional resonant modes of a beam with rectangular cross section can be approximated using Eqn. (3.48) below [29, 30, 35]:

\[
(f_i^T) = \frac{it}{2Lb} \sqrt{\frac{G}{\rho}}
\]  

(3.48)

where \((f_i^T)\) is the \(i\)th natural frequency of torsional vibration of the beam. For a cantilever, \(i = 1, 3, 5\ldots\) and for a fixed-fixed beam, \(i = 2, 4, 6\ldots\) Equation (3.48) assumes \(b >> t\), which is usually the case for thin film beams. Maeda et al. [35] found however for microcantilever beams with \(L/b\) ratios equal to or greater than 5, Eqn (3.48) provides a reasonable estimation. Figure 3.11(b) plots the effect of \(L/b\) ratio and changes in \(t\) on \(G_{calc}\), which was calculated using Eqn. (3.48) and the first order torsional resonant frequencies of the simulated beams. The theoretical value of \(G\) for the simulated beams can be obtained from Eqn. (3.6) giving a value of \(G\) as 40 GPa. As the \(L/b\) ratio increases, \(G_{calc}\) approaches \(G\). For low \(L/b\) ratios however, \(G_{calc}\) greatly overestimates \(G\). The analyses shown in Fig. 3.11 suggests that Eqns. (3.47) and (3.48) provide reasonable approximations for \(E\) and \(G\) determined from the bending and torsional resonant frequencies of a microcantilever beam of width 20 \(\mu\)m, thickness equal to, or less than 1 \(\mu\)m and high \(L/b\) ratio. For example, for a beam of length 160 \(\mu\)m \((L/b = 8)\), Eqn. (3.47) overestimates \(E\) by less than 1.3 %, whereas Eqn. (3.48) overestimates \(G\) by less than 6.4 %.

The above equations describe the bending and torsional vibration frequencies of stress-free, flat microcantilever beams and fixed-fixed beams. The inclusion of stress gradients or residual stresses will result in changes in resonant frequency of the beams. For a fixed-fixed beam, increasing the tensile residual stress \((\sigma_0)\) will increase the beam’s stiffness, resulting in an increase in its resonant frequency. A large stress gradient in a microcantilever beam or a buckled fixed-fixed beam due to a compressive residual stress will effectively change the dimensions of the beams and hence change their resonant frequency. Computational methods are commonly employed to describe a structure’s resonance behaviour that exhibits internal stresses. However, the Rayleigh-Ritz method can be employed to approximate resonant frequencies of non-ideal beams. Details on this method, as well methods to determine stresses in resonating structures can be found in Ref. [36-39].
3.7.2. Squeeze Film Damping

Squeeze film damping is a form of air damping that can occur during actuation of MEMS devices. Consider the plate shown in Fig. 3.12 of area \( A \) and restoring force \( k \). In this instance the cavity between the plate and the substrate is of size \( d \) such that \( d \ll A \). When the plate is displaced towards the substrate, the fluid (i.e. air) trapped inside the cavity will compress (i.e. pressure increases) and is squeezed out from the edges of the plate. When the plate moves back to its original position, the pressure inside the cavity decreases and air is sucked back in. The viscous drag of the air during flow retards the plate motion. This opposing force is known as squeeze film damping [4]. The squeezed film can be described using a dashpot \((c_s)\) and spring \((k_s)\) as shown in Fig 3.12.

Due to their generally small cavity sizes, squeeze film damping is a common occurrence when surface micromachined structures are actuated at high frequencies. If the plate is assumed to be rigid, its motion can be described using the following equation:

\[
m \frac{d^2z}{dt^2} + c_s \frac{dz}{dt} + (k + k_s)z = f(t)
\]

where \( m \) is the mass of the plate, \( z \) describes the displacement of the plate normal to the substrate and \( f(t) \) describes the external excitation force. Equation 3.49 can be rearranged such that it is expressed in terms of damping ratio \((\zeta)\) and the resonant frequency of the undamped system \((\omega_n)\) [27]:

\[
\frac{d^2z}{dt^2} + 2\zeta\omega_n \frac{dz}{dt} + \omega_n^2z = \frac{f(t)}{m}
\]

where

\[
\omega_n = \sqrt{\frac{k + k_s}{m}}
\]

When oscillatory motion exists (i.e. \( \zeta < 1 \)), the general solution to Eqn. (3.50) is:

\[
z = e^{-\zeta\omega_n t} \left( A \sin \sqrt{1-\zeta^2} \omega_n t + B \cos \sqrt{1-\zeta^2} \omega_n t \right)
\]

where \( A \) and \( B \) are determined from initial conditions. Hence the frequency of the damped system \((\omega_d)\) is equal to:

\[
\omega_d = \omega_n \sqrt{1-\zeta^2}
\]

Equation (3.53) shows that not only does squeeze film damping result in smaller plate amplitudes but also changes the plate’s resonant frequency. More accurate models which take into account the flexibility of MEMS structures can be found in ref. [40].
The damping ratio is directly related to the effective viscosity ($\eta_{\text{eff}}$) of the fluid, given by \[4\]:

\[
\eta_{\text{eff}} = \frac{\eta}{1 + 6K_n}
\]  

(3.54)

where $\eta$ is the bulk viscosity of the fluid and $K_n$ is the Knudsen number, which is defined as the ratio of molecular mean free path ($\lambda_m$) to the size of the cavity. Equations (3.53) and (3.54) show that by changing the fluid, or its properties (i.e. pressure) will result in a change in the performance of the MEMS when actuated.

3.8. Summary

This chapter has outlined the behaviour of MEMS structures fabricated from thin film materials when subjected to internal and external forces. Specific emphasis has been made on thin film microcantilever beams, fixed-fixed beams and membranes. The equations and solving methods presented in this chapter will be used in the following chapters to determine the mechanical and physical properties of thin film materials, as well as to characterise and optimise a new Fabry-Pérot filter design.

References


Chapter 4

Properties of Silicon Nitride Thin Films

Silicon nitride thin films, commonly synthesized by various chemical vapour deposition (CVD) processes, are of particular interest for the micro-fabrication of self-supporting structures. Plasma-enhanced CVD (PECVD) films are generally amorphous in structure, and exhibit mechanical properties that are quite different from their conventional bulk crystalline counterparts. Mechanical properties of thin films, including Young’s modulus \((E)\), density \((\rho)\), Poisson’s ratio \((\nu)\), residual stress \((\sigma_0)\), strength, fatigue resistance, and creep resistance are primary design parameters in the fabrication of micro-electro-mechanical systems (MEMS) and devices. Therefore, reliable and accurate determination of these properties is of critical importance.

Direct experimental measurement of the mechanical and physical properties of micro- and nano-scale thin films has always been technically challenging due to the small dimensions of test specimens. To overcome this problem, a number of methods have been developed and attempted. The choice of which method to use depends on the availability of equipment necessary for the undertaking of the measurement as well as the compatibility of the film in question to be fabricated into a particular test structure. It is the latter point which has brought about the demand for a more standard measurement technique which is easy to implement and can be applied to a large range of thin films.

This chapter studies the mechanical properties of PECVD silicon nitride \((\text{SiN}_x\text{H}_y)\) thin films deposited under varying deposition conditions. The density, Young’s modulus, Poisson’s ratio and residual stress of the films have been determined using a range of
techniques. Understanding the dependence of these material properties with changes in deposition conditions will contribute to the realisation of an optimised tunable Fabry-Pérot filter.

4.1. Methods Used to Determine Thin Film Properties

4.1.1. Density

The two most common methods used to determine $\rho$ of thin films are the weighing method [1] and the quartz crystal microbalance (QCM) method [2]. The weighing method simply involves measuring the change in mass of a substrate of known surface area due to the deposition of a thin film of known thickness on its surface. The accuracy of this method is dependent on the resolution of the scales used to measure the film’s mass, as well as the uniformity of the film itself. The QCM method involves measuring the change in the resonant frequency of a quartz crystal of known surface area when coated with a thin film material of a known thickness. The method is based on the principle that the resonant frequency of a solid body is dependent on its mass. QCMs are commonly used as thickness monitors in physical vapour deposition (PVD) systems, mainly due to their low cost and high accuracy.

Another method that utilises resonant frequency principle is the oscillation of microcantilever beams fabricated from thin films. The density of the microcantilever beam is determined from its fundamental bending resonant frequencies, however $E$ as well as the dimensions of the beam needs to be known. Measurement of a microcantilever beam’s bending resonant frequencies is often performed using an atomic force microscope (AFM). Resonating microstructures using an AFM has been used previously to characterise the long-term stability of micromechanical structures [3].

4.1.2. Young’s Modulus

Many methods have been developed to determine $E$ of thin films, including nanoindentation, the microcantilever beam bending test, the resonance method, the bulge test and the micro tensile test. Nanoindentation has been used to determine $E$ and hardness ($H$) of PECVD silicon nitride [4], LPCVD silicon nitride and thermally grown silicon dioxide thin films [5], as well as materials not in thin film form, such as fused
silica, soda-lime glass, and single crystals of aluminium, tungsten, quartz, and sapphire [6]. Also known as depth sensing indentation (DSI), nanoindentation originates from hardness tests, such as Brinell, Knoop, Vickers and Rockwell tests which involve pressing or indenting a hard steel ball or diamond tip of known dimensions against a material with known force, so that a depression is made [7]. The main difference between these methods and nanoindentation is the indentation depths. Typical indentation depths used for nanoindentation are measured in nanometres (hence the name ‘nanoindentation’) rather than microns or millimetres which are common depths used for conventional hardness testers. As the thickness of thin-film materials are commonly of the order of \(\sim 1 \mu m\) or less, nanoindentation is a suitable method to determine \(E\) of thin-film materials. Nanoindentation however does require the knowledge of Poisson’s ratio of the film.

The microcantilever beam bending test utilises a nanoindenter to apply a point load onto a microcantilever beam fabricated from a thin film. The Young’s modulus of the beam is determined directly from its restoring force and the beam dimensions. As microcantilever beams are generally described using the plate modulus, \(\nu\) of the film needs to be known. This method has been used to determine \(E\) of polysilicon [8], indium phosphide [9], aluminium and gold films [10]. The resonance method is similar to the microcantilever beam bending test in that it involves the fabrication of microcantilever beams. It is essentially the same method as that used to determine \(\rho\) in section 4.1.1 however \(\rho\) must be known in order to extract \(E\). This method has been used to determine \(E\) of thermally grown silicon dioxide [11], silicon nitride [12] and gold thin films [13].

Another method that can be employed to determine \(E\) is the bulge test, which involves pressurising a membrane that is constrained around the perimeter and has been released from the substrate, and measuring its maximum bulge height, usually with an optical profilometer. The membrane is of a well-defined geometry, usually being square, rectangular or circular, and conveniently fabricated on single crystal Si or amorphous glass substrates. The method has been used to determine \(E\) of tensile stressed low-pressure chemical vapour deposited (LPCVD) silicon nitride films deposited at 785 °C [14].

Edwards et al. [15] report on measuring \(E\) for tension-stressed LPCVD silicon nitride films using a micro-tensile test. This method is desirable since \(E\) is determined directly from the axial strain in a tensile specimen; however, the method requires accurate
measurement of the strain on an extremely small scale, and access to an advanced measurement system that is not commercially available.

4.1.3. Poisson’s Ratio

The Poisson’s ratio is an extremely difficult mechanical property to measure and current methods used to measure \( \nu \) provide only estimations at best. Utilising the micro-tensile test to measure \( \nu \) is desirable since it is determined directly from the axial and lateral strains in a tensile specimen. However, for reasons discussed above, this method is not always available. Tsai et al. [11] utilised the resonance method to determine \( \nu \) of silicon dioxide thin films by fabricating microcantilever beams using bulk micromachining techniques, and calculating \( \nu \) directly from their bending and torsional resonant frequencies. The measurement of torsional resonant frequencies of a microcantilever beam does require the use of a vibrometer.

Another method that can be used to determine \( \nu \) is the bulge test. Both square and rectangular membranes need to be employed and \( \nu \) is calculated from their bulge height ratios. Vlassak et al. [14] employed this method to determine \( \nu \) of tensile stressed LPCVD silicon nitride thin films. Kim et al. [16] determined \( \nu \) of polysilicon thin films by utilising a nanoindenter to displace surface micromachined double-ring shaped specimens. Though robust, this method requires the knowledge of \( E \), which was determined using nanoindentation.

4.1.4. Thin Film Residual Stress

The most common method used to determine \( \sigma_0 \) is the substrate curvature method. This method has been used previously to determine \( \sigma_0 \) of PECVD silicon nitride [17, 18] and porous silicon [19] thin films. This method relates the induced curvature of the substrate to the stress in the deposited film. The radius of curvature of a substrate can be determined from the divergence/convergence of three parallel laser beams once reflected from the substrate [17, 18], or by optical profilometry [19]. This method requires the knowledge of \( E \) and \( \nu \) of the substrate.

The bulge test also allows for measurement of \( \sigma_0 \). The bulge height of a membrane is directly proportional to the film’s \( \sigma_0 \), and consequently can be easily extracted. This
method has been used to determine $\sigma_0$ of tensile stressed LPCVD silicon nitride [14], tungsten [20] and titanium nitride [21] thin films. Pruessner et al. [9] utilised fixed-fixed beams to determine $\sigma_0$ of InP thin films. For a compressively stressed film, $\sigma_0$ was determined directly from the height of the buckled beams, whilst a nanoindenter was employed to measure the restoring force of a tensile stressed fixed-fixed beam.

4.1.5. Comments on Property Measurement Techniques

The above discussion presents many methods used to determine $\rho$, $E$, $\nu$ and $\sigma_0$ of thin-film materials. Methods such as the bulge test and the resonance method used to determine $E$, $\nu$ and $\sigma_0$ are appealing as more than one parameter can be determined from the same measurement. These methods however do require the fabrication of microstructures, which is contrary to methods such as nanoindentation and substrate curvature methods, which are used to determine $E$ and $\sigma_0$. When performing nanoindentation on a thin film deposited on a substrate, the properties of the substrate inevitably contribute to the measurement. Thus, a certain mathematical treatment is required to extract the true properties of the thin films from the measurements. The microcantilever beam bending test used to determine $E$ is advantageous compared to nanoindentation as the measurements are not influenced by the substrate and thus directly determines the properties of the film. It is advantageous compared to the bulge test as both compressive and tensile stressed thin films can be measured. In comparison to the resonance method, it does not require knowledge of $\rho$ however it does require the knowledge of $\nu$.

4.2. PECVD Silicon Nitride Thin Films

The mechanical and physical properties of PECVD Si$_x$N$_y$H$_z$ thin films have previously been studied by Winchester [22], Dell et al. [23] and Martyniuk [24]. In these studies, the Young’s modulus was determined using nanoindentation assuming $\nu$ of the films was equal to 0.25 and independent of deposition conditions. The residual stress of the films was also determined in these studies and was measured via the substrate curvature method. The deposition system used was an Oxford Instruments PECVD 80 system at an RF frequency of 13.6 MHz. The process gases consisted of ultra-high purity SiH$_4$, NH$_3$ and N$_2$. Winchester and Dell et al. [22, 23, 25] concluded that slightly tensile stressed films (<20 MPa) could be achieved for SiN$_x$H$_y$ films deposited at 125 °C, an RF
power equal to 75 W, a chamber pressure of 875 mTorr and SiH$_4$:NH$_3$:N$_2$ flow ratios equal to approximately 1:10:20. Due to the low deposition temperature and tensile nature of the film, these conditions are ideal for the realisation of flat FP filters monolithically integrated with Hg$_{1-x}$Cd$_x$Te detectors.

Martyniuk measured $E$ and $\sigma_0$ of Si$_x$N$_y$H$_z$ thin films with similar SiH$_4$:NH$_3$:N$_2$ flow ratios, however the chamber pressure and RF power were set at 450 mTorr and 100 W. The lower pressure and higher power resulted in higher $E$ values compared to films deposited using Winchester and Dell’s conditions, however compressively stressed films were realised for films deposited at chuck temperatures below 200 °C. Table 4.1 summarises the conditions used to deposit Si$_x$N$_y$H$_z$ thin films for this study. The same Oxford Instruments PECVD 80 system was utilised. The chamber pressure and RF power of the $\alpha$-series were set to 600 mTorr and 100 W, respectively, whilst the $\beta$-series conditions were set to those proposed by Winchester and Dell. The chuck temperatures were varied from 100-300 °C and 100-150 °C for the $\alpha$-series and $\beta$-series films, respectively. The chamber pressure of the $\alpha$-series was purposely chosen to lie between pressures used by Winchester, Dell and Martyniuk.

In this study, nanoindentation and substrate curvature methods were utilised to determine the Young’s modulus and residual stress of the $\alpha$ and $\beta$-series films. In order to determine the validity of the nanoindentation method, the Young’s modulus of the films was also realised using the microcantilever beam bending method, the bulge test and the resonance method. To determine the validity of the substrate curvature method, the residual stress of the films was realised using the bulge test. This work will not only provide $E$ and $\sigma_0$ values of PECVD Si$_x$N$_y$H$_z$ films of different process conditions (i.e. the $\alpha$-series films), but will also strengthen our confidence in the traditional measurement techniques used to determine $E$ and $\sigma_0$ of thin film materials. Density and Poisson’s ratio measurements of the films will also be carried out in this study. These are integral parameters required for the design and optimization of FP filters using PECVD Si$_x$N$_y$H$_z$.

<table>
<thead>
<tr>
<th>Film Set</th>
<th>SiH$_4$:NH$_3$:N$_2$ (SCCM)</th>
<th>Chamber Pressure (mTorr)</th>
<th>RF Power (W)</th>
<th>Chuck Temp. (°C)</th>
</tr>
</thead>
<tbody>
<tr>
<td>$\alpha$-series</td>
<td>5:50:100</td>
<td>600</td>
<td>100</td>
<td>100-300</td>
</tr>
<tr>
<td>$\beta$-series</td>
<td>5:50:100</td>
<td>875</td>
<td>75</td>
<td>100-150</td>
</tr>
</tbody>
</table>
thin films, however were not measured by Winchester, Dell and Martyniuk. The density of the films was measured using the resonance method as well as the quartz crystal microbalance method. The Poisson’s ratio of the films was measured using the bulge test, as well as the resonance method. By measuring all four parameters ($E$, $\sigma_0$, $\rho$ and $\nu$) of the $\alpha$ and $\beta$-series films with changes in deposition temperature, a better understanding of how changes in process conditions affect these parameters will be realised.

4.3. Surface Morphology and Etch Studies of PECVD SiN$_x$H$_y$ Thin Films

The surface morphology of the SiN$_x$H$_y$ thin films were characterised using atomic force microscopy (AFM) techniques. Figure 4.1 illustrates an AFM image of an $\alpha$-series film deposited at 125 °C. The grain structure of the film is very uniform and is continuous.
over the entire film’s surface. The analysis also showed no evidence of porosity on the surface of the films. The root mean square (RMS) surface roughness of the film is less than 1 nm over a 1 µm² area. It was found that changes in surface roughness were negligible with changes in deposition conditions. This observation agrees with Martyniuk [24] who carried out extensive studies on the surface morphology of SiNₓHᵧ thin films which were deposited using the same process equipment as in this study.

The chemical properties of the films were characterised by etch rate measurements in potassium hydroxide (KOH) solution. Etch studies can provide information on the chemical structure of a thin film, as well as physical information such as density or porosity. Hydrofluoric acid (HF) or buffered oxide etch (BOE) solutions are often used to study etch rates of thin film materials. KOH was chosen in this study as the etch rate of low-temperature PECVD SiNₓHᵧ in HF or BOE solution was too fast to quantify. Figure 4.2 plots the etch rates of the α and β-series films in 40-wt. % KOH solution at 80 °C. This solution is commonly used to etch Si in bulk micromachining processes [26]. As the deposition temperature decreases, the etch rate of the films increases hyperbolically. The results also show that the etch rates of the α-series films are less than the β-series films relative to deposition temperature. These results suggest that the chemical and/or physical compositions of the materials vary with process conditions.

![Figure 4.2: SiNₓHᵧ thin film etch rates in 40-wt.% KOH solution at 80 °C.](image)

FIG. 4.2. SiNₓHᵧ thin film etch rates in 40-wt.% KOH solution at 80 °C.
4.4. Determination of Density Using the Quartz Crystal Microbalance Method

4.4.1. Basic Formulations for Property Determination

For the quartz crystal microbalance (QCM) method used to determine $\rho$, the mass of the film deposited on the quartz crystal can be calculated using the following [2]:

$$M_f = M_q \left( \frac{\Delta F}{F_q} \right)$$

(4.1)

where $M_f$ is the mass of the deposited film, $M_q$ is the mass of the quartz crystal, $\Delta F$ is the change in resonant frequency of the crystal after the film has been deposited on its surface, and $F_q$ is the initial resonant frequency of the crystal. The density of the film can be calculated for given film thickness and area.

4.4.2. Experimental Details

The resonant frequency of a virgin quartz crystal of known surface area was first measured using an INFICON Thin Film Deposition Controller and then measured again after the film was deposited on the crystal. The thickness of the film was obtained using a surface profiler on a witness sample that experienced the same deposition event as the crystal. For each deposition event, three crystals were positioned inside the chamber in order to determine the error in the measurement.

4.4.3. Results to the Quartz Crystal Microbalance Method

Figure 4.3 shows the effect of chuck temperature on the density of the SiN$_x$H$_y$ thin films, as determined from the QCM measurements. The error bars represent the standard deviation in the measurement. It is seen that $\rho$ of the films increases continuously with increasing deposition temperature and that $\alpha$-series films exhibit higher $\rho$ values compared to $\beta$-series films. The highest $\rho$ achieved was 2.80 ± 0.05 g/cm$^3$ for a SiN$_x$H$_y$ film deposited at 300 °C, 600 mTorr and 100 W. This is close to the theoretical value of 3.3 g/cm$^3$ for bulk polycrystalline Si$_3$N$_4$ [27]. The lowest $\rho$ achieved was 1.55 ± 0.06 g/cm$^3$ for a SiN$_x$H$_y$ film deposited at 100 °C, 875 mTorr and 75 W. Figure 4.3 shows that $\rho$ of the films is highly sensitive to changes in process conditions and that bulk silicon nitride values cannot accurately represent the true $\rho$ value of the deposited $\alpha$ or $\beta$-series films.
4.5. Determination of Young’s Modulus and Poisson’s Ratio Using the Resonance Method

4.5.1. Basic Formulations for Property Determination

The Young’s modulus of a flat microcantilever beam can be determined from its bending resonant frequency by rearranging Eqn. (3.47):

\[ E = \frac{48\pi^2 \rho L^4}{(\beta_n D)^2 f_b^2} \]  

Similarly, the shear modulus can be determined from its torsional resonant frequency by rearranging Eqn. (3.48):

\[ G = \frac{4\rho L^2 b^2}{i^2 t^2} (f_t)_i^2, \text{ for } i = 1, 3, 5... \]  

The Poisson’s ratio can be calculated directly from the $E$ and $G$ values by rearranging Eqn. (3.6):

\[ \nu = \frac{E}{2G} - 1 \]
For the first bending and torsional resonant frequencies, substituting Eqns. (4.2) and (4.3) into Eqn. (4.4) yields:

\[
\nu = \frac{6\pi^2 L^2}{(\beta_L b^2)} \left( \frac{f_b}{f_t} \right)^2 - 1
\]  

(4.5)

Hence the Poisson’s ratio of a microcantilever beam can be determined directly from the measurements of its bending and torsional resonant frequencies if its length and width are known. If the thickness and the density of the beam are known, \(E\) and \(G\) can be determined using Eqns. (4.2) and (4.3). From the FEM simulations shown in section 3.7, beams of width 20 \(\mu m\) and of thickness less than 1 \(\mu m\) with \(L/b\) ratios greater than 8, Eqn. (4.5) is accurate to within 7 %, thus providing a good approximation of \(\nu\) of thin film microcantilever beams.

4.5.2. Experimental Details

SiN\(_x\)H\(_y\) microcantilevers were fabricated on 3-inch (100)-oriented Si wafers using bulk micromachining techniques, as shown in Fig. 4.4. Due to the high etch rate of low-temperature PECVD SiN\(_x\)H\(_y\) thin films in the KOH solution (see Fig. 4.2), thin layers (5 nm) of thermally evaporated Cr are deposited before and after the SiN\(_x\)H\(_y\) deposition (Fig. 4.4(a)). The Cr layers act as a barrier between the KOH solution and the SiN\(_x\)H\(_y\) thin film during the etching of the substrate. The films were then patterned (Fig. 4.4(b)) and the substrates etched in 40-wt. % KOH aqueous solution at 80 °C until the beams were fully released from the substrate (Fig. 4.4(c)). The Cr layers were then removed in a hydrochloric acid (HCl) solution (Fig. 4.4(d)). Care was taken not to over-etch the Cr bottom layer, thus causing undercut at the beams’ supports. This method resulted in fabricated SiN\(_x\)H\(_y\) microcantilevers with constant moment of inertia along their lengths, thus allowing the application of Eqns. (4.2), (4.3) and (4.5) to determine \(E\), \(G\) and \(\nu\) from the beams’ resonant frequencies, respectively.

The dimensions of the fabricated microcantilevers ranged from 200-300 \(\mu m\) in length, 15-25 \(\mu m\) in width and 200-900 nm in thickness. The thickness of the SiN\(_x\)H\(_y\) beams was obtained using a surface profiler on witness samples that experienced the same deposition event as the fabricated samples. The lengths and widths of the beams were measured using an optical microscope.
The resonant frequencies of the microcantilevers were measured using the experimental setup shown in Fig. 4.5, which is similar to that proposed by Tsai et al. [11]. The piezoelectric stack is used as an impact hammer to excite the beams [11]. The vibrations of the beams were measured using a vibrometer attached to an optical microscope. To eliminate the effects of air damping to the vibration, the samples were mounted inside an evacuated chamber of ~0.1 Torr pressure. The deflection of the beams was measured using a reference-measurement dual laser system, which helps to minimise the measurement of external vibrations, whilst removing the piezoelectric stack resonant vibration frequencies. Though the samples were excited in the vertical direction, torsional vibration modes were still excited due to cross-talk [12]. For each film, up to ten microcantilevers were measured to minimise errors in the measurement.

4.5.3. Results to the Resonance Method

Figure 4.6 shows the frequency response of an $\alpha$-series microcantilever beam deposited at 250 °C with dimensions 205 $\mu$m long, 21 $\mu$m wide and 830 nm thick. The first three bending vibration modes, $f_{b1}$, $f_{b2}$ and $f_{b3}$ are clearly visible, as well as the first torsional vibration mode, $f_{t1}$. The first bending vibration mode occurs at 25.711 kHz while $f_{t1}$ occurs at 490.875 kHz. Substituting these values, as well as the microcantilever beam’s
dimensions and density value (obtained using the QCM method) into Eqns. (4.2), (4.3) and (4.5) yields calculated $E$, $G$ and $\nu$ values equal to 182 GPa, 72.6 GPa and 0.25, respectively. Substituting the calculated $E$ and $G$ values into Eqn. (4.4) yields a $\nu$ value equal to 0.25, which agrees with the $\nu$ value calculated using Eqn. (4.5).

Figure 4.7 shows the calculated $E$, $G$ and $\nu$ values of the $\alpha$ and $\beta$-series thin films as a function of chuck temperature using the resonance method. The density of the films required to determine $E$ and $G$ were obtained from the QCM measurements. The error bars represent the standard deviations in the measurements. The values shown were calculated using the first bending and torsional vibration frequencies. It is evident that $E$ and $G$ of the films increase continuously with increasing deposition temperature. The $\alpha$-series thin films have higher $E$ and $G$ values compared to $\beta$-series films, relative to chuck temperature. The highest $E$ and $G$ values achieved were $193 \pm 7$ GPa and $77 \pm 3$ GPa, respectively, for SiN$_x$H$_y$ films deposited at 300 °C, 600 mTorr and 100 W. This $E$ value is still far below that of bulk polycrystalline Si$_3$N$_4$, which is 304 GPa [27]. The lowest $E$ and $G$ values achieved were $54 \pm 6$ GPa and $22 \pm 3$ GPa, respectively, for SiN$_x$H$_y$ films deposited at 100 °C, 875 mTorr and 75 W.
The $\nu$ values show little difference between the $\alpha$-series and $\beta$-series films. Though the relative errors are large, both sets of films show clearly that $\nu$ increases with increasing chuck temperature for temperatures below 200 °C. For films deposited above 200 °C, the measured Poisson’s ratio appears to be independent with changes in chuck temperature. At these temperatures, a Poisson’s ratio of approximately 0.25 is realised. The highest $\nu$ value achieved was 0.26 ± 0.01 for SiN$_x$H$_y$ films deposited at 250 °C, 600 mTorr and 100 W. This $\nu$ value is close to that of bulk polycrystalline Si$_3$N$_4$, which is 0.27 [27]. The lowest $\nu$ value achieved was 0.20 ± 0.01 for SiN$_x$H$_y$ films deposited at 100 °C, 875 mTorr and 75 W.

Similarly to the $\rho$ measurements presented in Section 4.4, the $E$, $G$ and $\nu$ of $\alpha$ and $\beta$-series films are highly sensitive to process conditions. It is also obvious that bulk silicon nitride values cannot accurately represent the true $E$, $G$ and $\nu$ values of the deposited $\alpha$ or $\beta$-series films. As deposition temperature increases, however, the $E$, $G$ and $\nu$ values of the films approach that of the bulk values.
FIG. 4.7. Resonant method measurements showing (a) $E$, (b) $G$, and (c) $\nu$ versus changes in chuck temperature for $\alpha$ and $\beta$-series films.
4.6. Determination of Young’s Modulus Using Nanoindentation and Microcantilever Beam Bending Methods

4.6.1. Basic Formulations for Property Determination

When an indenter of known dimensions is pressed onto the surface of a specimen, the force on the indenter will increase with increasing penetration depth. If the projected area of the indenter increases with increasing indentation depth (for example, a spherical or pyramidal shaped indenter), this force increases non-linearly. For a purely elastic material, the force induced onto the indenter relative to indentation depth will be the same for both loading and unloading events. For a purely inelastic (i.e. plastic) material, no force will be induced onto the indenter during an unloading event. Generally, materials behave both elastically and inelastically, and the loading-unloading curves of an indented material will look similar to that depicted in Fig. 4.8(a). In this case a SiN$_x$H$_y$ thin film has been indented with a Berkovich tip, which is commonly used in nanoindentation experiments and is illustrated in Fig. 4.8(b) [28]. For the case where plastic deformation occurs, the indentation event will leave an impression in the specimen and its height is equal to the difference in the loading-unloading curve at zero applied load (denoted $h_i$ in Fig. 4.8(a)).

During a loading event, both elastic and plastic deformation occurs. During unloading, the force induced onto the tip by the film is purely elastic, and is due to the relaxation of elastic strains within the material [29]. Hence, from nanoindentation experiments, $E$ is determined from the unloading contact stiffness, $S$, which can be obtained from the slope of the initial portion of the unloading curve [6]:

$$ S = \frac{dP}{dh} \quad (4.6) $$

From the parameter $S$, the reduced modulus $E_r$, which is a measure of the combined elastic response of the specimen and the indenter, can be determined as:

$$ E_r = \frac{\sqrt{\pi}}{2\eta} \cdot \frac{S}{\sqrt{A}} \quad (4.7) $$

which in turn can be related to the Young’s modulus of the sample in:

$$ \frac{1}{E_r} = \frac{1-v_D^2}{E} + \frac{1-v^2}{E_D} \quad (4.8) $$
In the equations above, \( P \) is the applied load, \( h \) is the tip displacement, \( A \) is the contact area, \( \eta \) is a constant depending on the geometry of the indenter, \( \nu \) and \( E \) are the Poisson’s ratio and the Young’s modulus of the specimen, \( \nu_D \) and \( E_D \) are the Poisson’s ratio and the Young’s modulus of the diamond tip. Here, \( \nu_D \) and \( E_D \) are equal to 0.07 and 1140 GPa, respectively, and for the Berkovich indenter tip \( \eta = 1.034 \).

The method described above is known as the Oliver and Pharr [6, 29] method, which is a refinement of the Doerner and Nix [30] approach. The Oliver and Pharr method is now the most commonly used method of analysis to determine \( E \) from nanoindentation loading data [29]. It has recently been shown however that the determination of \( E \) using nanoindentation of a thin film material deposited onto a

![Graph showing load versus tip displacement](image)

**FIG. 4.8.** Nanoindentation of a SiN, H, thin film showing, (a) a typical load versus tip displacement curve, and (b) SEM image of a Berkovich indenter tip [28].
substrate is directly dependent on the thickness of the film and the Young’s modulus of the substrate \( (E_s) \) relative to the Young’s modulus of the film \( (E_f) \). When nanoindentation is performed on a film deposited on a substrate, such that \( E_s > E_f \), the calculated \( E \) using the Oliver and Pharr method increases with increasing indentation depth towards \( E_s \). For the case where \( E_s < E_f \), the calculated \( E \) decreases with increasing indentation depth towards \( E_s \). Previous methods used to negate the effect of the substrate on the calculated \( E_f \) have simply been to neglect all values of \( E_f \) calculated from indentation depths greater than 10% of the film thickness [31, 32]. Jung et al. [5] provides a more elegant approach whereby the contributions from substrate deformation on the calculated \( E_f \) can be accounted for. In this method, \( E_f \) is determined using the following equations:

\[
E = E_s \left( \frac{E_f}{E_s} \right)^{L}
\]

(4.9)

\[
L = \frac{1}{1 + C \left( \frac{h}{t} \right)^D}
\]

(4.10)

where \( h \) and \( t \) are the indentation depth and film thickness, respectively and \( C \) and \( D \) are adjustable coefficients. Equations (4.9) and (4.10) satisfy the boundary conditions: \( h/t \to 0, E = E_f \) and \( h/t \to \infty, E = E_s \). Hence \( E_f \) is determined from multiple indentations carried out over a range of indentation depths. By plotting \( E \) verses \( h \) and fitting the data with Eqns. (4.9) and (4.10), \( E_f \) can be determined.

As outlined in Chapter 3, the Young’s modulus can also be inferred from the deflection of a microcantilever. When a nanoindenter is used to deflect a microcantilever beam, it acts as point load. For a beam with constant moment of inertia through its length, \( E \) of the beam can be calculated by rearranging Eqn. (3.25):

\[
E = \frac{4L^2}{bI^3} \left( \frac{F}{w} \right) (1 - \nu^2)
\]

(4.11)

In this case, \( F/w \) is equal to the slope of the applied force versus displacement curve of the tip of the nanoindenter, and \( L \) is the distance from the base of the beam to the tip of the nanoindenter. Similarly for a tapered beam, \( E \) is calculated by rearranging Eqn. (3.26):

\[
E = \left( \frac{6}{bI^3} \right) \left[ 2 \ln \left(1 + \frac{mL}{t} \right) - \frac{mL(3mL + 2t)}{(t + mL)^2} \right] \left( \frac{F}{w} \right) (1 - \nu^2)
\]

(4.12)
4.6.2. Experimental Details

Nanoindentation was used to determine $E$ of $\alpha$-series thin films using a Hysitron Triboscope III nanomechanical testing instrument, connected to a Digital Instruments atomic force microscope (AFM) with a Berkovich indenter tip. The indentation profile used is shown in Fig. 4.9. A holding time of 10 s at the maximum load was used to eliminate any creep effect on the measurements. For each film, three sets of up to 100 indents were performed with varying indentation loads ranging from 100 µN up to 9000 µN.

For the microcantilever beam bending method used to determine $E$, the SiN$_x$H$_y$ thin films were patterned into microcantilever beams using the same bulk micromachining methods shown in Fig. 4.4, except the Cr layers were not included. The SiN$_x$H$_y$ thin films were thus exposed to KOH during the substrate etching process, which resulted in tapered beams. The microcantilevers were fabricated with widths ranging from 15 µm to 30 µm and in various lengths ranging from 30 µm to 150 µm. The thicknesses of the beams ranged from 300 nm to 1000 nm.

Bending tests of the microcantilever beams were carried out using the Hysitron Triboscope III nanomechanical testing instrument, connected to a Digital Instruments

![FIG. 4.9. Indentation profile used for the nanoindentation experiments.](image)
AFM. The microcantilevers were initially scanned using the AFM, thus creating an image, which enabled precise positioning of the nanoindenter tip on the microcantilevers. The beams were deflected with a conical diamond tip of 20 µm tip radius and a 90° cone angle. The large tip radius ensures that no indentation of the beam occurred during bending. The beams were deflected up to 1 µm and the load versus tip displacement was recorded. For each film, up to 10 microcantilevers of varying dimensions were deflected.

4.6.3. Nanoindentation and Microcantilever Beam Bending Results

The results of the nanoindentation experiments for three α-series thin films deposited at different chuck temperatures are shown in Fig. 4.10, where \( E \) is a function of normalised indentation depth. The dashed reference line in Fig. 4.10 marks the value of the Young’s modulus of the Si substrate, \( E_{Si} \), as determined from nanoindentation experiments. This value is similar to that found in the literature [33]. It is seen that for all three films, \( E \) approaches \( E_{Si} \) with increasing indentation depth. The power-law method proposed by Jung et al. [5] was used to determine \( E \) from the film/substrate composite to the values for the SiN\(_x\)H\(_y\) films. From these plots, the Young’s modulus of the 125 °C, 150 °C and 200 °C deposited films were calculated to be 98 GPa, 131 GPa and 162 GPa, respectively.

Figure 4.11(a) shows an SEM top view image of three released microcantilevers of 25×125 µm in dimensions. It is seen that they are flat and free of any stress gradient. During the wet release process, etching occurred preferentially on selected atomic planes in the Si substrate, with the etching front propagating from the tip to the base of the beams. As the selectivity of the KOH etchant is not infinite, i.e. some slow etching also occurred on the SiN\(_x\)H\(_y\) thin film, a thickness gradient resulted along the length of the beam [11]. The exact amount of tapering, i.e. the thickness at the base and the tip of the beam was determined using SEM. An SEM image showing the thickness of a microcantilever beam near its base is illustrated in Figure 4.11(b). Care was taken not to etch the samples for too long in an attempt to minimise the amount of undercut at the base of the beam.

Figure 4.12 illustrates the applied load of the nanoindenter tip relative to its displacement on an α-series microcantilever beam deposited at 300 °C. The curve is linear, indicating that the deformation of the beam was totally within the elastic regime.
As the beams were tapered, the slope of the curve in Fig. 4.12 is equivalent to the \((F/w)\) term in Eqn. (4.12). The actual dimensions of the deflected beam were 19 µm in width, 38 µm in length and 900 nm in thickness at its base and 830 nm at its end. The thickness variation of the beam across its length was determined using SEM. In this particular example, the length from the base of the beam to the loading point was 32 µm. Using the calculated \(v\) value of the film determined using the resonance method, the Young’s modulus of this beam was determined using Eqn. (4.12) to be 193 GPa.

The average \(E\) values for \(\alpha\)-series thin films calculated from microcantilever beam bending tests and nanoindentation measurements are shown in Fig. 4.13. It is evident that the two methods, as well as the \(E\) values calculated using the resonance method, exhibited excellent agreement. The Young’s modulus was found to increase rapidly with increasing chuck temperature. The Young’s modulus appeared to reach saturation at above 200 °C. The highest \(E\) achieved was 195 GPa at the deposition temperature of 300 °C. For the film deposited at 100 °C, \(E\) was measured to be 83 GPa, ~43% of that of the film deposited at 300 °C.

**FIG. 4.10.** Normalised nanoindentation data of three \(\alpha\)-series thin films deposited at varying chuck temperatures. On the x-axis, \(h/t\) is the ratio of indentation depth \((h)\) to film thickness \((t)\).
FIG. 4.11. SiN$_x$H$_y$ thin film microcantilever beams; (a) SEM image of three 25 µm wide, 125 µm long microcantilevers, and (b) SEM image illustrating the cross-section thickness of a microcantilever near its base.
FIG. 4.12. Load-deflection curve for an $\alpha$-series microcantilever beam deposited at 300 °C.

FIG. 4.13. Effect of deposition temperature on the Young’s modulus of $\alpha$-series thin films measured using microcantilever
Chapter 4 - Properties of Silicon Nitride Thin Films

4.7. Determination of Density from Microcantilever Beams Using an Atomic Force Microscope

4.7.1. Experimental Details

The AFM was utilised to gain another measurement of the density of the films. As shown in Fig. 4.14, the AFM was used in a similar manner to the vibrometer and $\rho$ of the beams was measured using the resonance method. In this instance, however, $E$ is a known parameter (determined from microcantilever beam bending method). With the AFM set to tapping mode, the microcantilever beams were mounted in the AFM head, replacing the standard AFM tapping tip. A laser was focused onto a microcantilever which was mechanically excited with a sweep. The reflection of the laser off the microcantilever was monitored via a detector, which allowed for the deflection at the free end of the cantilever beam to be measured continuously.

FIG. 4.14. The AFM head setup for the cantilever beam resonant frequency measurement. The piezoelectric crystal drives the cantilevers in the $z$-direction and their deflections are measured by the detector.
4.7.2. Results to the Density Measurements

Figure 4.15 shows the effect of chuck temperature on the density of α-series thin films, as determined from resonating tapered microcantilever beams measured using the AFM. For comparison, \( \rho \) values obtained using the QCM method are also plotted. As Eqn. (4.2) is only applicable for a beam whose thickness is uniform across its length, the density of the beams was determined using the commercial finite element modelling package ANSYS. A modal analysis for a simple microcantilever beam of variable thickness across its length was modelled using shell elements. Using the measured first order bending resonant frequencies of microcantilever beams obtained from the AFM, as well as their measured \( E \) values determined from microcantilever beam bending tests, \( \rho \) could be extracted. The results presented in Fig. 4.15 show that both the QCM method and the AFM method shown excellent agreement. It is seen that \( \rho \) of the films increased continuously with increasing deposition temperature. The highest \( \rho \) achieved was 2.9 g/cm\(^3\) at the deposition temperature of 300 °C. For the film deposited at 100 °C the density was estimated as 1.8 g/cm\(^3\), ~64\% of that of the film deposited at 300 °C.

![Figure 4.15: Effect of chuck temperature on density of α-series thin films, as determined from resonating, tapered microcantilever beams. The resonant modes of the beams were measured using an AFM. For comparison, density measurements obtained using the QCM method are also plotted.](image-url)
It has been reported in the literature that the density of PECVD Si$_{x}$N$_{y}$H$_{z}$ films are dependent on film thickness [34]. A variation in $\rho$ through the thickness of an amorphous CVD deposited thin film suggests that during film growth, a variation in process conditions has occurred. For example, the temperature of the substrate may increase during deposition due to ion bombardment. Also, an increase in chamber pressure during deposition can result from the out-gassing of chamber deposits. This can occur if the chamber is not cleaned before deposition. For very dirty chambers, the increase in pressure can be significant.

To determine whether the mechanical and physical properties of PECVD Si$_{x}$N$_{y}$H$_{z}$ thin films are dependent on film thickness, the density of three $\alpha$-series films of different thicknesses, all deposited at the same temperature of 200 °C, were measured. Through microcantilever beam bending measurements, Young’s modulus measurements were also carried out on three $\alpha$-series thin films of different thicknesses, also deposited at the same temperature of 200 °C. It is important to note that for these film depositions, as well as for all PECVD Si$_{x}$N$_{y}$H$_{z}$ depositions carried out in this study, the PECVD chamber was thoroughly cleaned before each deposition event. Figure 4.16 shows the $E$ and $\rho$ measurements as a function of film thickness. The $E$ and $\rho$ values were measured from bending and resonating microcantilever beams, respectively. The resonant frequencies were measured using the AFM. It is evident that the variation of $E$ and $\rho$ with changes in film thickness is within the error of the measurement, which suggests that during each deposition event the variation in process conditions is negligible.

4.8. Determination of Residual Stress Using the Substrate Curvature Method

4.8.1. Basic Formulations for Property Determination

The residual stress in a film on a film-substrate assembly was determined using the curvature measurement method. The relationship between the residual stress in the film and the induced curvature of the substrate is expressed in the Stoney formula [35]:

$$\sigma = \frac{E_s t_s^2}{6R(1-v_s) t_f}$$  \hspace{1cm} (4.13)$$

where $E_s$ and $v_s$ are the Young’s modulus and the Poisson’s ratio of the substrate, respectively, $t_s$ and $t_f$ are the thickness of the substrate and film, respectively and $R$ is the radius of curvature of the assembly. This method is true only for situations where $t_s>>t_f$. 
FIG. 4.16. The dependence of (a) \( \rho \), and (b) \( E \) of 200 °C deposited \( \alpha \)-series thin films with changes in film thickness. The density was calculated from the beams’ first order bending resonant modes, measured using an AFM, and \( E \) was measured from microcantilever beam bending tests.
and determines the average residual stress in the film over a large deposition area (typically many cm$^2$) [18]. Large deposition areas are required to minimise the error in $R$ which is usually measured using an optical profilometer or parallel laser beam set-up. Consequently, the method does assume the properties of the film are uniform over the measured area.

4.8.2. Experimental Details

The radius of curvature of SiN$_x$H$_y$ coated substrates was determined from the divergence/convergence of three parallel laser beams created using an optically flat half-silvered 9.78 mm thick glass plate shown in Fig. 4.17. ND filters were positioned before the beam splitter in order to decrease the intensity of the laser. The sample was positioned with the film facing the laser beams. This setup resulted in convergence of the laser beams for a tensile stressed film and divergence of the laser beams for a compressively stressed film.

![Experimental setup](image)

**FIG. 4.17.** Experimental setup used to determine $\sigma_0$. The sample is positioned so that the parallel laser beams reflect off the substrate on the same side that the film is deposited on. In this case, the sample creates a divergence of the laser beams due to the compressive nature of the film.
The substrate’s radius of curvature can be determined using the following equation [17]:

\[ R = \frac{2L}{\cos\theta} \frac{D}{\Delta D} \]  \quad (4.14)

where \( L \) is the distance from the CCD camera to the sample, \( \theta \) is the angle of incidence, \( D \) is the interspot distance of the parallel beams and \( \Delta D \) is the change in the interspot distance of the reflected beams at the CCD camera. In this particular setup, \( D \) was equal to 6.5 mm. Double-sided polished (100)-oriented Si wafers of thickness 100 \( \mu m \) were used. Prior to SiN\(_x\)H\(_y\) deposition the wafers were dipped in an HF solution to remove the native oxide layer and then cleaved into rectangles of size 60 mm \( \times \) 5 mm. The curvatures of the cleaved Si samples were then measured using the apparatus shown in Fig. 4.17 to ensure they were optically flat. Three cleaved wafers were positioned in the chamber during each deposition event to determine the error in the measurement.

4.8.3. Results to the Substrate Curvature Method

Figure 4.18 shows the effect of chuck temperature on the residual stresses of the \( \alpha \) and \( \beta \)-series films. The standard deviation errors of the measurements are smaller in magnitude than the size of the symbols used to represent the measurement points and thus

![Graph showing residual stress measurements](image)

FIG. 4.18. Residual stress measurements of \( \alpha \) and \( \beta \)-series SiN\(_x\)H\(_y\) thin films determined using the substrate curvature method. Negative \( \sigma_0 \) values represent compressive stress.
are not shown. It is seen that the residual stresses increase with increasing deposition temperature for both $\alpha$- and $\beta$-series films. The magnitude of the stress in $\alpha$-series films is greater than in $\beta$-series films whereas $\beta$-series films have higher tensile stress. The highest compressive residual stress achieved was 133 MPa for SiN$_x$H$_y$ films deposited at 100 °C, 600 mTorr and 100 W. The highest tensile residual stress achieved was 59 MPa for SiN$_x$H$_y$ films deposited at 150 °C, 875 mTorr and 75 W.

4.9. Determination of Young’s Modulus, Residual Stress and Poisson’s Ratio

Using the Bulge Test

4.9.1. Basic Formulations for Property Determination

For square and rectangular thin-film membranes with zero residual stress that are constrained around the perimeter, the bulge height at their central point of the square is related to the applied pressure by Eqns. (3.27) and (3.28). If both a square and an infinite rectangular membrane are fabricated on the same substrate and from the same thin film and that they are both subjected to the same uniform pressure, then Eqns. (3.27) and (3.28) may be combined to yield:

$$f(\nu)(1+\nu) = \frac{4}{3} \left( \frac{w_r}{w_s} \right)^3 \left( \frac{a_s}{a_r} \right)^4$$

(4.15)

Furthermore, if the membranes are designed such that $a_s = a_r$, then Eqn. (4.15) simplifies to:

$$f(\nu)(1+\nu) = \frac{4}{3} \left( \frac{w_r}{w_s} \right)^3$$

(4.16)

Equation (4.16) demonstrates that the $w_r/w_s$ ratio is solely determined by the Poisson’s ratio of the film, and is independent of the applied pressure $P$.

For the case where a membrane contains a tensile residual stress $\sigma_0$, the relationship between the applied pressure and the central point bulge height of a square and an infinite rectangular membrane are given by Eqns. (3.34) and (3.35). As mentioned in section (3.4), it is evident that for large bulge heights, Eqns. (3.34) and (3.35) are reduced to Eqns. (3.27) and (3.28), respectively, since the linear terms become negligible. Hence, if a square membrane and an infinite rectangular membrane of the same dimension $a_s = a_r$ are fabricated on the same substrate and from the same thin film having
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the same residual tensile stress $\sigma_0$, and are tested under the same uniform pressure, then $\nu$ can be calculated from the ratio of $w_r/w_s$ at high pressures using Eqn. (4.16). However, for thin films with relatively low failure strengths, such as low-temperature PECVD SiN$_x$H$_y$ films, large bulge heights cannot be achieved prior to failure. In such cases, $\nu$ needs to be determined by measuring $w_r$ and $w_s$ over a range of pressures, and curve-fitting the data to the following equations:

$$P = A_{1s} w_s + A_{2s} w_s^3$$

$$P = A_{1r} w_r + A_{2r} w_r^3$$

where $A_{1s}, A_{2s}, A_{1r},$ and $A_{2r}$ are fitting parameters. Referring to Eqns. (3.34) and (3.35), it is evident that:

$$A_{1s} = c_s \frac{\sigma_0 t}{a_s^2}$$

$$A_{2s} = \frac{f(\nu)Et}{a_s^3(1-\nu)}$$

$$A_{1r} = c_r \frac{\sigma_0 t}{a_r^2}$$

$$A_{2r} = \frac{4Et}{3a_r^3(1-\nu^2)}$$

For the special case when $a_s = a_r$, dividing $A_{2s}$ by $A_{2r}$ yields:

$$f(\nu)(1+\nu) = \frac{4}{3} \left( \frac{A_{2s}}{A_{2r}} \right)$$

Using Eqn. (4.20), $\nu$ can be determined directly from the $A_{2s}/A_{2r}$ ratio, and is independent of the initial tensile residual stress. This implies that Eqn. (4.20) is applicable for both stress-free membranes and membranes containing tensile residual stress. Equation (4.19) can also be used to determine $E$ and $\sigma_0$ if the membrane dimensions are accurately known. By dividing $A_{1s}$ by $A_{1r}$, the following relationship can also be derived:

$$\left( \frac{A_{1s}}{A_{1r}} \right) = \frac{c_s \left( \frac{\sigma_0 t}{a_s^2} \right)}{c_r \left( \frac{\sigma_0 t}{a_r^2} \right)} = \frac{c_s}{c_r} = 1.6965$$

Equation (4.21) will hold true for any thin film material tested so long as $a_s = a_r$ and the film thickness and residual stress is uniform. Consequently, Eqn. (4.21) can be used as a
measure of possible experimental errors associated with thin film uniformity and membrane dimensions.

The analysis procedure discussed above can be used to determine \( \nu \), \( E \) and \( \sigma_0 \) for any thin film membrane material. However, Vlassak and Nix [14] have suggested that Eqns. (3.27) and (3.28) become invalid for \( \sigma_0 \) values above 500 MPa. In a recent study, Mitchell et al. [36] were unable to determine \( \nu \) using the bulge test for 3C-SiC films that were known to exhibit residual tensile stress greater than 300 MPa. In the present study, only tensile low stress thin films have been used for all bulge test experiments.

4.9.2. Experimental Details

Similar to the bulk micromachining processes described above, square and rectangular SiN\(_x\)H\(_y\) thin film membranes were patterned and released from (100)-oriented Si substrates. As the bulge test is only applicable to tensile stressed thin films, an \( \alpha \)-series film deposited at 205 °C and a \( \beta \)-series film deposited at 125 °C are fabricated into square and rectangular membranes. A fully-fabricated membrane is shown schematically in Fig. 4.19(a). CF\(_4\) plasma was used to dry etch the top SiN\(_x\)H\(_y\) film, followed by wet etching of the Si substrate in 40-wt. % KOH solution at 80 °C. As expected, some etching of the SiN\(_x\)H\(_y\) films did occur in the KOH solution, however, note that the samples were completely immersed in the KOH solution, and etching was terminated immediately on removal of the Si substrate. In order to accurately determine \( E \) and \( \sigma_0 \) of the thin films (see Eqn. (4.19)), their thickness needed to be measured after membrane release. This was achieved by wet etching an area of the SiN\(_x\)H\(_y\) thin film that was still attached to the substrate and using a surface profiler to measure the SiN\(_x\)H\(_y\) step height.

Multiple pairs of square and rectangular SiN\(_x\)H\(_y\) membranes were fabricated on 3-inch Si wafers. The square membranes were approximately 1900×1900 µm in size, and the rectangular membranes had dimensions of approximately 1900×9000 µm. For each set of square and rectangular membranes, the variation in \( a_r \) relative to \( a_s \) was measured to be less than ±0.2%.

Bulge height measurements of the pressurized thin film membranes were carried out using a Zygo Maxim-GP optical profilometer, and the experimental setup is shown in Fig. 4.19(b). The samples were mounted on the pressure chamber with an adhesive
epoxy, and the chamber was pressurised using high purity nitrogen gas. The chamber pressure was monitored using a temperature compensated and calibrated pressure sensor, and all measurements were taken at room temperature. A light source of 673 nm wavelength was used to produce three-dimensional images of the bulge shapes. Cross-sections of these images were plotted in order to determine the maximum bulge height under the uniform applied pressure.

4.9.3. Results to the Bulge Test Measurements

Figure 4.20(a) and (b) shows examples of the interference fringes observed with the optical profilometer due to the convex curved surfaces of a square membrane and an infinite rectangular membrane, respectively, under a uniform applied pressure. Figure 4.20(c) plots the bulge cross-sectional profiles for a set of square and rectangular membranes fabricated on the same silicon substrate from an α-series thin film deposited
FIG. 4.20. Optical profilometer measurements with 673 nm illumination showing: (a) interference fringes of the middle section of a square membrane under a uniform applied pressure, (b) interference fringes of the middle section of an infinite rectangular membrane under a uniform applied pressure, and (c) cross-sectional bulge profiles through the centre (see insert) of a square and a rectangular membrane under 300 Pa of pressure for a 305 nm thick PECVD α-series thin film deposited at 205 °C.
at 205 °C. The applied pressure corresponding to the two profiles was 300 Pa. The inset illustrates a top view of the membranes, indicating the positions of the bulge cross-sectional profiles. It is evident that under the same applied pressure, the square membrane produced a lower bulge profile compared to the infinite rectangular membrane, which is due to the greater degree of constraint to bulging that result from the fixed edges of the square membrane.

It was found that for rectangular membranes, the bulge profile across the width at any pressure was accurately described by a parabola, with correlation coefficients better than 0.999. This observation agrees with Kalkman et al. [20] who found the same parabolic correlation for rectangular free-standing tungsten thin films. However, for the square membranes, it was found that a fourth order polynomial was well-suited to describing the bulge profile across the centre at any pressure.

Figure 4.21 shows the relationship between bulge height and applied pressure for two pairs of membranes fabricated from a β-series thin film deposited at 125 °C (Fig. 3(a)) and an α-series thin film deposited at 205 °C (Fig. 3(b)). The data were fitted using Eqns. (4.17) and (4.18), with the fitting parameters summarized in Table 4.2. Substituting the $A_{2s}/A_{2r}$ ratios for this particular set of membranes into Eqn. (4.20) yields a Poisson’s ratio of 0.24 for the 125 °C deposited film and 0.25 for the 205 °C deposited film, when using the Vlassak and Nix expression for $f(\nu)$ in Table 3.1 [14]. The SiN$_x$H$_y$ thicknesses of these two sets of membranes were measured to be 680 nm and 305 nm for the 125 °C and 205 °C deposited films, respectively. These thickness values and the parameters listed in Table 4.2 were substituted into Eqn. (4.19) to determine the Young’s modulus, $E$, and residual stress, $\sigma_0$, for each film. The calculated $E$ values for this particular set of membranes were thus calculated to be 82 GPa and 146 GPa for the 125 °C deposited film and the 205 °C deposited film, respectively. The corresponding $\sigma_0$ values were calculated to be 4.3 MPa and 32 MPa for the 125 °C deposited film and the 205 °C deposited film, respectively.

<table>
<thead>
<tr>
<th>SiN$_x$H$_y$ Deposition Temp. (°C)</th>
<th>$A_{1s}$ (kPa/µm)</th>
<th>$A_{2s}$ (kPa/µm$^3$)</th>
<th>$A_{1r}$ (kPa/µm)</th>
<th>$A_{2r}$ (kPa/µm$^3$)</th>
</tr>
</thead>
<tbody>
<tr>
<td>125</td>
<td>1.094×10^{-2}</td>
<td>1.630×10^{-4}</td>
<td>6.462×10^{-3}</td>
<td>9.490×10^{-5}</td>
</tr>
<tr>
<td>205</td>
<td>3.426×10^{-2}</td>
<td>1.259×10^{-4}</td>
<td>2.114×10^{-2}</td>
<td>7.291×10^{-5}</td>
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</table>
FIG. 4.21. Measured relationship between applied pressure and bulge height for square and rectangular $\beta$ and $\alpha$-series thin film membranes deposited at; (a) 125 °C, and (b) 205 °C, respectively. The solid lines represent fits to the data using Eqns. (4.17) and (4.18) and the fitting parameters listed in Table 4.2.
Figure 4.22 shows the measured relationships between $w_r$ and $w_s$ extracted from the fits shown in Fig. 4.21 for the two films. The dashed lines in both plots are linear, representing the $w_r/w_s$ ratio under the condition of zero residual stress: that is $\sigma_0=0$, as represented by Eqn. (4.16). It needs to be pointed out that the slope of all the curves in Fig. 4.22 is always greater than unity. This is due to the bulge height of an infinite rectangular membrane being always greater than that of a square membrane. From the data in Fig. 4.22, it is evident that the relationship between $w_r$ and $w_s$ is non-linear for low bulge heights, with the experimental curves deviating from the reference dashed lines. This implies that the films contain residual tensile stresses, which act to restrict bulging of the membranes.

The curve for the 205 °C deposited film deviates further from the reference line in comparison to the 125 °C deposited film, which is consistent with the fact that the 205 °C deposited film had a higher tensile stress than the 125 °C deposited film. The experimental curves appear above the reference lines because the residual tensile stress is more effective in resisting bulging for the square membrane than for the rectangular membrane, as indicated in the relative values of $c_r$ and $c_s$. It is also evident that the experimental curves for increasing $w_r$ and $w_s$ values are approaching linearity (that is, merging with the dashed lines), as would be expected according to Eqns. (3.34) and (3.35).

Substituting $A_{1s}$ and $A_{1r}$ values from Table 4.2 into Eqn. (4.21) yields $c_s/c_r$ ratios of 1.693 and 1.701 for the 125 °C deposited film and the 205 °C deposited film, respectively. These values are very close to the theoretical $c_s/c_r$ ratio of 1.6965. It is expected that the small error is due mainly to the variation in $a_s$ relative to $a_r$. Due to the large exponents in Eqn. (4.19) a 0.2 % error in $a_s$ relative to $a_r$ will cause an error of approximately ±0.01 in the calculated Poisson’s ratio. Fabricating larger sized membranes will help to reduce the relative error between $a_s$ and $a_r$, as dimension variations created during patterning processes will become less significant.

In the course of this study, three sets of square and rectangular membranes were measured for both $\beta$ and $\alpha$-series thin films deposited at 125 °C and 205 °C, respectively, and the average value of Poisson’s ratio, $\nu$, was calculated using Eqn. (4.20). For $\beta$-series films deposited at 125 °C, $\nu$ was calculated to be 0.23 ± 0.02. In this case, the $f(\nu)$
FIG. 4.22. The solid lines represent the measured relationship between maximum bulge height for square and rectangular $\beta$ and $\alpha$-series thin film membranes deposited at; (a) 125 °C, and (b) 205 °C, respectively. The dashed lines were calculated neglecting the effects of residual stress, as represented by Eqn. (4.16).
expression proposed by Vlassak and Nix was used for this calculation [14]. The indicated
errors represent the standard deviation in the experimental results. The calculated

ν

value is close to the value of 0.22 ± 0.01 obtained using the resonance method for the same
film. For α-series films deposited at 205 °C ν was calculated to be 0.25 ± 0.01, which is
the same as the ν value obtained for the α-series films deposited at 200 °C using the
resonance method. Table 4.3 summarises the calculated ν values for the two SiNₓHᵧ films
using the different f(ν) expressions listed in Table 3.1. The models of Pan et al. and
Vlassak and Nix appear to indicate a slightly better agreement between each other in
comparison to the Maier-Schneider et al. model.

The average calculated E and σ₀ values for each SiNₓHᵧ film are shown in
Table 4.4. In this case, the ν values for each film required to calculate E and σ₀ were
determined using the Vlassak and Nix expression for f(ν) [14]. For the β-series film, the
calculated E and σ₀ values are comparable to the respective values 78 ± 4 GPa and
14 ± 2 MPa determined using the resonant method and substrate curvature measurements
for films with the same deposition conditions. For the α-series film, the calculated E and
σ₀ values are comparable to the respective values 168 ± 6 GPa and 6 ± 1 MPa determined
using the resonant method and substrate curvature measurements for films deposited at
200 °C.

<table>
<thead>
<tr>
<th>Authors</th>
<th>f(ν)</th>
<th>ν</th>
</tr>
</thead>
<tbody>
<tr>
<td>Pan et al. (1990) [37]</td>
<td>1.37(1.446 - 0.427 ν)</td>
<td>0.24</td>
</tr>
<tr>
<td>Vlassak and Nix (1992) [14]</td>
<td>(0.800 + 0.062 ν)³</td>
<td>0.23</td>
</tr>
<tr>
<td>Maier-Schneider et al. (1995) [38]</td>
<td>1.994(1 - 0.271 ν)</td>
<td>0.21</td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>Authors</th>
<th>f(ν)</th>
<th>ν</th>
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<tbody>
<tr>
<td>Pan et al. (1990) [37]</td>
<td>1.37(1.446 - 0.427 ν)</td>
<td>0.26</td>
</tr>
<tr>
<td>Vlassak and Nix (1992) [14]</td>
<td>(0.800 + 0.062 ν)³</td>
<td>0.25</td>
</tr>
<tr>
<td>Maier-Schneider et al. (1995) [38]</td>
<td>1.994(1 - 0.271 ν)</td>
<td>0.23</td>
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</tbody>
</table>

<table>
<thead>
<tr>
<th>Film</th>
<th>Chuck Temperature (°C)</th>
<th>E (GPa)</th>
<th>σ₀ (MPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>β-series</td>
<td>125</td>
<td>79 ± 4</td>
<td>5.1 ± 1.2</td>
</tr>
<tr>
<td>α-series</td>
<td>205</td>
<td>151 ± 6</td>
<td>29 ± 3</td>
</tr>
</tbody>
</table>
4.10. Discussion

The resonance method, nanoindentation and microcantilever beam bending methods used to determine the Young’s modulus of thin film materials show excellent agreement for the $\alpha$ and $\beta$-series SiN$_x$H$_y$ thin films measured in this study. The resonance method and quartz crystal microbalance method used to measure the density of the films also showed excellent agreement. This study has also shown that the modified bulge test method provides similar Young’s modulus, residual stress and Poisson’s ratio values for slightly tensile thin films compared to the other methods presented.

The experimental evidence presented above demonstrates the Young’s modulus and shear modulus of the SiN$_x$H$_y$ films exhibited similar dependencies on deposition temperature. The increasing trend of $E$ with increasing deposition temperature for SiN$_x$H$_y$ thin films was also observed by Winchester et al. [17] and Martyniuk [18]. It is seen that the density of the films also increases with increasing deposition temperature. Figure 4.23(a) plots the Young’s modulus against density for the films as determined using the resonance method. It is evident that the Young’s modulus increases with increasing density of the films.

The density measurements presented in Fig. 4.3 determined using the quartz crystal microbalance method shows that $\rho$ may decrease by 44% (for the $\alpha$-series films) with decreasing deposition temperature from 300 °C to 100 °C. This is too large to be accounted for by any minor variations in the structure of the films. Low-temperature deposited PECVD SiN$_x$H$_y$ thin films have been previously described as being porous in structure, and the amount of porosity in the films increases with decreasing deposition temperature [39]. Besser et al. [40] suggested that at higher substrate temperatures, there is a decrease in film porosity due to higher surface mobility of the deposited species during the deposition process. With such large variation in density, the porosity is expected to be massive. The current study disagrees with this claim. The films deposited are found to be perfectly dense without any sign of pinholes or physical porosity [41]. In this regard, the large variation in density can only be attributed to significant variations in the chemical and molecular structure of the films, i.e. the films deposited at high and low temperatures are chemically different. Decreasing the deposition temperature decreases the surface mobility of the deposited species, as well as the completeness of the film deposition reactions [42]. This is expected to increase the retention of hydrogen from the reactant species in the films, resulting in higher concentrations of Si-H and N-H
FIG. 4.23. Results obtained from the resonance method, (a), (b) and (c) plot $E$ versus $\rho$, $\nu$ versus $\rho$ and $\nu$ versus $E$ of $\alpha$-series and $\beta$-series Si$_{N_x}$H$_y$ thin films, respectively.
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bonds [43]. Larger concentrations of hydrogen in a film will result in a lower density, whilst higher concentrations of Si-H and N-H bonds in replacement of Si-N bonds will result in a lower Young’s modulus as the strength of the Si-N bond is greater compared to those of the Si-H and N-H bonds [44].

Changes in the atomic composition of PECVD SiN$_x$H$_y$ thin films with changes in deposition temperature have been reported by Martyniuk [24] through Heavy-Ion Elastic Recoil Detection Analysis (HIERDA). For films deposited using the same process equipment as in this study, the concentration of hydrogen decreased from 35% to 21% with increasing deposition temperature from 100 ºC to 300 ºC. In comparison, the concentration of silicon and nitrogen increased from 21% to 31% and 38% to 47%, respectively, over the same deposition temperature range. Such large changes in atomic composition of the films help to explain the large variations in the mechanical and physical properties measured in this study.

The $E$ and $\rho$ values for the $\beta$-series films are lower than those for the $\alpha$-series films. This is attributed to the higher chamber pressures and lower RF powers for the $\beta$-series films. A higher pressure decreases the mean free path of the reactants, resulting in lower energies of the ions as they accelerate towards the substrate. A lower RF power decreases the intensity of ion bombardment on the surface of the substrate. During film growth, the breaking of the Si-H and N-H bonds is directly influenced by the extent of ion bombardment [42]. High retention of H in the structure leads to formation of films of low density and low Young’s modulus. In this regard, a decrease in ion bombardment intensity has similar effects on $E$ and $\rho$ as does decreasing the deposition temperature. This appears to be consistent with the work carried out by Claassen et al. [42], who found that the percentage of hydrogen in PECVD silicon nitride films deposited at 300 ºC increased with increasing chamber pressure, whilst the percentage of nitrogen decreased.

The Poisson’s ratio values obtained using the resonance method and the modified bulge test method agree with Ziebart et al. [45] who report a $\nu$ value of 0.253 for PECVD SiN$_x$H$_y$ thin films obtained from buckled membranes. These $\nu$ values obtained for PECVD SiN$_x$H$_y$ thin films are lower than the measurement of 0.28 obtained by Vlassak and Nix [14] for low-pressure chemical vapour deposited (LPCVD) silicon nitride films. The relationships between the $\nu$ and $\rho$ values and between $E$ and $\nu$ values are shown in Figures 4.23(b) and (c), respectively. It is seen that $\nu$ increases with increasing $\rho$. Similar
observations were also made by Alexandre et al. [46] on hot isostatically pressed zirconium nitride materials, Maruyama et al. [47] on high-density isotropic graphites, and Artamonov et al. [48] on porous iron. The change in the measured $\nu$ with changes in $\rho$ in these studies was attributed to the changes in porosity of the materials. In this sense the measured $\nu$ is only the apparent Poisson’s ratio of the porous materials, instead of the true physical property of the solid material. The films used in this study contained no physical porosity. In this regard, the large variation in $\nu$ can only be attributed to the large variations in atomic compositions of the films deposited under different conditions. The variations in atomic composition also affect the density of the films, as discussed. The exact explanation of the relation between atomic composition and Poisson’s ratio is yet to be established.

From the substrate curvature results, it is clear that increasing the chuck temperature increases the tensile nature of the SiN$_x$H$_y$ thin films. It is also clear that decreasing the RF power and increasing the chamber pressure also increases the tensile nature of the film. This trend is contrary to the $\rho$ and $E$ parameters. For example, increasing the chuck temperature increases $E$, however decreasing RF power and increasing chamber pressure decreases $E$. Classen et al. [42] suggests that tensile stress is caused by densification of the film due to hydrogen desorption, which increases with increasing chuck temperature. To explain how a tensile stress is produced by decreasing RF power and increasing chamber pressure however, another mechanism must be proposed. Winchester [22] suggests even though both the temperature of the deposited species and the extent of ion bombardment affects hydrogen desorption, it is the temperature which is the dominating factor. In actual fact, increasing ion bombardment (i.e. increasing RF power and decreasing chamber pressure) increases the packing and densification of a film, resulting in an increase in the compressive stress. This agrees with the results found in this study. The denser $\alpha$-series films were found to be more compressive than the $\beta$-series films for a given chuck temperature.

4.11. Summary

This chapter has presented Young’s modulus, density, Poisson’s ratio and residual stress measurements of SiN$_x$H$_y$ thin films deposited under varying process conditions. A range of methods have been used to determine the properties of the films, and show a strong agreement in the values of the measured parameters, as well as a strong
dependence of $E$, $\rho$, $\nu$ and $\sigma_0$ with changes in chuck temperature, chamber pressure and RF power. In general, increasing the chuck temperature will increase the $E$, $\rho$, and $\nu$ values of the deposited films. Higher chuck temperatures also resulted in tensile films being deposited. For a given chuck temperature, increasing the RF power and decreasing the chamber pressure resulted in higher $\rho$, $E$ and $\nu$ values, however increased the compressive stress in the films.

References


Chapter 5

Filter Design and Optimisation

In this chapter, a new MEMS-based short-wave infrared (SWIR) tunable Fabry-Pérot (FP) filter design is presented, which endeavours to solve the problem of fabricating a filter array with high fill factor and large tuning range. The filter is optimised both optically and electro-mechanically. The fabrication processes were designed and implemented such that the filter can be monolithically integrated with Hg$_{1-x}$Cd$_x$Te. In this work however, filter fabrication was carried out on (100)-oriented Si substrates. The transmission spectrum of fabricated fixed cavity filters and the behaviour of tunable filters when subjected to electrostatic loads are also presented.

5.1. Review of MEMS-Based Fabry-Pérot Filters

The X-style MEMS-based SWIR tunable FP filter fabricated by Dell et al. [1] (shown in Section 1.2) is a common filter design which has been employed for the fabrication of both short-wave (1-3 $\mu$m) and mid-wave (3-5 $\mu$m) IR (MWIR) detectors [1-5]. Using this design, Dell et al. [1] managed to tune from 1.8-2.2 $\mu$m for the SWIR designed filters and from 3.6-4.5 $\mu$m for the MWIR designed filters. In both cases, distributed Bragg reflector (DBR) mirrors were utilised, and consisted of alternate layers of thermally evaporated germanium and silicon monoxide thin films. For the SWIR and MWIR filters, the actuation voltages required to obtain these tuning ranges were 7.6 V and 17 V, respectively.

The finesse of a cavity ($F_c$) is a parameter often used to describe the resolution of an FP filter. It is defined using:
\[ F_c = \frac{\text{FSR}}{\text{FWHM}} \]  

(5.1)

where FSR is free spectral range and FWHM is the full-width at half-maximum of the resonant peaks (also known as the line width). The line widths of the SWIR filters fabricated by Dell et al. [1] were less than 40 nm, resulting in \( F_c \) equal to more than 10. The required cavity finesse can be large or small, depending on the application. For thermal imaging applications, this value of \( F_c \) is acceptable, however for spectroscopy applications, smaller line widths are required. Shah et al. [2] employed the X-style for the fabrication InP-based FP filters, which were theoretically able to tune from 1.475-1.550 \( \mu \text{m} \) with an \( F_c \) value greater than 180. Though the FSR is small compared to filters fabricated by Dell et al. [1], the cavity finesse is much greater and is a requirement for wavelength division multiplexing applications.

Huang et al. [5] report on modelling the deformation behaviour of four different FP filters when subjected to electrostatic loading. All four filters were designed such that the top mirror was positioned on top of supporting flexures, as shown in Fig. 5.1. This method is common in MEMS-based FP filter designs [1-7]. Figure 5.1(a) illustrates the X-style filter, as well as three other filter designs. Their associated mirror displacements versus applied bias voltage curves are shown in Fig. 5.1(b). For these filters, the resonant cavity size was 1.6 \( \mu \text{m} \) and the top mirror size was 100×100 \( \mu \text{m} \). The top mirrors in designs (i), (ii) and (iii) are supported by two fixed-fixed beams normal to each other. For these three designs the beams are subjected to tensile load when the mirror is displaced downwards. For design (iv) the arms are subjected to flexural deformation when the mirror is displaced. In comparison, the relationship between the stress in the arms and the displacement of the mirror is more linear for design (iv) and more parabolic for the other three designs. Due to their non-linear nature, larger displacements of the top mirror are achieved with designs (i), (ii) and (iii) compared to design (iv) before pull-in occurs. The tuning curve for design (iv) is undesirable as the control of the mirror displacement is difficult at higher voltages, however it does have the advantage of a higher fill factor compared to the other three designs. The top mirror deformation in designs (i), (ii) and (iii) in Fig. 5.1 is generally insensitive to residual stress (\( \sigma_0 \)) in the flexures provided the stress is tensile or less than the Euler buckling limit if the stress is compressive [5]. Design (iv) however is highly sensitive to residual stress in the flexures and large tensile or compressive stress will result in warping of the mirror even without an applied voltage [5].
When an electrostatic load is applied to a filter that incorporates the top mirror sharing a common surface with the flexures (i.e. deposition of the top mirror directly onto the flexures, or vice versa), the deformation induced in the flexures will inevitably be transferred to the top mirror. It is for this reason that the designs proposed by Huang et al. suffer in terms of parallelism (or flatness), especially at large mirror displacements. Figure 5.2 illustrates schematically cross sectional views of filters that are fabricated with the top mirror deposited onto a cantilever beam (Fig. 5.2(a)) and a fixed-fixed beam (Fig. 5.2(b)). When an electrostatic load is applied between the top and bottom electrodes, the flexures will deform, resulting in the top mirror moving towards the bottom mirror as
FIG. 5.2. Schematic illustration of mirror distortion of surface-micromachined FP filters when subjected to electrostatic loads for (a) mirrors supported by a cantilever beam and (b) mirrors supported by a fixed-fixed beam.
the applied voltage is increased. Ideally, the top mirror should stay parallel to the bottom mirror with increasing bias. Any curvature or tilt in the mirror will adversely affect the finesse of the cavity [8]. The electrostatic force however will induce curvature into the flexures, which will be transferred to the top mirrors.

Attempts have been made to minimise the induced deformation of the mirror by increasing the stiffness of the mirror, decreasing the stiffness of the flexures (i.e. changing their thickness), changing the shape or positioning of the electrodes, or increasing the length of the flexures [5, 9]. These methods however can affect the performance of the filter, both mechanically and optically. Design (iv) is unique in that the supporting four flexures behave similar to cantilevers that run parallel to the sides of the top mirror. This design leads to very little curvature induced into the top mirror when a voltage is applied.

5.2. New Fabry-Pérot Filter Design

It is obvious from the above discussion that depositing the top mirror directly onto the flexure limits the device in terms of tuning range, parallelism and fill factor. To overcome these limitations, a different filter structure is designed in this study, as shown schematically in Fig. 5.3. This design, termed the suspension filter design, varies from those shown in Fig. 5.1 inasmuch as the top mirror is freely suspended underneath the supporting flexures. When a bias is applied between the top and bottom electrodes, the flexures deform towards the substrate, thus changing the length of the resonant cavity between the mirrors. It is intended that nearly all deformation will occur in the flexures, without affecting the top mirror. This design allows for large mirror displacements whilst maintaining parallelism. Consequently, high cavity finesse is maintained over the entire tuning range.

A similar MEMS structure which supports a mirror at its centre is the Texas Instrument’s Digital Micromirror Device (DMD). This MEMS device (shown in Fig. 1.3) utilises a torsional flexure which is electrostatically actuated via a yoke. A mirror is supported from the underneath by a centre support attached to the yoke. This design allows for all deformation to occur in the flexure, whilst maintaining a perfectly flat mirror. Similar to an FP filter, it is essential that the mirrors of the DMD stay perfectly flat during actuation as it is a requirement for high resolution imaging.
FIG. 5.3. (a) Deformation behaviour of the proposed suspension filter when subjected to an electrostatic load, (b) isometric view of the filter, and (c) 4×4 array.
An isometric view of the proposed suspension filter is shown in Fig. 5.3(b). The flexures consist of two fixed-fixed beams joined at the centre, where they attach to the top mirror. The ends of the flexures are attached to the substrate via end supports. As the flexures of the filter shown in Fig. 5.3(b) are positioned above the mirrors, the optically active area is increased. Figure 5.3(c) shows a $4 \times 4$ array of filters. For this particular set of filters, the mirror sizes are $100 \times 100 \ \mu$m, the flexure widths are $6 \ \mu$m and the spacing between adjacent mirrors is $4 \ \mu$m. The resulting fill-factor is 79%. The fill-factor can be further increased by modifying the flexures such that they consist of only one fixed-fixed beam. It is expected however that this flexure design will be less reliable when actuated due to its susceptibility to torsion.

In comparison to the filters proposed by Huang et al., the suspension filter combines the advantages of the four designs:

- Similar to designs (i), (ii) and (iii) in Fig. 5.1, the suspension filter utilises fixed-fixed beams which are advantageous compared to cantilever beams due to larger achievable displacements, and better displacement control of the top mirror at higher actuation voltages.
- Similar to designs (i), (ii) and (iii) in Fig. 5.1, the mirror deformation of the suspension filter is, in principle, independent of $\sigma_0$ in the flexures provided the stress is tensile or less than the Euler buckling limit if the stress is compressive.
- Similar to design (iv), the suspension filter design results in less mirror deformation during actuation as the mirror is mechanically independent of the flexures.

The fill factors of the four designs proposed by Huang et al. however are inferior to the new suspension filter. Designs (i), (ii) and (iii) in Fig. 5.1 have fill factors of approximately 17%, whilst the fill factor of design (iv) is 25%.

As mentioned in Section 3.6, the displacement of a fixed-fixed beam when subjected to an electrostatic force is typically around 40 % of the cavity length. A further increase in the applied bias will result in pull-in of the beam onto the substrate. In the suspension filter design, positioning the flexures above the top mirror increases the electrostatic cavity size (i.e. the distance between the top and bottom electrodes). Consequently, designing the filter so that the flexures are positioned above the top mirror should, in principle, increase the displacement of the mirror before pull-in. This is a major
advantage compared to the four filter designs proposed by Huang et al. Larger mirror displacements result in a larger achievable FSR. If a high degree of parallelism is maintained across the suspension filter’s tuning range, smaller line widths of the resonant peaks can be expected, resulting in higher cavity finesse.

5.3. Fabrication of the Proposed Suspension Filter

Fabrication of the new FP filter design was initially carried out to determine its feasibility, as well as optimising the required process steps. Due to resolution limitations of emulsion masks, the dimensions of the FP filter illustrated in Fig. 5.3 were scaled up. The resolution required for the successful fabrication of the suspension filter however can be easily achieved using Cr masks. For initial device fabrication, the bottom mirror and the top and bottom electrodes were omitted from the fabrication steps, and a Si substrate was used.

Figure 5.4 shows the steps used to fabricate single mirror filters on Si substrates. The Si substrates were initially dipped in HF to remove the native oxide layer. Polyimide (PI)-2616 was then spun onto the substrates and baked at 130 °C for 20 minutes (Fig. 5.4(a)). More information on partial curing of polyimides can be found in Ref. [10]. The spin speed versus film thickness curve of partially cured (130 °C for 20 minutes) PI-2616 is shown in Fig. 5.5. For the first polyimide layer, a spin speed of 3000 rpm was used. The top mirrors were then deposited and defined using a dry etch process consisting of CF₄ plasma (Fig. 5.4(b)). The top mirrors were fabricated from e-beam evaporated Ge/SiO/Ge stacks of thickness 120/280/120 nm, as proposed by Dell et al. and Antoszewski et al. [1, 6]. The Ge and SiO layers were both deposited at 10 Å/s. The mirror deposition was carried out using a BOC Edwards Auto 500 Vacuum Coater. This system enables alternate layers of Ge and SiO thin films to be deposited for the one pump-down process, thus minimising contamination of the layers.

Photo-definable PI-2737 was then spun on at 7000 rpm and baked initially at 100 °C for 1 minute, followed by a second bake at 130 °C for 1 minute (Fig. 5.4(c)). After baking, the sample was flood exposed under a UV source for 40 s. This second PI layer defines the spacing between top mirror and the flexures and acts as the second sacrificial layer. PI-2616 cannot be used for this step as solvent transfer will occur between the two PI layers, resulting in delamination of the patterned top mirrors. The spin speed versus
FIG. 5.4. Process steps necessary for the fabrication of a single mirror FP filters on Si.
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The film thickness curve of PI-2737 using this process is also shown in Fig. 5.5. The polyimide layers were then patterned using a dry etch process consisting of 1:20 gas mixture ratio of CH₄:O₂ (Fig. 5.4(d)). This step was carried out using a two mask process – one mask to etch both polyimide layers down to the substrate (for the ends supports) and another mask to etch the second polyimide layer down to the top mirror (for the centre support). This method minimised etching times of the centre support etch and thus reduced etching in directions parallel to the substrate. To improve step coverage, a photoresist with an overcut profile was used (see Section 2.5).

The flexures were then deposited from $\beta$-series 500 nm thick PECVD SiN$_x$H$_y$ thin films deposited at 125 °C (Fig. 5.4(e)). In Chapter 4, it was shown that films deposited under these conditions contain only moderate residual tensile stresses, which are desirable for producing flat filters. Etching of the flexures was then carried out using CF$_4$ plasma. The filters were then released from the polyimide sacrificial layers using a wet process (Fig. 5.4(f)) consisting of a weak base (see Section 2.2). Samples were then washed in deionised water and transferred to methanol to be dried using a critical point dryer (see Section 2.5).

FIG. 5.5. Dependences of film thickness on spin speed for Polyimide (PI)-2616 and PI-2737.

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Figure 5.6 shows optical images of arrays of FP filters (bottom mirror and electrodes omitted) fabricated on Si wafers. Figure 5.6(a) and (b) shows unreleased filters with double and single support flexures, respectively. Figure 5.6(c) and (d) shows the filters after the sacrificial layers have been removed. The sizes of the top mirrors are 250×250 μm. It is obvious from Fig. 5.6(c) and (d) that the mirrors are not planar once the sacrificial layers are removed. Figure 5.7 plots the height variation across the centre of one of the released suspension filters. The height measurement was obtained using an optical profilometer. The centre support is not plotted in the figure as it was out of focus during the measurement. The plot show the mirror curves right down to the substrate. As the mirrors are symmetrical through their thickness, the curvature can only be due to stress gradients in the mirror materials (see Section 3.4).

It is well known that some physical vapour deposited (PVD) thin films exhibit thickness-dependent stress distributions (i.e. stress gradients) [11, 12]. During a PVD event, source material is boiled onto the substrate. Depending on the material, the

![Optical images of fabricated FP filters](image-url)  
(a)  
(b)  
(c)  
(d)

FIG. 5.6. Optical images of fabricated FP filters with double supports (a) and single supports (b) in the unreleased state, and released state (c and d). The sizes of the mirrors are approximately 250×250 μm.
evaporant can be many thousands of degrees in temperature, and consequently, the temperature of the substrate will inevitably increase during deposition. Changes in substrate temperature during deposition will lead to changes in grain size and growth through the thickness of the film. The stress in PVD films is also dependent on the rate of deposition. If the rate of deposition is too high, energy minimisation by atom diffusion across the sample’s surface is restricted, resulting in a build-up of stress in the film. Mayr et al. [11] showed as film thickness increases, the grain structure of e-beam evaporated Zr_{65}Al_{7.5}Cu_{27.5} film and the surface roughness of the film increase. Guisbiers et al. [12] showed the average tensile stress in e-beam deposited Ta, Al, Mo and Pd thin films decreased with increasing film thickness, suggesting the existence of stress gradients.

Winchester [8] showed using a ray approach that even a small deviation (a few nanometres) in parallelism between the top and the bottom mirrors will have detrimental effects on the finesse and transmission of the cavity. Depending on its transmission, the finesse of a cavity can be reduced by as much as an order of magnitude if deviations from parallel exceed 10 nm [8]. Figures 5.6 and 5.7 clearly show the mirror system used is unacceptable, and highlights the vulnerability of the suspension filter to stress gradients in

![Graph showing height variation across the top mirror of a released suspension filter measured using an optical profilometer. The variation was measured across the centre of a filter (see insert).](image)
the top mirror. The stress gradients in the Ge/SiO/Ge mirror stack have less of an influence towards cavity finesse for an X-style filter design [1, 6, 8, 13, 14]. In order to remove the curvature in the top mirror of the suspension filter, three methods are considered:

- Change the deposition conditions (deposition rate, substrate temperature etc.) to change grain growth in Ge and SiO films.
- Deposit an additional thin film layer in the mirror stack to compensate for the stress non-uniformity through the thickness of the mirrors.
- Use an alternative mirror stack system.

Deposition systems containing substrate temperature controllers can be utilised to change the grain growth of a PVD film [15, 16]. Due to processing temperature constraints, increasing the substrate temperature to improve the stress uniformity of the Ge and SiO films is not feasible. A thermal evaporation system fitted with a substrate temperature controller was utilised to evaporate Ge and SiO films whilst maintaining a constant substrate temperature of 25 °C. It was found however that this made little difference to the stress gradients. Changing the deposition rates of the Ge and SiO thin films is another alternative to improving stress gradients [17]. However, the optical performance of the mirror stack is optimised at deposition rates of 10 Å/s [6].

Depositing an additional thin film layer is feasible, provided it does not affect the optical properties of the filter. The material would need to be free of stress gradients itself, or at least result in curvature of the mirror in the opposite direction (i.e. curve mirror up instead of down). PECVD SiNₓHᵧ is an ideal choice as its stress profile can be easily controlled with changes in process conditions. The film (deposited at temperatures at, or below 125 °C) would be deposited before or after the deposition of the Ge/SiO/Ge stack. In order to flatten the mirror, a SiNₓHᵧ film deposited before the mirror deposition event would need to be compressive in nature, or tensile if deposited after the mirror.

Using alternative mirror materials in the stack is also a feasible option. Due to its excellent mechanical properties, low surface roughness [18] and high refractive index, PECVD a-Si:H is an ideal choice for use as the high refractive index material in the top mirror. Further, it has been shown that PECVD a-Si:H can be deposited at low temperatures with zero stress gradient [19], allowing for highly parallel mirrors to be
realised once the sacrificial layers are removed. By combining PECVD a-Si:H with low refractive index PECVD SiN$_x$H$_y$ thin films, a DBR mirror stack could be realised without removing the sample from the reactor chamber between depositions. Minimising the exposure of the inner surfaces of the stack to atmosphere between depositions will result in less contaminated mirrors.

5.4. Optical Modelling of the Proposed Filter

In order to conduct accurate spectral analysis and optical optimisation of the FP filter, the optical transfer matrix method was utilised [20]. This method takes into account the finite thickness of all films used in the DBR mirror stacks, as well as their wavelength dependence of their individual refractive indices and absorption coefficients. A summary of the method can be found in Appendix B of this thesis. Detailed development of this technique can be found in Refs. [20, 21].

The top and bottom DBR mirrors consist of alternating layers of high index material (Ge) and low index material (SiO) thin films of thicknesses ($t_i$) equal to:

$$ t_i = \left( \frac{2m + 1}{4n_i} \right) \lambda_0 $$

(5.1)

where $m = 0, 1, 2, \ldots, n_i$ is the refractive index of the $i$th layer and $\lambda_0$ is the mean wave number wavelength equal to:

$$ \lambda_0 = \frac{2}{\frac{1}{\lambda_{\text{min}}} + \frac{1}{\lambda_{\text{max}}}} $$

(5.2)

where $\lambda_{\text{max}}$ and $\lambda_{\text{min}}$ are the maximum and minimum wavelengths of interest, respectively.

In this case, $\lambda_{\text{max}} = 2.5$ μm, $\lambda_{\text{min}} = 1.5$ μm and thus $\lambda_0 = 1.875$ μm. From Ref. [22], $n(1.875$ μm) = 4.13 for Ge. For SiO, $n(1.875$ μm) = 1.84. Using Eqn. (5.1), the first order thicknesses (i.e. one-quarter wavelength optical thicknesses) were calculated to be $t_{\text{Ge}} = 113$ nm and $t_{\text{SiO}} = 255$ nm. The refractive index of the cavity was set to 1.0. Though the filters in this study were fabricated on Si substrates, they were designed with the notion of being monolithically integrated with Hg(1-x)Cd$_x$Te detector material. Consequently, the modelled substrate consists of 100 nm CdTe passivation layer on a short-wave IR Hg(1-x)Cd$_x$Te detector.
5.4.1. Transmission Spectrum of Stress-Free Mirrors

Figure 5.8 illustrates two different mirror stack designs and their corresponding transmission spectrums. In the design shown in (a), the top mirror is of a three-layered stack consisting of one-quarter wavelength thicknesses of Ge/SiO/Ge thin films. The bottom mirror has an identical stack structure. In the design shown in (b), the top mirror remains the same as in (a) but the bottom mirror contains an extra one-quarter wavelength thick SiO layer below. The respective transmission spectrums of the two stacks are shown in Fig. 5.8(c) and (d). To obtain these spectrums, the $n$ and extinction coefficient ($k$) values for each material were taken from Ref. [22] and the top mirrors are assumed perfectly flat and free of stress gradients. The cavity lengths corresponding to each peak are shown above the graphs. The FSR for each filter is similar (~1.5-2.5 $\mu$m), however the transmission intensity and finesse of the cavity for the three-layered bottom mirror stack are both lower compared to the four-layered bottom mirror stack. Table 5.1 lists the FWHM and associated cavity finesse for each transmission peak shown in Fig. 5.8.

For a cavity size of 1400 nm, the first and second order resonant peaks occur at wavelengths of 2502 nm and 1486 nm, respectively for the filter with the three-layered bottom mirror, and 2486 nm and 1491 nm, respectively for filter with the four-layered bottom mirror. This corresponds to FSR values for the three- and four-layered bottom mirror filters equal to 1016 nm and 995 nm, respectively. Increasing the number of layers of a mirror stack in a filter will increase the cavity finesse, whilst reducing its FSR [21]. The additional SiO layer in the bottom mirror increases the filter’s transmission due to better matching of the top and bottom mirror with the inclusion of the substrate.

5.4.2. Inclusion of a Stress Compensating Film

As mentioned in Section 5.3, the inclusion of a SiN$_x$H$_y$ layer in the top mirror stack should aid in improving parallelism. SiN$_x$H$_y$ thin films have a low refractive index ($n \approx 1.8$), and very low extinction coefficients [6]. Using $n_{\text{SiN}_x\text{H}_y} = 1.8$, from Eqn. (5.1) the one-quarter wavelength optical thickness of SiN$_x$H$_y$ is 260 nm. Figure 5.9 plots the theoretical transmission spectrums of filters on Hg$_{(1-x)}$Cd$_x$Te substrates with the inclusion of a SiN$_x$H$_y$ layer positioned at the bottom of the top mirror. The bottom mirror is identical to that shown in Fig. 5.8(b). Figure 5.9(b), (c) and (d) plot the transmission spectrums of filters with SiN$_x$H$_y$ optical thicknesses equal to one-quarter wavelength (260 nm), one-eighth wavelength (130 nm) and one-sixteenth wavelength (65 nm),
FIG. 5.8. Optical impedance matrix modelling of two different FP filters, (a) and (b) illustrate FP filters with a three and four-layered bottom mirrors, respectively, (c) and (d) show the transmission spectrums of the filters shown in (a) and (b), respectively.
respectively. The maximum cavity size for each filter was chosen to obtain a first-order transmission peak close to 2500 nm. As the SiN<sub>x</sub>H<sub>y</sub> thickness decreases, the FSR increases, as well as the filter’s transmission. The effect on the FWHM of the transmission peaks however is only minor.

Figure 5.10 plots the changes in FSR of the filter with changes in SiN<sub>x</sub>H<sub>y</sub> film thickness. Also plotted is the cavity length which was varied in order to obtain a first order transmission peak at a wavelength of 2500 nm. The change in cavity length varies substantially with changes in SiN<sub>x</sub>H<sub>y</sub> film thickness, and is also plotted in Fig. 5.10. The FSR decreases with increasing SiN<sub>x</sub>H<sub>y</sub> film thickness for thicknesses ranging from 0-300 nm. For SiN<sub>x</sub>H<sub>y</sub> film thicknesses greater than 300 nm, the FSR increases. At these higher thicknesses however, the cavity length becomes too small to be a feasible design. Hence in order to maximise the FSR, the SiN<sub>x</sub>H<sub>y</sub> film thickness should be made as small as possible.

Figure 5.11 plots the theoretical transmission spectrums of filters on Hg(1-x)Cd<sub>x</sub>Te substrates with the inclusion of a SiN<sub>x</sub>H<sub>y</sub> layer positioned on top of the top mirror. Figure 5.11(b), (c) and (d) plot the transmission spectrums of filters with SiN<sub>x</sub>H<sub>y</sub> optical
FIG. 5.9. Theoretical transmission spectrums of an FP filter with a Si$_n$H$_y$ layer included at the bottom of the top mirror, (a) illustrates the filter layers, and (b), (c) and (d) show the transmission spectrums of the filter for Si$_n$H$_y$ thicknesses of 260 nm, 130 nm and 65 nm, respectively.
thicknesses equal to one-quarter wavelength (260 nm), one-eighth wavelength (130 nm) and one-sixteenth wavelength (65 nm), respectively. The plots show that as the thickness of the SiN$_x$H$_y$ layer increases, the FSR increases slightly, as well as the FWHM. Similar to the filters shown in Fig. 5.9, the transmission of the filter in Fig. 5.11 improves with decreasing SiN$_x$H$_y$ thickness. The change in cavity length required to obtain a first order transmission peak at 2500 nm is almost independent of changes in SiN$_x$H$_y$ thickness. This result alone shows the performance of the filter is less influenced by changes in SiN$_x$H$_y$ thickness, which is a desired result when tuning SiN$_x$H$_y$ thicknesses to obtain flat mirrors.

5.4.3. Alternative Mirror Designs

Figure 5.12 plots the transmission spectrums of alternative mirror designs using $a$-Si:H as the high refractive index material. At 1875 nm, the refractive index of $a$-Si:H is 3.44 [22]. From Eqn. (5.1), the one-quarter wavelength optical thickness of the $a$-Si:H layer is 136 nm. Figure 5.12(a) and (b) shows a filter design and its respective transmission spectrum consisting of $a$-Si:H and SiN$_x$H$_y$ layers as the high and low refractive index layers in the top mirror, respectively. The high transmission, high FSR
FIG. 5.11. Theoretical transmission spectrums of an FP filter with a SiN$_x$H$_y$ layer included at the top of the top mirror, (a) illustrates the filter layers, and (b), (c) and (d) show the transmission spectrums of the filter for SiN$_x$H$_y$ thicknesses of 260 nm, 130 nm and 65 nm, respectively.
FIG. 5.12. (a), (c) and (e) illustrate alternative FP filter designs utilising $\alpha$-Si:H thin films, (b), (d) and (e) plot their respective theoretical transmission spectrums.
and low FWHM suggest that this filter design would serve as a good replacement for the previous filter design which incorporates Ge and SiO in the top mirror. Figure 5.12(c) shows another alternative filter design which incorporates an air cavity in the top mirror to replace the SiN$_x$H$_y$ layer. The transmission spectrum for this filter design is shown in Fig. 5.12(d). The air cavity in the top mirror is designed so that it has a fixed cavity length equal to one-quarter wavelength optical thickness (equal to 469 nm using Eqn. (5.1) assuming $n_{air} = 1.0$). This filter design has improved line widths, as well as a higher FSR, however the transmission is lower. Better matching of the top and bottom mirror is achieved using the bottom mirror design shown in Fig. 5.12(e), with its associated transmission spectrum shown in Fig. 5.12(f). In this case the bottom mirror consists of alternating layers of Ge and SiN$_x$H$_y$. Due to its lower absorption properties, the SiO layers have been replaced with SiN$_x$H$_y$ layers. The six-layered bottom mirror results in much smaller line widths (FWHM = 9 nm for transmission peak at 1961 nm), and consequently much higher cavity finesse ($F_c = 111$ for transmission peak at 1961 nm).

Adopting air as the low refractive index material in a DBR mirror system has previously been shown to improve a FP filter’s optical performance, in terms of reducing line-widths and improving cavity finesse [23]. In order to fabricate such a top mirror, the 469 nm separation between the two a-Si:H layers would need to be defined using a sacrificial layer. As shown in Fig. 5.13, incorporating this mirror design into the new

![FIG. 5.13. Top Mirror design incorporating a-Si:H layers and a one-quarter wavelength optical thick sacrificial layer (thicknesses have been increased by a factor 5 for viewing purposes only). The end supports are known shown.](image)
filter design would be achieved by connecting the two a-Si:H layers via an extra a-Si:H support layer, which would run underneath the supporting flexures. This would help to maintain parallelism between the two a-Si:H mirror layers whilst not affecting the fill-factor.

5.5. Mirror Curvature Optimisation

Due to time constraints, the filter designs incorporating a-Si:H thin films were not considered for the fabrication of the tunable shortwave IR filter. Though the cavity finesse of the filter design shown in Fig. 5.12(e) is desirable, fabrication of low-temperature a-Si:H thin films separated by a 469 nm sacrificial layer requires extensive research. In this work, the use of stress compensation layers incorporating SiN$_x$H$_y$ into the Ge/SiO/Ge top mirror stack is studied. As discussed in Section 5.4.2, it is desirable to control the stress in the top mirror by utilising a SiN$_x$H$_y$ layer, deposited as thin as possible, on top of the mirror stack (i.e. outside of the cavity). Surface micromachined microcantilevers consisting of e-beam evaporated one-quarter wavelength optical thickness Ge/SiO/Ge layers and PECVD SiN$_x$H$_y$ layers were fabricated to determine the optimum thickness of the SiN$_x$H$_y$ layer required to flatten the structure.

Figure 5.14 illustrates the process steps used to fabricate multilayer microcantilevers. Polyimide-2616 was initially spun onto Si substrates at 3000 rpm and baked at 130 °C for 20 minutes. Patterning of the polyimide was carried out using a dry etch process as explained previously. E-beam evaporated Ge and SiO thin films were then deposited, followed by the deposition of a SiN$_x$H$_y$ layer of known deposition time. The films were then patterned into microcantilevers of various dimensions using CF$_4$ plasma. The polyimide layer was then removed using a weak base, followed by rinsing. The samples were dried using a critical point dryer. The curvature of the beams was measured using an optical profilometer.

For Ge/SiO/Ge mirrors that curve downwards due to stress gradients, the top SiN$_x$H$_y$ layer needs to be tensile in nature in order to flatten the mirror. To cancel out the stress gradient in the mirror with the thinnest possible SiN$_x$H$_y$ film, it needs to be deposited with the highest possible tensile stress. As presented in Chapter 4 (Fig. 4.18), the highest tensile residual stressed SiN$_x$H$_y$ film that can be integrated with Hg$_{(1-x)}$Cd$_x$Te is a β-series film deposited at 125 °C. Figure 5.15 plots the inverse in radius of curvature (R)
FIG. 5.14. Process steps used to fabricate surface micromachined multilayer thin film microcantilevers.
against changes in SiN$_x$H$_y$ thickness for microcantilevers consisting of β-series films deposited at 125 °C on the topside of Ge/SiO/Ge quarter wavelength stack layers. As seen in the figure, the SiN$_x$H$_y$ film thickness required to flatten the cantilever ($1/R = 0$) is large (approaching 2 µm). From the optical analysis shown in Fig. 5.11, thick SiN$_x$H$_y$ films deposited on top of the top mirror have detrimental effects on the optical properties of the filter. Also, depositing such a thick film using PECVD is not practical. Consequently, in order to flatten the top mirror, a compressively stressed SiN$_x$H$_y$ film deposited on the underside of the Ge/SiO/Ge stack is required. As presented in Chapter 4 (Fig. 4.18), the highest compressive residual stressed SiN$_x$H$_y$ film that can be integrated with Hg$_{(1-x)}$Cd$_x$Te is an α-series film a deposited at 100 °C. For the purposes of this work however, an α-series film deposited at 125 °C was used to compensate for the stress in the top mirror. SiN$_x$H$_y$ films deposited under these conditions are more chemically inert compared to lower temperature deposited film when subjected to a base (see Fig. 4.2). This will occur during the PI release process. Though it is expected the amount of etching of a low-temperature deposited SiN$_x$H$_y$ film subjected to this process will be minimal, the sensitive nature of compensating for the top mirror curvature requires a high degree of SiN$_x$H$_y$ film
thickness uniformity of across the mirror’s surface. As the release process of the suspension filter involves etching of the PI underneath the top mirror, the etching front starts at the edges of the mirror and will finish underneath its centre. Any etching of the SiN,Hₓ film during this process will result in a non-uniform film thickness across the width of the top mirror.

Figure 5.16 plots the inverse in radius of curvature (R) against changes in SiN,Hₓ thickness for microcantilevers consisting of α-series films deposited at 125 °C on the topside of Ge/SiO/Ge quarter wavelength stack layers. The figure shows that the stress gradients in the mirror materials can be compensated for by depositing an α-series SiN,Hₓ thin film of thickness approximately 100-140 nm. From Fig. 5.9, a SiN,Hₓ film of this thickness will have negligible effects on the FSR and finesse of the cavity. Figure 5.17 illustrates an array of released single arm suspension filters fabricated on Si substrates with α-series SiN,Hₓ thin films of thickness 130 nm used as the stress compensating layers. From optical profilometry measurements, the mirrors are bending up ~20 nm along the diagonal from the centre supports, corresponding to a curvature of

![Graph](image)

**FIG. 5.16.** α-series SiN,Hₓ thin film compensating layer thickness and deposition time versus 1/R for microcantilevers consisting of SiN,Hₓ/Ge/SiO/Ge multilayer stacks (SiN,Hₓ on bottom).
approximately 0.001 mm\(^{-1}\). Though not completely optimised, Fig. 5.17 shows that adopting a compressively stressed film to the underside of an e-beam deposited Ge/SiO/Ge mirror can greatly improve its flatness.

### 5.6. Finite Element Modelling

Finite element modelling (FEM) was carried out using the software package CoventorWare version 2006.1 to determine the mechanical performance and resonance behaviour of the suspension filter that utilises a 130 nm thick Si\(\text{N}_x\text{H}_y\) thin film compensation layer deposited underneath the top mirror. The modelled filter is shown in Fig 5.18(a). It consists of 100×100 µm mirrors and 6 µm wide flexures. Since the bottom mirror is a stationary object, it is not included in the model. The top Ge layer of the bottom mirror serves as the bottom electrode. Metal-induced crystallisation of Ge results in an increase in its conductivity, which is desirable for its function as the electrode. The crystallisation can be achieved by incorporating a monolayer of In [24]. The supporting
flexures consist of a 300 nm thick SiN$_x$H$_y$ layer and a 30 nm Au layer deposited on top as
the top electrode. The Au layer covers the SiN$_x$H$_y$ layer in a conformal manner and hence
has continuity into the centre support. Due to its tensile nature, the SiN$_x$H$_y$ flexure layer
will consist of a $\beta$-series film deposited at 125 $^\circ$C. From the measurements presented in
Chapter 4, the average residual stress, density and Young’s modulus of this film are
10 MPa, 1.68 g/cm$^3$ and 78 GPa, respectively. From Ref [25], $E$(Au) = 66 GPa and
$\rho$(Au) = 19.3 g/cm$^3$. Work carried out by Gall et al. [26] has shown that the average
residual stress in evaporated Au thin films can be easily tuned to a zero via low-
temperature annealing. For the purpose of this work, the residual stress in the Au layer
was set to 0 MPa.

The top mirror was modelled as a single material so that the Young’s modulus,
density and Poisson’s ratio were equivalent to the four-layer SiN$_x$H$_y$/Ge/SiO/Ge stack.

FIG. 5.18. Finite element model of proposed filter showing (a) total modelled filter, and (b)
meshed quarter model. In both cases, the dimensions in the z-direction have been increased by
a factor of 5 for viewing purposes only.
From Chapter 4, $E$ and $\rho$ of an $\alpha$-series SiN$_x$H$_y$ thin film deposited at 125 ºC are equal to 104 GPa and 2.18 g/cm$^3$, respectively. The thickness of the $\alpha$-series SiN$_x$H$_y$ stress compensating layer was set to 130 nm. From Ref [10], $E$(Ge) = 102 GPa, $E$(SiO) = 90 GPa, $\rho$(Ge) = 5.35 g/cm$^3$, $\rho$(SiO) = 2.13 g/cm$^3$. The dielectric constant ($\varepsilon_0$) of the mirror was modelled such that its total capacitance was the equivalent to the actual mirror stack for the same thickness. The dielectric constant of the SiN$_x$H$_y$ layers was approximated to be equivalent to the square of its refractive index. From Fig. 5.9, a 130 nm thick SiN$_x$H$_y$ stress compensation layer is equivalent to an optical thickness of $(1/8)\lambda$ and thus a cavity length of $\sim$1.3 µm is required to obtain a first order resonant peak at 2.5 µm. The distance between the flexures and the mirror was set to 800 nm. From Fig. 5.5, a PI-2737 layer spun on at 9000 rpm will achieve this thickness.

To minimise the effects of non-symmetrical meshing, only a quarter of the filter was meshed and symmetry boundary conditions utilised. Figure 5.18(b) shows the meshed quarter filter model. As the centre support is circular in shape, the mirror and flexures were meshed with a 5 µm tetrahedral mesh. The bottom electrode was meshed with a 5 µm, hexahedral mesh. The boundary conditions of the nodes at the end of the flexure were set so that all displacements and rotations were set to zero. Figure 5.19 plots the centre mirror displacement versus applied bias to the top electrode for the meshed filter shown in Fig. 5.18, with and without the inclusion of residual stress in the SiN$_x$H$_y$ flexure layer. From the optical modelling shown in Fig. 5.9, the cavity length required to obtain a first order resonant peak at 1502 nm is approximately 650 nm. To achieve this cavity length, the top mirror needs to be displaced by $\sim$750 nm, which is $\sim$57% of the initial cavity length. Figure 5.18 suggests that this displacement is easily achievable if the residual stress in the flexures is zero, however it cannot be achieved without pull-in occurring if the residual stress in the flexures is included. The tuning range of the mirror can be increased by increasing the distance between the mirror and the flexures. This will however increase the required actuation voltage, which is not feasible as the maximum required voltage shown in Fig. 5.19 is already close to 30 V – the maximum allowable voltage for the integration of this technology with application-specific integrated circuits (ASICs) required for filter control and detector signal acquisition.

Milne et al. [27] has shown by varying the dimensions of a fixed-fixed beam, the displacement of the beam can be increased before pull-in occurs. Figure 5.20 shows the centre mirror displacement versus applied bias for the meshed filter shown in Fig. 5.18,
FIG. 5.19. Results to FEM simulations showing centre mirror displacement versus applied bias to the top electrode with and without the inclusion of residual stress in the SiN,H$_y$ flexures.

FIG. 5.20. Results to FEM simulations for a filter with SiN,H$_y$ flexure thickness equal to 150 nm. The residual stress in the SiN,H$_y$ layer was included in the model.
except with the thickness of the SiN$_x$H$_y$ flexure layer halved to 150 nm. A residual stress of 10 MPa was included in the model for this layer. As seen in the figure, the required displacement of 750 nm is easily achieved using this design, at an actuation bias voltage just over 20 V. This is a desirable result.

An interesting feature of this filter design is the curling of the mirror towards the flexures (top electrode) when a bias voltage is applied. As discussed in section 5.2, the suspension filter is free of the problem of large mirror distortion during mirror displacement driven by electrostatic force on the top electrode because of the separation of the electrode and the mirror. However, a much minor attraction force may occur across the gap between the mirror and the top electrodes (the flexures) when a voltage is applied. Under the influence of this force, the corners of the mirror situating just under the flexures will curve up, causing mirror distortion. Modelling results suggest that the mirror distortion is negligible. For example, under the influence of a 20 V bias voltage, the corners of the mirror deflect only 5.1 nm towards the top electrode. This level of mirror distortion is of the same order of magnitude as the surface roughness of evaporated SiO and Ge thin films [10] and will have negligible effect on the finesse of the cavity.

A modal analysis was conducted on the filter designed with 150 nm thick SiN$_x$H$_y$ flexure supports to determine the frequencies the filter will resonate. Figure 5.21 illustrates the first three resonant modes and their associated frequencies. The first mode occurs at 26.1 kHz and involves bending of the flexures. The second and third resonant modes occur at much higher frequencies (199 kHz and 281 kHz) and involve bending in the top mirror. Ideally, the filter needs to operate at a high frequency (~10 kHz). As the vibration modes of the filter are much higher than 10 kHz, there will not be any detrimental effects regarding filter resonance on the performance of the device.

5.7. Device Fabrication

Suspension filters were fabricated specifically for optical and electro-mechanical testing. Figure 5.22 illustrates schematically the cross-sectional profile of a filter fabricated on double-sided polished Si substrates. Emulsion masks were utilised, so in order to obtain the desired resolution, large devices were fabricated. The devices were fabricated with 200×200 μm mirrors as well as longer flexures. Thicker SiN$_x$H$_y$ layers
FIG. 5.21. The first three resonant modes of the proposed filter with 150 nm thick SiN$_x$H$_y$ flexures. The residual stress in the SiN$_x$H$_y$ flexure layer was included in the model.
(410 nm) were also employed for the flexures. An Au layer of 100 nm thickness was used to act as a shield for the optical measurements (see section 5.8). Thick Au layers (300 nm) were deposited on areas of the support arms where changes in step height occurred and in areas where bonding pads were positioned. All Au depositions employed a thin layer (5 nm) of Cr to improve Au adhesion. For these devices, the Si substrate served as the bottom conductor. A detailed summary on the processing steps used to fabricate such a device is presented in Appendix C of this thesis.

Figure 5.23 shows scanning electron microscopic (SEM) images of two filters, which are poorly fabricated. Fig. 5.23(a) shows a filter with two fixed-fixed beam supports for the top mirror. The top mirror has detached from its flexures. This is due to poor adhesion between SiN$_x$H$_y$ and Ge. This problem was eradicated by performing an additional CF$_4$ etch to remove the Ge layer in the area where the flexures attach to the top mirror, allowing the SiN$_x$H$_y$ layer in the flexures to be deposited directly onto the SiO layer in the top mirror at the centre support. The adhesion of SiN$_x$H$_y$ on SiO is far better than that of SiN$_x$H$_y$ on Ge. Figure 5.23(b) illustrates a single fixed-fixed beam filter. This filter features residual PI-2737 left on the filter. This is due to the polyimide layer not being properly etched during the release process. Though PI-2616 can be easily etched at
FIG. 5.23. Scanning electron micrographs of poorly fabricated filters showing (a) detached mirror of a filter consisting of double arm fixed-fixed beam flexures, and (b) residual PI-2737 hanging from a filter consisting of a single arm fixed-fixed beam flexure. The size of the mirrors are 200×200 µm.
room temperatures, PI-2737 needs to be etched at elevated temperatures (~50-60ºC) in order for it to be fully removed from the filter. Figure 5.24 illustrates a correctly fabricated device consisting of two fixed-fixed beams. In this particular device, the mirror is properly etched so that its SiO layer is in contact with the SiN$_x$H$_y$ flexure layer. Correct removal of the PI-2737 layer was also employed to release this structure. Hence there is no evidence of residual polyimide, which would otherwise have detrimental effects on the optical and electro-mechanical performance of the filter.

5.8. Device Characterisation

5.8.1. Optical Measurements

Transmission spectrum measurements were carried out on fixed cavity filters (i.e. unreleased filters) using the optical arrangement shown in Fig. 5.25. A tungsten filament blackbody source was used for illumination, while a filter and collimator were employed to remove the filament image and to collimate the light beam, respectively. The sample was mounted on a positioning stage to align the filters normal to the incident beam. A
20X objective lens was used to produce a recognisable image of the aligned suspension filters, which was observed using a beam splitter and a camera. A pin hole was employed to reduce the size of the beam such that only incident light passing through one suspension filter could be transmitted to the InGaAs detector. A monochromator was employed to filter particular wavelengths of the incident light through to the detector. The estimated incident beam diameter was 150 $\mu$m, which was smaller than the size of the mirrors of the suspension filter. Employing an incident beam of this diameter, as well as including an Au shield underneath the bottom mirror (see Fig. 5.22) ensured no leakage of incident light occurred around the patterned mirrors that would otherwise reach the InGaAs detector. The optical transmission spectrums were obtained by normalising the filter output relative to a reference. The reference consisted of an identical Si substrate that the filters were fabricated on. The reference substrate was patterned with the Au shield and supporting flexures of identical thicknesses as the measured filters.

Fig. 5.26 plots the optical transmission spectra of unreleased FP filters as well as the modelled spectra with polyimide included in the model. The cavity length of these filters was 1.8 $\mu$m – higher than the target thickness of 1.3 $\mu$m. Figure 5.26(a) plots the spectra of three filters consisting of single arm fixed-fixed beam flexures whilst

![FIG. 5.25. Schematic illustration showing the optical arrangement used to measure the transmission spectrum of Fabry-Pérot filters.](image-url)
FIG. 5.26. (a) Modelled and measured optical transmission spectra of three FP filters consisting of single arm fixed-fixed beam flexures, and (b) modelled and measured transmission spectrum of three FP filters consisting of double arm fixed-fixed beam flexures.
Fig. 5.26(b) plots the spectra of three filters consisting of double arm fixed-fixed beam flexures. The six filters were fabricated on the same Si wafer and spread across an area of 10×20 mm. The measured spectra of the filters show excellent agreement with the model in terms of peak position. The measured transmissions of the peaks however are approximately 20% lower compared to the modelled transmissions. This is most likely due to the non-perfect nature of the mirrors in terms of surface roughness and parallelism which the model does not take into account. The average line width of the peaks which occur at ~2130 nm of the three measured filters plotted in Fig. 5.26(a) is 35 nm. A similar average line width value of 36 nm is obtained for the filters plotted in Fig. 5.26(b). The wavelength spread of the peaks which occur at ~2130 nm of the three measured filters plotted in Fig. 5.26(a) is 10 nm, which is lower than the value of 25 nm obtained for the filters in Fig. 5.26(b). The heights of the peaks in Fig. 5.26(a) are also slightly higher on average, compared to the peaks in Fig. 5.26(b).

5.8.2. Electro-Mechanical Measurements

Figure 5.27 illustrates the deformation behaviour of the released FP filter shown in Fig. 5.24 when subjected to an applied bias. The cavity length of this particular filter was 1.31 µm. Figure 5.27(a) plots height measurements of the filter along one of its support arms when zero volts were applied to the top electrode. Also plotted are height measurements of the filter once it had collapsed down to the substrate due to pull-in. Measurements were obtained using an optical profilometer. These results show that under zero applied bias, the flexure profile is flat across its length, except for its ends. The unusual step at the beam’s right support is due to an optical profilometer error, which can occur if there is a discontinuity in the data. Figure 5.27(b) plots the central mirror displacement versus applied bias of the filter. Also plotted are simulated data obtained from finite element modelling using the dimensions of the filter. The measured results show excellent agreement with the model assuming the residual stress in the SiN_xH_y flexure layer is 10 MPa, as obtained from stress measurement data in Chapter 4. The mirror was displaced by applying a voltage across the top and bottom electrodes in steps of 2.5 V. When 17.5 V was applied to the top electrode, the mirror displaced 620 nm. A further increase in bias by 2.5 V resulted in pull-in, which lead to the top mirror being permanently adhered to the bottom mirror and consequent failure of the device. The simulation data however does suggest that if smaller voltage steps were used, stable mirror displacements greater than 750 nm could be achieved.
FIG. 5.27. Results to the electro-mechanical measurements of the FP filter shown in Fig. 5.24. (a) Height measurements of the filter along one of its support arms (see insert) when zero volts was applied to the top electrode, as well as once the filter had collapsed due to pull-in. Measurements were obtained using an optical profilometer. (b) Centre mirror displacement versus applied bias of the filter shown in Fig 5.25. Also plotted are modelled results of the filter for various residual stress values in the SiN$_x$H$_y$ flexure layer.
5.9. Discussion

The optical measurements discussed in Section 5.8 suggest that FP filters consisting of single arm fixed-fixed beam flexures perform slightly better optically compared to FP filters consisting of double arm fixed-fixed beam flexures. This could be due to dispersion which occurs due to the flexures being positioned directly above the mirrors. As a double arm fixed-fixed beam flexure has a greater surface area compared to a single arm fixed-fixed beam flexure, more dispersion will occur, thus having a detrimental effect on the optical transmission spectra of the filter. Unfortunately no spectra were measured on a released filter. The results presented above however do suggest that the device design is feasible both optically and mechanically.

References


Chapter 6

Conclusions and Future Work

6.1. Conclusions

This study has investigated the mechanical and physical properties of low-temperature plasma enhanced chemical vapour deposited (PECVD) silicon nitride (SiN$_x$H$_y$) thin films for the fabrication of short-wave infrared tunable Fabry-Pérot (FP) filters with high fill factor, high cavity finesse and low actuation voltages. It has been the intention of this work to fabricate a tunable filter that can be monolithically integrated with Hg$_{1-x}$Cd$_x$Te photoconductors and photodiodes. The main outcomes of this work are in three groups: (1) development and verification of micromechanical testing techniques of thin film materials, (2) determination and characterisation of mechanical and physical properties of PECVD SiN$_x$H$_y$ thin films, and (3) design and fabrication of the suspension FP filter.

6.1.1. Development and Verification of Micromechanical Testing Techniques of Thin Film Materials

A range of methods have been used to determine the Young’s modulus ($E$), Poisson’s ratio ($v$) residual stress ($\sigma_0$) and density ($\rho$) of thin film materials. These techniques include nanoindentation, microcantilever beam resonance method, the microcantilever beam bending method, the micromembrane bulge test, and the quartz crystal microbalance method.

(i) Nanoindentation, the resonance method, the microcantilever beam bending method and the bulge test were used to determine the $E$ of PECVD SiN$_x$H$_y$ thin films. All
methods have shown very good agreement to within 10% of the nanoindentation method.

(ii) This study has shown the dependence of the resonance method on the dimensions of resonating microcantilever beams and quantified the accuracy of standard bending and torsional resonance equations using finite element modelling (FEM). For microcantilever beams with length-to-width ratios greater than 8, the resonance method used in this study to determine \( \nu \) is accurate to within 7%.

(iii) The resonance method has also been used to determine \( \rho \) of thin film materials. This method has shown very good agreement (to within 10%) of the quartz crystal microbalance method.

(iv) This study has also derived an equation (Eqn. 4.12) which can be used to determine \( E \) of microcantilever beams of variable thickness along their lengths using the microcantilever beam bending method. This equation is necessary for determining \( E \) from tapered beams caused by chemical microfabrication processes.

(v) The standard bulge test has been modified in this study to determine \( \nu \) of thin films. The modified bulge test varies from the standard bulge test in that both square and rectangular-shaped membranes are fabricated alongside, and their relative deflections are measured when subjected to an applied pressure. This new invention allows for a more accurate determination of \( \nu \) as the measurement is less affected by variations in membrane dimensions. It has been shown that this method closely agrees (to within 5%) with the resonance method used to determine \( \nu \).

(vi) The bulge test has also been used to determine \( \sigma_0 \) of thin films. It has been shown in this study that this method provided similar residual stress values compared to the substrate curvature method, which was also utilised in this study.

6.1.2. Mechanical and Physical Properties of PECVD SiN\(_x\)H\(_y\) Thin Films

The Young’s modulus, Poisson’s ratio, residual stress and density of PECVD SiN\(_x\)H\(_y\) thin films have been determined using the methods outlined above. In this study, all PECVD SiN\(_x\)H\(_y\) thin films were deposited using an Oxford Instruments PECVD 80 system at an RF frequency of 13.6 MHz with SiH\(_4\):NH\(_3\):N\(_2\) flow rates of 5:50:100 SCCM.
The deposition temperature varied between 100-300 °C. The effects of changes of other processing parameters such as chamber pressure and RF power on $E$, $\nu$, $\sigma_0$ and $\rho$ have also been evaluated.

(i) The highest $E$ measured was 195 GPa for a film deposited at 300 °C, 600 mTorr and 100 W. The lowest $E$ measured was 54 GPa for a film deposited at 100 °C, 875 mTorr and 75 W. An increase in chuck temperature, increase in RF power and decrease in chamber pressure all increase $E$.

(ii) The highest $\rho$ measured was 2.90 g/cm$^3$ for SiN$_x$H$_y$ films deposited at 300 °C, 600 mTorr and 100 W. The lowest $\rho$ measured was 1.55 g/cm$^3$ for a SiN$_x$H$_y$ film deposited at 100 °C, 875 mTorr and 75 W. Increasing chuck temperature, increasing RF power and decreasing chamber pressure all increase $\rho$, similar to the case of $E$.

(iii) The highest $\nu$ value measured was 0.26 ± 0.01 for SiN$_x$H$_y$ films deposited at 250 °C, 600 mTorr and 100 W. The lowest $\nu$ value achieved was 0.20 ± 0.01 for SiN$_x$H$_y$ films deposited at 100 °C, 875 mTorr and 75 W. An increase in chuck temperature increases $\nu$. Changes in RF power and chamber pressure have negligible effect on $\nu$.

(iv) The highest compressive residual stress measured was 133 MPa for SiN$_x$H$_y$ films deposited at 100 °C, 600 mTorr and 100 W. The highest tensile residual stress achieved was 59 MPa for SiN$_x$H$_y$ films deposited at 150 °C, 875 mTorr and 75 W. Increasing the chuck temperature, decreasing the RF power and increasing the chamber pressure all increase the tensile nature of the SiN$_x$H$_y$ thin films.

6.1.3. Design and Fabrication of the Suspension FP Filter

This study has conducted the theoretical design, analysis and fabrication of a new surface micro-machined short-wave infrared tunable Fabry-Pérot filter for adaptive infrared photon detectors.

(i) A new concept support-membrane configuration for Fabry-Pérot filter design is proposed. This design, termed the suspension filter, separates the top mirror from the supporting flexures which allows for improved fill-factors (up to 79%), as well
as increased tuning range. In order to improve parallelism of the filter, a PECVD SiN$_x$H$_y$ stress compensation layer was employed in the top mirror stack.

(ii) The suspension filter uses broad spectral range, high reflectivity distributed Bragg reflector (DBR) mirrors, which results in very high finesse filters. It has been shown a theoretical free spectral range (FRS) of approximately 1 µm (tuning across wavelengths from 1.5 µm to 2.5 µm) is achievable using filters which incorporate thermally evaporated Ge and SiO as the high and low-refractive index materials in the DBR mirrors, respectively. A cavity finesse of more than 40 can be achieved using this mirror system. An alternative mirror stack is also presented, which incorporates a-Si:H and air as the high and low-refractive index materials in the top mirror, respectively. This mirror system can theoretically achieve a similar FSR as the Ge and SiO mirrors, however with an achievable cavity finesse of more than 110.

(iii) Finite element modelling of the suspension filter has demonstrated top mirror displacements of more than 900 nm (~64% of the cavity length) are achievable with actuation voltages less than 30 V. As the required displacement for the filter to tune across wavelengths from 1.5-2.5 µm is only 750 nm, it has been shown the suspension filter can easily tune across this wavelength range. It has also been shown the filter is able to maintain high cavity finesse during tuning, showing negligible top mirror deformation across the desired wavelength range.

(iv) The fabrication process of this new design has also been established and arrays of suspension filters, as well as a filter that could be actuated were successfully fabricated. The filters were fabricated using a multiple sacrificial layer process to define the resonant cavity spacer and the separation of the top mirror from the supporting flexures. The flexures were fabricated from low-temperature (PECVD) SiN$_x$H$_y$ thin films.

(v) Optical characterisation of fabricated fixed cavity suspension filters on Si substrates has shown transmissions of ~60% with small line widths (35 nm) are achievable using the proposed mirror system incorporating Ge and SiO as the high and low refractive index materials, respectively.
(vi) A mirror displacement versus an applied bias voltage curve was obtained from a released suspension filter fabricated on Si. A mirror displacement of 620 nm was achieved whilst theoretical results suggest the required 750 nm mirror displacement was possible if smaller voltage increments were used.

6.2. Future Work

It was the intention of this work to obtain transmission spectra of a released Fabry-Pérot filter that could be tuned to a desired wavelength in the range 1.5-2.5 µm. This work has shown that such a target is feasible and that more work in this area is worth considering. The main challenges in realising this concept to a final engineering product are:

- Improve device yield through better fabrication processes;
- Improving the optical performance of the filter;
- Gaining a better understanding of the dynamics of the suspension filter.

Recommended future work for research includes:

(i) More robust designs should be made possible using smaller mirror sizes, which would require the use of Cr masks. Much higher resolutions are achievable using Cr masks compared to emulsions masks, which were used in this study. Increasing the resolution of the masks will enable smaller features to be fabricated, including the size of the mirrors. Smaller mirror sizes will decrease their sensitivity to stress gradients that exist in the mirror materials. High resolution masks will also help to decrease errors associated with mask alignment and consequently result in better performing devices both optically and mechanically, as well as increasing device yield.

(ii) This work has also presented new mirror structure designs that utilise a-Si:H thin films. Future work that focuses on optimising a-Si:H thin film processes could prove useful in terms of improving cavity finesse and parallelism of the filter.

(iii) The dynamics of the suspension filter when actuated at high frequencies has not been considered in this work. For hyperspectral imaging capabilities to be realised using this technology, the filter would be required to move across the spectral
tuning range in increments. At each increment, the filter would have a fixed voltage applied between its electrodes for a finite period of time, whereby a measurement would be taken. The time required for a measurement at each increment is dependent on computation time as well as device settling time. It is expected that the settling time of the proposed filter will be higher compared to designs previously fabricated by researchers. It is the opinion of the writer that a highly damped environment would be required to minimise the settling time however, issues regarding squeeze-film damping need to be considered.
# Appendix A

## List of Symbols

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<td>Length</td>
<td>$\lambda_m$</td>
<td>Mean free path</td>
</tr>
<tr>
<td>$E$</td>
<td>Young’s modulus</td>
<td>$M_f$</td>
<td>Mass of deposited film</td>
</tr>
<tr>
<td>$\nu$</td>
<td>Poisson’s ratio</td>
<td>$M_q$</td>
<td>Mass of quartz crystal</td>
</tr>
<tr>
<td>$\tau$</td>
<td>Shear stress</td>
<td>$F_q$</td>
<td>Quartz crystal resonance</td>
</tr>
<tr>
<td>$G$</td>
<td>Shear modulus</td>
<td>$S$</td>
<td>Contact stiffness</td>
</tr>
<tr>
<td>$E'$</td>
<td>Biaxial modulus</td>
<td>$E_r$</td>
<td>Reduced modulus</td>
</tr>
<tr>
<td>$E^*$</td>
<td>Plate modulus</td>
<td>$F_c$</td>
<td>Finesse of cavity</td>
</tr>
<tr>
<td>$\gamma$</td>
<td>Shear strain</td>
<td>$n$</td>
<td>Refractive index</td>
</tr>
<tr>
<td>$\kappa$</td>
<td>Curvature</td>
<td>$k$</td>
<td>Extinction coefficient</td>
</tr>
<tr>
<td>$w$</td>
<td>Deflection</td>
<td>$\lambda_0$</td>
<td>Mean wave number</td>
</tr>
<tr>
<td>$b$</td>
<td>Beam width</td>
<td>$N$</td>
<td>Complex refractive index</td>
</tr>
<tr>
<td>$t$</td>
<td>Beam thickness</td>
<td>$\delta$</td>
<td>Phase shift of a wave</td>
</tr>
<tr>
<td>$z$</td>
<td>Distance</td>
<td>$\eta$</td>
<td>Admittance of a film</td>
</tr>
<tr>
<td>$M$</td>
<td>Bending moment</td>
<td>$Y$</td>
<td>Optical admittance</td>
</tr>
<tr>
<td>$I$</td>
<td>Moment of inertia</td>
<td>$E$</td>
<td>Electric field</td>
</tr>
<tr>
<td>$m$</td>
<td>Gradient of beam thickness</td>
<td>$H$</td>
<td>Magnetic field</td>
</tr>
<tr>
<td>$P$</td>
<td>Pressure</td>
<td>$R$</td>
<td>Reflectance</td>
</tr>
<tr>
<td>$a_{w,r}$</td>
<td>Half width of a square/rectangle</td>
<td>$T$</td>
<td>Transmittance</td>
</tr>
<tr>
<td>$b_r$</td>
<td>Half length of rectangular</td>
<td>$A$</td>
<td>Absorbance</td>
</tr>
</tbody>
</table>
Appendix B

Deflection of a Tapered Beam due to a
Point Load

This section derives the equation found in Chapter 3 on page 64 which describes the deflection of a tapered beam due to a point load. Referring to Fig. B1, let \( t \) be the thickness of the beam at the point where the load \( F \) is applied and \( m \) be the gradient of the beam thickness. Also, let \( L \) and \( b \) be the length and width of the beam, respectively. The moment of inertia of the beam can thus be described using:

\[
I = \frac{b}{12}(t + mx)^3
\]  
(B.1)

From Eqn. (3.24):

\[
E^*I \frac{d^2w}{dx^2} = M = Fx \Rightarrow \frac{E^*b}{12F} \frac{d^2w}{dx^2} = \frac{x}{(t + mx)^3}
\]  
(B.2)

Integrating with respect to \( x \) yields:

\[
\frac{E^*b}{12F} \frac{dw}{dx} = -\frac{t + 2mx}{2m^2(t + mx)^2} + C_1
\]  
(B.3)

![FIG. B1. Tapered microcantilever beam due to bulk micromachining processes.](image-url)
Appendix B – Deflection of a Tapered Beam due to a Point Load

\[ \frac{dw}{dx}(L) = 0 \Rightarrow C_1 = \frac{t_i + 2mL}{2m^2(t_i + mL)^2} \]  \hspace{1cm} (B.4)

Thus:

\[ \frac{E'bw}{12F} \frac{dw}{dx} = -\frac{t_i + 2mx}{2m^2(t_i + mx)^2} + \frac{t_i + 2mL}{2m^2(t_i + mL)^2} \]  \hspace{1cm} (B.5)

Integrating with respect to \( x \) yields:

\[ \frac{E'bw}{12F} = \frac{t_i}{2m^2(t_i + mx)} - \frac{1}{m^2} \left[ \frac{t_i}{(t_i + mx)} + \ln(t_i + mx) \right] + \frac{(t_i + 2mL)x}{2m^2(t_i + mL)^2} + C_2 \]  \hspace{1cm} (B.6)

\[ w(L) = 0 \Rightarrow C_2 = -\frac{t_i}{2m^2(t_i + mL)} + \frac{1}{m^2} \left[ \frac{t_i}{(t_i + mL)} + \ln(t_i + mL) \right] - \frac{(t_i + 2mL)L}{2m^2(t_i + mL)^2} \]  \hspace{1cm} (B.7)

Maximum deflection occurs at \( F \), i.e. when \( x = 0 \). Hence:

\[ \frac{E'bw}{12F} = \frac{1}{2m^4} - \frac{1}{m^2} \left[ 1 + \ln(t_i) \right] - \frac{t_i}{2m^2(t_i + mL)} + \frac{1}{m^2} \left[ \frac{t_i}{(t_i + mL)} + \ln(t_i + mL) \right] - \frac{(t_i + 2mL)L}{2m^2(t_i + mL)^2} \]  \hspace{1cm} (B.8)

Rearranging yields:

\[ w = \left( \frac{F}{E'} \right) \left( \frac{6}{bm^3} \right) \left( 2 \ln(1 + \frac{mL}{t_i}) - \frac{mL(3mL + 2t_i)}{(t_i + mL)^2} \right) \]  \hspace{1cm} (B.9)
Appendix C

Summary of the Optical Transfer Matrix Method

This section provides a brief summary on the basic theory required to calculate the properties of multilayer thin film coatings. An introduction into the methods behind calculating the reflectance of a single layer thin film on a substrate, as well as a multilayer thin film assembly, is presented. More detail on these topics can be found in references [1-4] found at the end of this Appendix.

C.1. The Reflectance of a Thin Film Covering a Substrate

This section outlines the equations used to determine the reflectance of an incident plane wave normal to a single thin film covering a substrate. The film is said to be ‘thin’
when its thickness is less than the coherence length of the incident light. Figure C1 illustrates the arrangement of a plane wave, incident on a thin film. The film’s complex refractive index \((N_1)\), can be described using the following:

\[
N_1 = n_1 - ik_1 \quad \text{(C.1)}
\]

where \(n_1\) is the real part of the refractive index of the film, and \(k_1\) is known as the extinction coefficient.

For a plane wave of normal incidence on a thin film covering a substrate, the sum of the incident electric field and the reflected electric field (denoted \(E_a\)), and the sum of the incident magnetic field and the reflected magnetic field (denoted \(H_a\)) are related to the transmitted electric and magnetic fields (denoted \(E_b\) and \(H_b\), respectively) via the optical transfer matrix:

\[
\begin{bmatrix}
E_a \\
H_a
\end{bmatrix} = 
\begin{bmatrix}
\cos \delta & (i \sin \delta) / \eta_1 \\
(i \eta_1 \sin \delta & \cos \delta
\end{bmatrix}
\begin{bmatrix}
E_b \\
H_b
\end{bmatrix} \quad \text{(C.2)}
\]

where \(\delta\) is equal to the phase shift of the wave and \(\eta_1\) is the admittance of the film. For a plane wave of normal incidence:

\[
\eta_1 = N_1 y \quad \text{(C.3)}
\]

where \(y\) is usually complex. In free space however, \(y\) is equal to \(2.6544 \times 10^{-3}\) S. The 2×2 matrix on the right side of Eqn. (C.2) is known as the characteristic matrix of the film. The phase shift of a normal incidence plane wave travelling through the film’s thickness, \(d\) is given by:

\[
\delta = \frac{2\pi N_1 d}{\lambda} \quad \text{(C.4)}
\]

where \(N_1\) is the complex refractive index of the film, and \(\lambda\) is the wavelength. Define the input optical admittance, \(Y\) as:

\[
Y = \frac{H_a}{E_a} \quad \text{(C.5)}
\]

The reflectance of an interface between an incident medium of admittance \(\eta_0\) and a medium of admittance \(Y\) is equal to:

\[
R = \left(\frac{\eta_0 - Y}{\eta_0 + Y}\right) \left(\frac{\eta_0 - Y^*}{\eta_0 + Y^*}\right) \quad \text{(C.6)}
\]

Equation (C.2) can be normalised by dividing through by \(E_b\):

\[
\begin{bmatrix}
E_a / E_b \\
H_a / E_b
\end{bmatrix} = 
\begin{bmatrix}
B \\
C
\end{bmatrix} = 
\begin{bmatrix}
\cos \delta & (i \sin \delta) / \eta_1 \\
i \eta_1 \sin \delta & \cos \delta
\end{bmatrix}
\begin{bmatrix}
1 \\
\eta_2
\end{bmatrix} \quad \text{(C.7)}
\]
where

\[
\begin{bmatrix}
  B \\
  C
\end{bmatrix}
\]  \hspace{1cm} (C.8)

is known as the characteristic matrix of the assembly. \(B\) and \(C\) represent the normalised electric and magnetic fields at the front interface. From Eqns. (C.5) and (C.7), we can write:

\[
Y = \frac{H_a}{E_a} = \frac{C}{B} = \eta_2 \cos \delta + i \eta_1 \sin \delta \cos \delta + i(\eta_2 / \eta_1) \sin \delta
\]  \hspace{1cm} (C.9)

Hence from Eqns. (C.6) and (C.9), the reflectance can be calculated.

### C.2. The Reflectance of Multiple Thin Films Covering a Substrate

Consider the thin film setup shown in Fig. C.2. The characteristic matrix of the film nearest the substrate is:

\[
\begin{bmatrix}
  \cos \delta_2 & (i \sin \delta_2) / \eta_2 \\
  i \eta_2 \sin \delta_2 & \cos \delta_2
\end{bmatrix}
\]  \hspace{1cm} (C.10)

and

\[
\begin{bmatrix}
  E_b \\
  H_b
\end{bmatrix} = \begin{bmatrix}
  \cos \delta_2 & (i \sin \delta_2) / \eta_2 \\
  i \eta_2 \sin \delta_2 & \cos \delta_2
\end{bmatrix} \begin{bmatrix}
  E_c \\
  H_c
\end{bmatrix}
\]  \hspace{1cm} (C.11)

Applying Eqn. (C.2), \(E_a\) and \(H_a\) can be calculated using:

![Diagram of Incident Plane Wave Normal to Two Thin Films on a Substrate](image)

**FIG. C.2.** Incident plane wave normal to two thin films on a substrate.
Appendix C – Summary of the Optical Transfer Matrix Method

\[
\begin{bmatrix}
E_o \\
H_o
\end{bmatrix} =
\begin{bmatrix}
\cos \delta_1 & (i \sin \delta_1) / \eta_1 \\
i \eta_1 \sin \delta_1 & \cos \delta_1
\end{bmatrix}
\begin{bmatrix}
\cos \delta_2 & (i \sin \delta_2) / \eta_2 \\
i \eta_2 \sin \delta_2 & \cos \delta_2
\end{bmatrix}
\begin{bmatrix}
E_v \\
H_v
\end{bmatrix}
\] (C.12)

and the characteristic matrix of the assembly is:

\[
\begin{bmatrix}
B \\
C
\end{bmatrix} =
\begin{bmatrix}
\cos \delta_1 & (i \sin \delta_1) / \eta_1 \\
i \eta_1 \sin \delta_1 & \cos \delta_1
\end{bmatrix}
\begin{bmatrix}
\cos \delta_2 & (i \sin \delta_2) / \eta_2 \\
i \eta_2 \sin \delta_2 & \cos \delta_2
\end{bmatrix}
\begin{bmatrix}
1 \\
\eta_1
\end{bmatrix}
\] (C.13)

\( Y \) is defined as before \((C/B)\), and the reflectance is calculated using Eqn. (C.6). Hence, for an assembly with \( q \) number of layers, the characteristic matrix is equal to the product of the individual matrices:

\[
\begin{bmatrix}
B \\
C
\end{bmatrix} = \prod_{r=1}^{q} \begin{bmatrix}
\cos \delta_r & (i \sin \delta_r) / \eta_r \\
i \eta_r \sin \delta_r & \cos \delta_r
\end{bmatrix} \begin{bmatrix}
1 \\
\eta_m
\end{bmatrix}
\] (C.14)

where \( \delta_r = 2\pi N_r d / \lambda \) and \( \eta_m \) is the optical admittance of the substrate.

C.3. Calculating the Transmittance and Absorbance of a Thin Film Assembly

The transmittance of a thin film assembly is equal to the ratio of the net irradiance at the exit side of the assembly (denoted the \( k \)th interface) to the net incidence irradiance:

\[
T = \frac{I_k}{I_i} = \frac{\text{Re}(\eta_m)(1 - R)}{\text{Re}(BC^*)}
\] (C.15)

The transmittance is related to the assembly’s reflectance \((R)\) and absorbance \((A)\) via the following equation:

\[
A + R + T = 1
\] (C.16)

Substituting Eqn. (C.15) into (C.16) yields:

\[
A = (1 - R) \left(1 - \frac{\text{Re}(\eta_m)}{\text{Re}(BC^*)}\right)
\] (C.17)

Rewriting the reflectance in terms of \( B \) and \( C \), Eqn. (C.6) can be written as:

\[
R = \left(\frac{\eta_0 B - C}{\eta_0 B + C}\right) \left(\frac{\eta_0 B - C}{\eta_0 B + C}\right)^*
\] (C.18)

Inserting Eqn. (C.18) into Eqns. (C.15) and (C.17) yields:

\[
T = \frac{4 \eta_0 \text{Re}(\eta_m)}{(\eta_0 B + C)(\eta_0 B + C)^*}
\] (C.19)
Appendix C – Summary of the Optical Transfer Matrix Method

\[ A = \frac{4\eta_0 \text{Re}(BC^\ast - \eta_m)}{(\eta_0 B + C)(\eta_0 B + C)^\ast} \quad (C.20) \]

C.4. Concluding Remarks

From the above analysis, the transmittance, absorbance and reflectance of either a singular thin film or a multilayer assembly are calculated directly from the complex refractive index \(N_i\) and the thickness \(d_i\) of the films, as well as the substrate in which the films are deposited on. It is also important to note that \(N_i\) is a function of \(\lambda\). Values for \(n\) and \(k\) used in this study were found in ref. [5].

References

Appendix D

Device Processing Summary

This section summarises the process steps necessary for the fabrication of the tunable filter on Si substrates described in Section 5.7.

D.1. Si Wafer Clean

<table>
<thead>
<tr>
<th>Step</th>
<th>Step Details</th>
</tr>
</thead>
<tbody>
<tr>
<td>Piranha etch</td>
<td>1:1 H₂SO₄:H₂O₂</td>
</tr>
<tr>
<td></td>
<td>Immerse wafer once H₂O₂ is added to H₂SO₄ for 10 min</td>
</tr>
<tr>
<td>Soak in hot trichloroethylene</td>
<td>50 °C for 5 min</td>
</tr>
<tr>
<td>Soak in hot acetone</td>
<td>50 °C for 5 min</td>
</tr>
<tr>
<td>Soak in hot methanol</td>
<td>50 °C for 5 min</td>
</tr>
<tr>
<td>Soak in hot isopropanol</td>
<td>50 °C for 5 min</td>
</tr>
<tr>
<td>Blow dry N₂</td>
<td></td>
</tr>
</tbody>
</table>

D.2. Bottom Electrode Deposition

<table>
<thead>
<tr>
<th>Step</th>
<th>Step Details</th>
</tr>
</thead>
<tbody>
<tr>
<td>Dehydration</td>
<td>110 °C for 1 min</td>
</tr>
<tr>
<td>Spin AZ 2035 resist</td>
<td>3000 rpm for 40 s</td>
</tr>
<tr>
<td>Bake</td>
<td>110 °C for 1 min</td>
</tr>
<tr>
<td>Mask 1 exposure</td>
<td>20 s exposure on Karl Suss mask aligner</td>
</tr>
<tr>
<td>Bake</td>
<td>110 °C for 1 min</td>
</tr>
<tr>
<td>Develop AZ 300 MIF</td>
<td>90 s, undiluted</td>
</tr>
<tr>
<td>Rinse and blow dry N₂</td>
<td></td>
</tr>
<tr>
<td>Load sample in thermal evaporator</td>
<td>Evacuate &lt;2×10⁻⁶ mBar</td>
</tr>
<tr>
<td>Evaporate 50/1000 Å Cr/Au</td>
<td>1 Å/s for Cr, 10 Å/s for Au</td>
</tr>
<tr>
<td>Cool down, remove from chamber</td>
<td></td>
</tr>
<tr>
<td>Lift-off</td>
<td>Soak in undiluted acetone</td>
</tr>
<tr>
<td>Solvent rinse</td>
<td>Soak in acetone, methanol and isopropanol</td>
</tr>
<tr>
<td>Blow dry N₂</td>
<td></td>
</tr>
</tbody>
</table>
### D.3. Bottom Mirror Deposition

<table>
<thead>
<tr>
<th>Step</th>
<th>Step Details</th>
</tr>
</thead>
<tbody>
<tr>
<td>Dehydration</td>
<td>110 °C for 1 min</td>
</tr>
<tr>
<td>Load sample in E-beam</td>
<td>Evacuate &lt;2×10^{-6} mBar</td>
</tr>
<tr>
<td>Evaporate 255/113/255/113 Å SiO/Ge/SiO/Ge</td>
<td>10 Å/s for all materials</td>
</tr>
<tr>
<td>Cool down, remove from chamber</td>
<td></td>
</tr>
</tbody>
</table>

### D.4. Bottom Mirror Etch for Bottom Electrode Contact

<table>
<thead>
<tr>
<th>Step</th>
<th>Step Details</th>
</tr>
</thead>
<tbody>
<tr>
<td>Dehydration</td>
<td>110 °C for 1 min</td>
</tr>
<tr>
<td>Spin 4562 resist</td>
<td>4000 rpm for 40 s</td>
</tr>
<tr>
<td>Bake</td>
<td>100 °C for 90 s</td>
</tr>
<tr>
<td>Mask 2 exposure</td>
<td>40 s exposure on Karl Suss mask aligner</td>
</tr>
<tr>
<td>Develop 400 K</td>
<td>60 s, dilute with H₂O 1:3</td>
</tr>
<tr>
<td>Rinse and Blow dry N₂</td>
<td></td>
</tr>
<tr>
<td>Load sample in Plasmalab 100 RIE CF₄ 10 sccm, 10 mTorr, 20 °C, 50/50 W RF/ICP, 30 min</td>
<td></td>
</tr>
<tr>
<td>Remove resist in acetone</td>
<td></td>
</tr>
<tr>
<td>Solvent clean</td>
<td>Soak in acetone, methanol and isopropanol</td>
</tr>
<tr>
<td>Blow dry N₂</td>
<td></td>
</tr>
</tbody>
</table>

### D.5. First Sacrificial Layer Deposition

<table>
<thead>
<tr>
<th>Step</th>
<th>Step Details</th>
</tr>
</thead>
<tbody>
<tr>
<td>Dehydration</td>
<td>130 °C for 1 min</td>
</tr>
<tr>
<td>Spin Polyimide 2616</td>
<td>6000 rpm for 40 s</td>
</tr>
<tr>
<td>Bake</td>
<td>130 °C for 20 min</td>
</tr>
</tbody>
</table>

### D.6. SiₙHₓ Compensating Layer Deposition

<table>
<thead>
<tr>
<th>Step</th>
<th>Step Details</th>
</tr>
</thead>
<tbody>
<tr>
<td>Deposit SiₙHₓ, using Oxford Instruments PECVD 80 System</td>
<td>Evacuate chamber for 15 min</td>
</tr>
<tr>
<td></td>
<td>Deposit at 125 °C</td>
</tr>
<tr>
<td></td>
<td>SiH₄:NH₃:N₂ flow ratios 5:50:100 sccm</td>
</tr>
<tr>
<td></td>
<td>600 mTorr</td>
</tr>
<tr>
<td></td>
<td>100 W RF</td>
</tr>
<tr>
<td></td>
<td>4 min, 35 sec</td>
</tr>
</tbody>
</table>

### D.7. Top Mirror Deposition

<table>
<thead>
<tr>
<th>Step</th>
<th>Step Details</th>
</tr>
</thead>
<tbody>
<tr>
<td>Dehydration</td>
<td>110 °C for 1 min</td>
</tr>
<tr>
<td>Load sample in E-beam</td>
<td>Evacuate &lt;2×10^{-6} mBar</td>
</tr>
</tbody>
</table>
Evaporate 113/255/113 Å
Ge/SiO/Ge
Cool down, remove from chamber

D.8. Top Mirror Etch

<table>
<thead>
<tr>
<th>Step</th>
<th>Step Details</th>
</tr>
</thead>
<tbody>
<tr>
<td>Dehydration</td>
<td>110 °C for 1 min</td>
</tr>
<tr>
<td>Spin HPr resist</td>
<td>3000 rpm for 40 s</td>
</tr>
<tr>
<td>Bake</td>
<td>100 °C for 60 s</td>
</tr>
<tr>
<td><strong>Mask 3 exposure</strong></td>
<td>40 s exposure on Karl Suss mask aligner</td>
</tr>
<tr>
<td>Develop HPr developer</td>
<td>60 s, dilute with H2O 1:3</td>
</tr>
<tr>
<td>Rinse and Blow dry N₂</td>
<td></td>
</tr>
<tr>
<td>Load sample in Plasmalab 100 RIE</td>
<td>CF₄ 10 sccm, 10 mTorr, 20 °C, 50/50 W RF/ICP, 25 min</td>
</tr>
<tr>
<td>Remove resist in acetone</td>
<td></td>
</tr>
<tr>
<td>Solvent clean</td>
<td>Soak in acetone, methanol and isopropanol</td>
</tr>
<tr>
<td>Blow dry N₂</td>
<td></td>
</tr>
</tbody>
</table>

D.9. Second Sacrificial Layer Deposition

<table>
<thead>
<tr>
<th>Step</th>
<th>Step Details</th>
</tr>
</thead>
<tbody>
<tr>
<td>Dehydration</td>
<td>100 °C for 1 min</td>
</tr>
<tr>
<td>Spin Polyimide 2737</td>
<td>9000 rpm for 40 s</td>
</tr>
<tr>
<td>Bake</td>
<td>100 °C for 1 min</td>
</tr>
<tr>
<td>Bake</td>
<td>130 °C for 1 min</td>
</tr>
<tr>
<td>Flood Expose</td>
<td>40 s exposure on Karl Suss mask aligner</td>
</tr>
</tbody>
</table>

D.10. Sacrificial Layer Patterning

<table>
<thead>
<tr>
<th>Step</th>
<th>Step Details</th>
</tr>
</thead>
<tbody>
<tr>
<td>Dehydration</td>
<td>110 °C for 1 min</td>
</tr>
<tr>
<td>Spin AZ 2035 resist</td>
<td>3000 rpm for 40 s</td>
</tr>
<tr>
<td>Bake</td>
<td>110 °C for 1 min</td>
</tr>
<tr>
<td><strong>Mask 4 exposure</strong></td>
<td>20 s exposure on Karl Suss mask aligner</td>
</tr>
<tr>
<td>Bake</td>
<td>110 °C for 1 min</td>
</tr>
<tr>
<td>Develop AZ 300 MIF</td>
<td>90 s, undiluted</td>
</tr>
<tr>
<td>Rinse and blow dry N₂</td>
<td></td>
</tr>
<tr>
<td>Load sample in Plasmalab 100 RIE</td>
<td>O₂/CF₄ 20/1 sccm, 20 mTorr, 20 °C, 75 W RF, 20 min</td>
</tr>
<tr>
<td>Remove resist in acetone</td>
<td></td>
</tr>
<tr>
<td>Solvent clean</td>
<td>Soak in acetone, methanol and isopropanol</td>
</tr>
<tr>
<td>Blow dry N₂</td>
<td></td>
</tr>
<tr>
<td>Dehydration</td>
<td>110 °C for 1 min</td>
</tr>
<tr>
<td>Spin AZ 2035 resist</td>
<td>3000 rpm for 40 s</td>
</tr>
<tr>
<td>Bake</td>
<td>110 °C for 1 min</td>
</tr>
<tr>
<td><strong>Mask 5 exposure</strong></td>
<td>20 s exposure on Karl Suss mask aligner</td>
</tr>
<tr>
<td>Bake</td>
<td>110 °C for 1 min</td>
</tr>
<tr>
<td>Develop AZ 300 MIF</td>
<td>90 s, undiluted</td>
</tr>
<tr>
<td>Rinse and blow dry N₂</td>
<td></td>
</tr>
<tr>
<td>Load sample in Plasmalab 100 RIE</td>
<td>O₂/CF₄ 20/1 sccm, 20 mTorr, 20 °C, 75 W RF, 30 min</td>
</tr>
<tr>
<td>Remove resist in acetone</td>
<td></td>
</tr>
</tbody>
</table>
### D.11. SiN\textsubscript{x}H\textsubscript{y} Flexure Layer Deposition

<table>
<thead>
<tr>
<th>Step</th>
<th>Step Details</th>
</tr>
</thead>
<tbody>
<tr>
<td>Deposit SiN\textsubscript{x}H\textsubscript{y} using Oxford Instruments PECVD 80 System</td>
<td>Evacuate chamber for 15 min&lt;br&gt;Deposit at 125 °C&lt;br&gt;SiH\textsubscript{4}:NH\textsubscript{3}:N\textsubscript{2} flow ratios 5:50:100 sccm&lt;br&gt;875 mTorr&lt;br&gt;75 W RF&lt;br&gt;15 min</td>
</tr>
</tbody>
</table>

### D.12. SiN\textsubscript{x}H\textsubscript{y} Etch for Bottom Electrode Contact

<table>
<thead>
<tr>
<th>Step</th>
<th>Step Details</th>
</tr>
</thead>
<tbody>
<tr>
<td>Dehydration</td>
<td>110 °C for 1 min</td>
</tr>
<tr>
<td>Spin 4562 resist</td>
<td>4000 rpm for 40 s</td>
</tr>
<tr>
<td>Bake</td>
<td>100 °C for 90 s</td>
</tr>
<tr>
<td><strong>Mask 6</strong> exposure</td>
<td>40 s exposure on Karl Suss mask aligner</td>
</tr>
<tr>
<td>Develop 400 K</td>
<td>60 s, dilute with H\textsubscript{2}O 1:3</td>
</tr>
<tr>
<td>Rinse and Blow dry N\textsubscript{2}</td>
<td></td>
</tr>
<tr>
<td>Load sample in Plasmalab 100 RIE</td>
<td>CF\textsubscript{4} 10 sccm, 10 mTorr, 20 °C, 50/50 W RF/ICP, 10 min</td>
</tr>
<tr>
<td>Remove resist in acetone</td>
<td></td>
</tr>
<tr>
<td>Solvent clean</td>
<td>Soak in acetone, methanol and isopropanol</td>
</tr>
<tr>
<td>Blow dry N\textsubscript{2}</td>
<td></td>
</tr>
</tbody>
</table>

### D.13. Top Electrode Deposition

<table>
<thead>
<tr>
<th>Step</th>
<th>Step Details</th>
</tr>
</thead>
<tbody>
<tr>
<td>Dehydration</td>
<td>110 °C for 1 min</td>
</tr>
<tr>
<td>Spin AZ 2035 resist</td>
<td>3000 rpm for 40 s</td>
</tr>
<tr>
<td>Bake</td>
<td>110 °C for 1 min</td>
</tr>
<tr>
<td><strong>Mask 7</strong> exposure</td>
<td>20 s exposure on Karl Suss mask aligner</td>
</tr>
<tr>
<td>Bake</td>
<td>110 °C for 1 min</td>
</tr>
<tr>
<td>Develop AZ 300 MIF</td>
<td>90 s, undiluted</td>
</tr>
<tr>
<td>Rinse and blow dry N\textsubscript{2}</td>
<td></td>
</tr>
<tr>
<td>Load sample in thermal evaporator</td>
<td>Evacuate &lt;2×10\textsuperscript{-6} mBar</td>
</tr>
<tr>
<td>Evaporate 50/500 Å Cr/Au</td>
<td>1 Å/s for Cr, 10 Å/s for Au</td>
</tr>
<tr>
<td>Cool down, remove from chamber</td>
<td></td>
</tr>
<tr>
<td>Lift-off</td>
<td>Soak in undiluted acetone</td>
</tr>
<tr>
<td>Solvent rinse</td>
<td>Soak in acetone, methanol and isopropanol</td>
</tr>
<tr>
<td>Blow dry N\textsubscript{2}</td>
<td></td>
</tr>
<tr>
<td>Dehydration</td>
<td>110 °C for 1 min</td>
</tr>
<tr>
<td>Spin AZ 2035 resist</td>
<td>3000 rpm for 40 s</td>
</tr>
<tr>
<td>Bake</td>
<td>110 °C for 1 min</td>
</tr>
<tr>
<td><strong>Mask 8</strong> exposure</td>
<td>20 s exposure on Karl Suss mask aligner</td>
</tr>
<tr>
<td>Bake</td>
<td>110 °C for 1 min</td>
</tr>
<tr>
<td>Develop AZ 300 MIF</td>
<td>90 s, undiluted</td>
</tr>
<tr>
<td>Rinse and blow dry N\textsubscript{2}</td>
<td></td>
</tr>
</tbody>
</table>
Appendix D – Device Processing Summary

<table>
<thead>
<tr>
<th>Step Details</th>
</tr>
</thead>
<tbody>
<tr>
<td>Load sample in thermal evaporator</td>
</tr>
<tr>
<td>Evaporate 50/3000 Å Cr/Au</td>
</tr>
<tr>
<td>Cool down, remove from chamber</td>
</tr>
<tr>
<td>Lift-off</td>
</tr>
<tr>
<td>Solvent rinse</td>
</tr>
<tr>
<td>Blow dry N₂</td>
</tr>
</tbody>
</table>

**D.14. SiNₓHᵧ Flexure Etch**

<table>
<thead>
<tr>
<th>Step</th>
<th>Step Details</th>
</tr>
</thead>
<tbody>
<tr>
<td>Load sample in Plasmalab 100 RIE</td>
<td>CF₄ 10 sccm, 10 mTorr, 20 °C, 50/50 W RF/ICP, 10 min (Au layers define SiNₓHᵧ pattern) O₂ 80 sccm, 20 mTorr, 20 °C, 75 W RF, 5 min (Oxygen ash)</td>
</tr>
<tr>
<td>Solvent clean</td>
<td>Soak in acetone, methanol and isopropanol</td>
</tr>
<tr>
<td>Blow dry N₂</td>
<td></td>
</tr>
</tbody>
</table>

**D.15. Filter Release and Drying**

<table>
<thead>
<tr>
<th>Step</th>
<th>Step Details</th>
</tr>
</thead>
<tbody>
<tr>
<td>Remove polyimide layers</td>
<td>Solution of HPr: photosensitive developer – 1:2</td>
</tr>
<tr>
<td>Rinse in H₂O</td>
<td>10 mins running H₂O in Petri dish in large rectangular glass dish</td>
</tr>
<tr>
<td>Dehydrate Step 1</td>
<td>1:2 acetone:H₂O 10mins</td>
</tr>
<tr>
<td>Dehydrate Step 2</td>
<td>2:1 acetone:H₂O 10mins</td>
</tr>
<tr>
<td>Dehydrate Step 3</td>
<td>Acetone 10mins</td>
</tr>
<tr>
<td>Dehydrate Step 4</td>
<td>Methanol 10 mins</td>
</tr>
<tr>
<td>Immerse in fresh methanol – transfer to critical point dryer</td>
<td></td>
</tr>
<tr>
<td>Run critical point dryer process</td>
<td></td>
</tr>
</tbody>
</table>