Global and local residual stress in silicon carbide films produced by plasma-enhanced chemical vapor deposition

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Abstract

Residual stress in films is normally created in the process and results in the unwanted deformation or failure of structures. It could be characterized by measuring the changes in the radius of curvature of an overall stressed film on a substrate, i.e. global residual stress or by the curvature change of locally micromachined suspended cantilever beams, i.e. local residual stress. In this study, global and local residual stress behavior of PECVD silicon carbide films is investigated by deposition parameters and post rapid thermal annealing to obtain low stress and flat free-standing microstructure. In the as-deposited state, a low global compressive film with a stress level of \(-160\) MPa has been obtained at medium pressure of 147 Pa and at low substrate temperature of 250 °C. Then the stress can be further reduced to less than 100 MPa with low local gradient stress by a post annealing at 420 °C. The final stress could be effectively controlled by a discrete annealing method, approaching a steady state only after several minutes. It is comparable with the complementary metal-oxide-semiconductor (CMOS) process. As the annealing temperature increases, a stress transition from the compressive region to the tensile one takes place. The stress relaxation is attributed to the dissociation of Si–H and C–H bonds, and incorporation of hydrogen molecules in the film for the formation of tensile Si–C bonds with large electronegativity difference or the outdiffusion of hydrogen molecules to increase the bond length between atoms for a more tensile stress. Self-deformed micromachined cantilevers are fabricated to exhibit the local gradient stress from the curvature of beams. The gradient stress is negative for the cantilevers with negative or downward curvature in the as-deposited state and then transforms to positive for the cantilevers with positive or upward curvature after annealing. The gradient stress varied from negative to positive is attributed to the difference of stress variation between the free surface and the interface of the film and substrate. A low stress film of less than 100 MPa with a near-zero gradient component could be adjusted by as-deposited process parameters and post annealing for the fabrication of the flat suspended MEMS structure.

Keywords: Residual stress; SiC; PECVD; MEMS

1. Introduction

Silicon carbide has become a potential material with very attractive properties for the application in large band gap electronics and micro-electro-mechanical systems (MEMS). The functions of films for MEMS include supporting membranes [1], coating in microchannels [2], passivation layer on microelectronics [2,3], etching mask [4,5] and thermistor material [6] due to their high mechanical strength, extremely chemical resistance in various etchants and excellent thermal conductivity. With respect to the mechanical properties of the silicon carbide films for the use in sensors or free-standing MEMS structures, one important issue is the residual stress, which is normally created in the process and results in the unwanted deformation or failure of structures. For example, the built-in stress may change the mechanical response or the resonant frequency of thin-film structures, even lead to cracking, buckling or delamination of films. Therefore, it is necessary to reduce and control the...
residual stress of thin films for the design and performance of microdevices.

In the past, several techniques have been used to obtain silicon carbide films with extreme high melting point, such as conventional chemical vapor deposition (CVD), gas-source molecular beam epitaxy (MBE) and electron cyclotron resonance (ECR) plasma [7]. These techniques are performed at high temperature above 600 °C. However, it is necessary to fabricate MEMS structures at low temperature for IC compatible process integration [1]. It is feasible to deposit silicon carbide films at low temperatures using plasma-enhanced chemical vapor deposition (PECVD) with a lot of variable parameters for film formation [8]. Therefore, it provides wide possibilities to tailor materials properties through correlations between the deposition parameters, microstructure control and the properties of films. The effects of deposition parameters [4,7,9–11] and annealing [12–14] on the variation of residual stress were investigated in the previous publications. However, the stress is usually measured by bent curvature method in large scale of several cm or more, which represents the global stress of the whole film. Local stress distribution on freestanding structures of silicon carbide films is less mentioned and rarely correlated to the global stress.

In this study, low stress silicon carbide films were controlled by adjusting the deposition parameters and an effective method through post rapid thermal annealing. The origin of residual stress and the mechanisms governing the stress relaxation during annealing were investigated from the viewpoint of chemical bonds and hydrogen molecules diffusion. Moreover, self-deformed micromachined cantilevers were fabricated to display the local stress distribution.

2. Experimental procedures

The amorphous silicon carbide (a-Si$_x$C$_{1-x}$;H) films were deposited by plasma-enhanced chemical vapor deposition system (STS Multiplex CVD, UK). The schematic diagram of the apparatus was shown in Fig. 1. Four-inch boron-doped p-Si(100) wafers of 4–10 Ω cm were initially cleaned in a solution of H$_2$SO$_4$/H$_2$O$_2$ in the ratio of 3:1. They were put into the load-lock chamber followed by rotary pumping and then transferred to reaction chamber for deposition. The reaction gases were introduced into the reaction chamber through a showerhead array in the upper electrode. They were consisted of a gas mixture of SiH$_4$ and Ar with a fixed SiH$_4$/Ar flow rate ratio at 20:1000, and the CH$_4$ gas with a fixed flow rate at 400 cm$^3$/min. The plasma was activated by 380 kHz RF power supply at 150 W. The thickness of films is concerned with the residual stress and controlled between 400 and 500 nm during deposition in this study. Two major operating parameters for stress control are the substrate temperature and total pressure. The total reaction pressure was controlled at 80–213 Pa while the substrate temperature was controlled at 250–350 °C for as-deposited stress formation. Post-annealing treatment was performed by rapid thermal annealing (RTA, AG Associate-310, USA) in a N$_2$ atmosphere to investigate the further stress variation of films. Local stress variation was examined by suspended cantilever beams. The fabrication process of suspended cantilevers was illustrated in Fig. 2. Silicon carbide was patterned by conventional photolithography and reactive ion etching (RIE, Trion Phantom system, USA) in CF$_4$ gas. Silicon beneath the films was removed by wet etching in 33 wt.% KOH solution at 85 °C. The deformation of released
The microstructure was examined by scanning electron microscopy (SEM, Hitachi S-3500H, Japan).

The stress of a-Si$_x$C$_{1-x}$:H films were determined using a Tencor FLX-2320 system (Tencor Instruments, USA), which measures the changes in the radius of curvature of a substrate caused by deposition of a stressed thin film. The stress ($\sigma$) is calculated using the following formula:

$$\sigma = \frac{Eh^2}{(1-\nu)6Rt}$$

where $E$ and $\nu$ are Young’s modulus and Poisson’s ratio of the substrate, respectively. The $E/(1-\nu)$ value is $1.805 \times 10^{11}$ Pa for 4 in. silicon wafers. Both symbols of $h$ and $t$ are the individual thickness of substrates and films. $R$ is radius of curvature determined from the equation as follows:

$$\frac{1}{R} = \frac{1}{R_1} - \frac{1}{R_2}$$

where $R_1$ is the average radius of the bare substrate. The substrate deforms to a new radius $R_2$ after the film is deposited.

In consideration of the relationship between the stress and the chemical bonding, the IR absorption measurements were carried out at room temperature with a Fourier transform infrared (FTIR) spectroscopy (Nicolet 460, USA) in the 400–4000 cm$^{-1}$ region and the resolution was 4 cm$^{-1}$. The IR spectra were recorded relative to an uncoated silicon substrate. According to the integrated absorbances and the inverse cross section of the C–H, Si–H and Si–C stretching bonds, the bond densities can be estimated by

$$N = A_s \int \frac{\alpha(v)}{\nu} d\nu = \frac{A_s}{\nu_0} \int \alpha(v) d\nu$$

where $A_s$ is the inversion absorption cross section of the considered mode and these for C–H, Si–H and Si–C bands are $(1.35 \pm 0.35) \times 10^{21}$, $(1.4 \pm 0.1) \times 10^{20}$ and $2.13 \times 10^{19}$ cm$^{-2}$, respectively. $\nu_0$ is the wavenumber corresponding to the maximum absorption of the band and $\alpha(v)$ is the absorption coefficient [13,15,16].

To understand the hydrogen effect on residual stress, the hydrogen content in the films was determined by Elastic Recoil Detection Analyses (ERDA). A $^4$He$^+$ ion ($\alpha$ particle) beam with incident energy of 2.9 MeV was used in ERDA. The incidence and the emergence angles were set at $15^\circ$ and the solid angle of detector was 0.97 m.s.r. Incident dose of 20 $\mu$C was used and the detector resolution energy was 30 keV.

3. Results and discussion

Fig. 3 shows the relationship between residual stress of as-deposited films and total pressures at different substrate temperatures of 250 °C, 300 °C and 350 °C, respectively. The as-deposited films were under compressive stress in the range from $-510$ MPa to $-160$ MPa. The stress values are relatively lower as compared to those in literature [7,9,12,13]. It is interesting to note that the lowest stress of films occurs at a medium pressure of 1100 mTorr during the amorphous silicon carbide (a-Si$_x$C$_{1-x}$:H) deposition at 80, 147 and 213 Pa, respectively. Residual stress is supposed to be related to the incorporation of hydrogen in the films [12] and also to the extent of the ion bombardment [10]. The stress variation shows an initial decrease from 80 Pa to 147 Pa, and then increase from 147 Pa to 213 Pa.
may imply the competition of two factors, that is, the incorporation of hydrogen and the ion bombardment affecting the residual stress of films. At lower pressure, ion bombardment dominates over the other. The ion bombardment is concerned with the ion energy in plasma activated by 380 kHz RF power at 150 W. The mean free path of ion is longer at lower pressure, so ions gain more energy before a collision to contribute more compressive stress to the film. Consequently, the effect of ion bombardment on the growing film is more pronounced at 80 Pa and the resultant compressive stress is relatively higher. The quantity of ion energy and ion flux on the film surface could not be measured due to the absence of port for an instrument attached to the deposition chamber. However, the effect of hydrogen incorporation becomes predominant as the pressure is increased with more hydrogen-contained reactants of SiH$_4$ and CH$_4$ gases. With higher concentration of hydrogen incorporation, atoms in the film will be closer with shorter bond length, leading to a higher compressive stress. Therefore, a relatively lower stress film is developed at a medium pressure of 147 Pa. Another factor influencing variation of stress is the substrate temperature. The films with higher residual stress were produced as the substrate temperature increases. The thermal stress is contributed from two competing factors: one is the difference of coefficient of thermal expansion (CTE) between the a-Si$_x$C$_{1-x}$:H film and Si substrate while the other is the atomic density of silicon carbide. The CTE of a-Si$_x$C$_{1-x}$:H is about $3.6 \times 10^{-6}$/K [17] larger than that of Si of about $2.6 \times 10^{-6}$/K. So, the CTE difference contributes a tensile stress as the film is deposited at high temperature, then cooled to room temperature. The atomic density of a film is related to the deposition rate and affected by the reactive fluxes of neutral radicals and positive ions arriving at the growing surface. The substrate temperature effects on the fluxes of radicals and ions are formulated as follows [18].

$$J_n \propto 1/T$$

(4)

$$J_i \propto n_i(T)^{1/2}$$

(5)

$$J = J_n + J_i$$

(6)

where $J_n$ and $J_i$ are the fluxes of neutral radicals and positive ions, respectively, $T$ is reaction temperature and $n_i$ is concentration of ions. At a lower substrate temperatures, the radical mechanism is dominant. Since the mechanism is less selective in bonding with the substrate surface, it exhibits a higher deposition rate to get a lower density [19]. When the substrate temperature is raised, the ionic mechanism is predominant. Some ionic species bond with the surface to become a part of the growing film, and others are bounced off after neutralizing. Since the number of sites available for bonding formation is limited, the deposition rate is lower. Moreover, due to the preference of charged species attaching to the surface, the bombardment of the surface by electrons and ions speeds up the rearrangement of adsorbed atoms on the surface, resulting in a denser film. The atomic density of films deposited at 250, 300 and 350 °C is measured and estimated at $(6.9 \pm 0.35) \times 10^{22}$, $(7.5 \pm 0.38) \times 10^{22}$, $(7.8 \pm 0.39) \times 10^{22}$ atoms/cm$^3$, respectively. The denser film with shorter bond length formed at a higher substrate temperature leads to a higher compressive stress contributing to the thermal stress. The atomic density effect on the thermal stress is overriding the CTE difference effect in the experimental data at 250–350 °C. So, the as-deposited film with the lowest stress level of −160 MPa could be obtained at a medium pressure of 147 Pa and at a low substrate temperature of 250 °C.

The residual stress of a-Si$_x$C$_{1-x}$:H film is controlled not only by process parameters but also by post annealing to further reduce and adjust the residual stress (<100 MPa) for better released MEMS structure fabrication. Fig. 4 shows the stress evaluation as a function of the annealing temperatures at a constant time of 30 s. The open symbols indicate the initial stress of as-deposited films at 147 Pa at 250 °C in square and at 350 °C in circle, respectively, and solid symbols represent the stress after annealing. It appears that the stress is shifted from the compressive region to the tensile region as the annealing temperature increases. The transition temperatures from compressive stress to tensile one for films deposited at 250 °C and 350 °C are around 400 °C and 500 °C, respectively. The higher the deposition temperature, the higher the annealing temperature required to relax the compressive stress. The process of stress relaxation may be distinguished into two regions: (i) from the as-deposited states to the vicinity of 500 °C, there is a slow evolution of stress. (ii) Above 500 °C, an abrupt stress variation is observed in the tensile region. Stress variation is attributed to the incorporation of hydrogen in the films [12] and the elimination of hydrogen can be promoted by annealing. Fig. 5 shows the relationship between the total hydrogen contents

![Image](332x100 to 534x243)

Fig. 4. Dependence of residual stress on rapid thermal annealing temperatures at a constant time of 30 s. Open marks indicate the initial stress of as-deposited films at 250 °C in square and at 350 °C in circle, respectively, and solid marks represent the stress after annealing.

\[J_n \propto \frac{1}{T}\]  
\[J_i \propto n_i(T)^{1/2}\]  
\[J = J_n + J_i\]
measured by ERDA and the RTA temperature of a-Si$_2$C$_{1-x}$H film deposited at a pressure of 147 Pa at 250 °C. The open square symbol indicates the initial hydrogen contents of as-deposited films and solid circle symbols represent the hydrogen contents after annealing at temperatures of 400–650 °C. The hydrogen contents remain nearly constant until the annealing temperature reaches 500 °C. For temperature up to 600 °C and more, the hydrogen contents decrease rapidly, suggesting the outdiffusion of all hydrogen molecules. It is in agreement with the stress variation in Fig. 4. The relationship between the stress and the chemical bonding can be established by the IR absorption measurements at room temperature with a Fourier transform infrared (FTIR) spectroscopy. The densities of C–H, Si–H and Si–C stretching bonds can be estimated by Eq. (3) as mentioned in experiments. Fig. 6(a)–(c) show the variation of the bond densities of C–H, Si–H and Si–C stretching modes of the films annealed from 400 °C to 650 °C at different deposition substrate temperatures of 250 °C, 300 °C and 350 °C, respectively. The open symbols also indicate the initial stress of as-deposited films and solid symbols represent the stress after annealing. The C–H bond density in square symbols exhibits an appreciable decrease in the vicinity of 500–600 °C, while it is almost constant for temperatures below 500 °C. The behavior of the Si–H bond density in circular symbols below 500 °C is similar to the trend in the C–H bond case. As to the Si–C bond density in triangular symbols, it starts to increase at 400 °C and shows a nearly linear increase with increasing temperatures. When the annealing temperature is as high as 650 °C, the hydrogenated Si–H and C–H bonds almost vanish, suggesting the formation of hydrogen molecules incorporated in the films or the outdiffusion of hydrogen molecules. Similar phenomena were also found in El Khakani et al.’s work [12], in which the C–H bond density showed an appreciable decrease in a narrow temperature region around 550–650 °C and the Si–H bond density continuously decreased for annealing temperature higher than 450 or 500 °C. The hydrogen atoms resulting from this breaking of hydrogenated bonds can either combine to form hydrogen molecules which are expected to outdiffuse towards the free surface of the film exposed to atmosphere during the annealing, or be trapped in the Si or C dangling bonds to form new hydrogenated bonds. From the as-deposited states to the vicinity of 500 °C, the relief of stress may be contributed to the relaxation and rearrangement of structures without composition and bonding changes, resulting in a more stable structure [20]. Above 500 °C, the substantial increase in tensile stress may be linked to the breaking of C–H and Si–H bonds and the formation of Si–C bond with large difference in electronegativities. Because the dissociation of the hydrogenated bonds increases obviously above 500 °C, it will lead to the additional Si–C bond formation and hydrogen incorporation as well as outdiffusion. The

Fig. 5. The relationship between the total hydrogen contents and the RTA temperature of a-SiC:H film deposited at 250 °C.

Fig. 6. Variation of the C–H, Si–H and Si–C bond densities with the rapid thermal annealing temperatures from 400 °C to 650 °C at different deposition substrate temperatures: (a) 250 °C, (b) 300 °C and (c) 350 °C, respectively. The deposition pressure is fixed at 147 Pa.
Due to the difference in electronegativities of Si and C atoms much larger than that in Si and H atoms or in C and H atoms, the interaction between Si and C atoms is enhanced and the atoms in films are prone to contract, resulting in a tensile stress to compensate the compressive stress. In addition, the outdiffusion of hydrogen molecules results in the increasing bond length between atoms and molecules and therefore generates an additional tensile stress to compensate the compressive stress. In the microscopic viewpoint, the amorphous films deposited by PECVD at low temperature have a smaller atomic surface mobility and generally lead to a porous microstructure. As

![Graph showing stress variation](image)

**Fig. 7.** Stress variation of the film deposited at 300 °C and annealed at 420 °C as a function of the accumulative annealing time from 30 s to 3 min.

![Images of microstructures](image)

**Fig. 8.** Substantial stress-induced deformation of microstructures under highly compressive stress: (a) spiral coils, (b) suspended membrane and (c) microbridges. Successful fabrication of microstructures without stress-induced deformation: (d) spiral coils, (e) suspended membrane and (f) microbridges.
temperature increases by annealing, the atomic surface mobility is enhanced and hydrogen molecules can diffuse out due to the increasing chemical potential. Then shrinkage of voids takes place to reduce the surface energy, and the bond length between atoms and molecules increases to induce a tensile residual stress. Variations of residual stress and bond densities during annealing indicate that the hydrogenated bonds are responsible for the stress relaxation in films.

On the basis of the observation on the stress releasing by rapid thermal annealing described above, it is not feasible for films with higher initial stress to approach a lower stress region without a higher annealing temperature. The higher the annealing temperature, the larger the temperature limitation of the front-process fabrication. In order to eliminate residual stress at lower annealing temperatures less than 450 °C, a longer period of annealing is adopted at 400 °C or 420 °C, which is used in semiconductor fabrication to decrease defects and to enhance performance. Fig. 7 shows the relationship between stress of the film deposited at 300 °C and annealed at 420 °C with the accumulative annealing time from 30 s to 3 min. The film is annealed by a discrete annealing process composed of several steps. The interval between two successive steps is 0.5 min. It is apparent that the compressive stress is decreased further with increasing annealing time and approaches to a steady state around −50 MPa after several minutes. Thus, the films with higher initial stress after deposition could be decreased to a lower stress with annealing at a low temperature around 400 °C. It is feasible for MEMS microstructure fabrication to control the final stress of films by post rapid thermal annealing. In fact, nearly half of the stress is relaxed in the first step at 0.5 min, i.e. the stress level is reduced from −220 MPa to −125 MPa, as shown in Fig. 7.

It is believed that the residual stress may bring about deformation, such as bending, buckling or twisting towards micromachined structures, particularly the floating structures suspending on the cavity. Fig. 8 illustrates several structures used in MEMS, including microbridges, spiral coils and suspended membranes. Fig. 8(a)–(c) represent the structures which would fail due to an unwanted deformation resulting from a rather high stress around −300 MPa. However, when the stress can be decreased or controlled adequately, the structure will work and relax the built-in stress caused by the deposition of other materials. Fig. 8(d)–(f) show the structures corresponding to Fig. 8(a)–(c), respectively, in which the initial stress after fabrication can be relaxed to around −100 MPa and no stress-induced deformation is observed. The residual stress mentioned above is the global stress measured by bending curvature method when the film is still constrained on the substrate. In general, the global residual stress constitutes uniform and gradient components [23] as illustrated in Fig. 9. The relation can be expressed using the equation as follows [24]:

$$\sigma_{\text{total}} = \sigma_0 + \sigma_1 \left( \frac{\gamma}{h/2} \right)$$

where $\sigma_0$ and $\sigma_1$ represent the uniform and gradient residual stress, respectively, and $h$ is the thickness of film. The uniform component can be relieved through the free end of the cantilever [24], causing a boundary effect on pre-bending as shown in Fig. 10(a). In addition, the cantilever will be bent by the gradient component, resulting in a bending effect on deflection orientation as illustrated in Fig. 10(b). Hence, self-deformed micromachined cantilevers are fabricated in this study to observe the gradient stress ($\sigma_1$) from the curvature of beams and the correlation is expressed as

$$\rho = \frac{Eh}{\sigma_1}$$

where $\rho$ is the radius of curvature and $h$ is the thickness of film. $E=E_f(1-\nu_f)$ is the biaxial modulus, and $E_f$ and $\nu_f$
are the elastic modulus and Poisson’s ratio, respectively, of the film [23].

Cantilevers of different lengths and widths were used to check the uniformity of the fabrication and the repeatability and consistency of the measurements. According to Eq. (8), the deflection curvature would be the same for the cantilevers of different lengths but uniform constitutions [25] due to the identical values of $E$, $h$ and $\sigma_1$. The widths of beams shown in Fig. 11 are fixed at 20 $\mu$m, while the lengths of beams are varied in the range from: (a) 20–80 $\mu$m, (b) 80–140 $\mu$m and (c) 140–200 $\mu$m. These beams are micromachined from the a-SiC:H film deposited at a pressure of 147 Pa at 350 °C without annealing. The length difference of two adjacent beams is 15 $\mu$m. Similar deflection orientation and curvature imply that the fabrication and structure would be uniform. It is easier to observe on longer beams due to larger deflections from the surface. The downward bending is due to the superposition of uniform compressive stress and negative gradient stress. Fig. 12(a)–(c) show the cantilevers with widths of 40 $\mu$m and length of 140–200 $\mu$m in the as-deposited states at a deposition pressure of 147 Pa at the substrate temperatures of 250, 300 and 350 °C, respectively, and their corresponding global residual stresses are $-160$, $-232$ and $-272$ MPa, respectively. The length difference of two adjacent cantilevers is 20 $\mu$m. The negative curvature of beams

![Fig. 11. Deformation of 20-μm-wide cantilevers fabricated from the same wafer with various lengths of beams: (a) 20 μm to 80 μm, (b) 80 μm to 140 μm, and (c) 140 μm to 200 μm. The downward bending is due to the superposition of uniform compressive stress and negative gradient stress.](image1)

![Fig. 12. Deflection of 40-μm-wide cantilevers with various global stresses of (a) $-160$, (b) $-232$ and (c) $-272$ MPa due to the difference in the substrate temperatures of (a) 250 °C, (b) 300 °C and (c) 350 °C, respectively. The length of cantilevers ranges from 140 μm to 200 μm. The deposition pressure is fixed at 147 Pa. The negative curvature of beams indicates the gradient residual stress is negative. The larger compressive stress at higher temperature of 350 °C leads to the larger curvature or gradient stress.](image2)
indicates the gradient residual stress is negative. In addition, the curvature or the gradient stress seems to be dependent on the deposition temperature. The larger compressive stress at higher temperature of 350 °C leads to the larger curvature or gradient stress. When the deposition temperature is reduced to 250 °C, appreciable deflection is not observed. It implies the gradient stress is small. Fig. 13(a)–(c) show the cantilevers identical to Fig. 12(a)–(c), respectively; however, they are fabricated with an additional post annealing at a temperature of 420 °C for 30 s. The global stresses are changed into 12, −40 and −50 MPa in films responsible for Fig. 13(a)–(c), respectively, and all exhibit positive curvature. The phenomena not only imply that the gradient stresses are positive, but also suggest that the annealing effect exhibits a more positive variation on curvature. It is related to the annealing induced more tensile stress as well as positive gradient stress in film. The trend of global stress variation is similar to that shown in Fig. 4 and a little larger than that in Fig. 7. The gradient stress varied from negative to positive after post annealing is related to the difference of stress variation between the free surface and the interface of the film and substrate. The stress variation at a free surface without constraints is much smaller than that at the interface with a substrate constraint. As the global stress becomes more tensile after annealing, most stress variation occurring at the interface leads to more positive gradient stress. The flat released structure is related to the zero curvature or gradient stress. So, the flat micro-machined suspended structure should consider the low residual stress with small gradient component. If low residual stress has high gradient component, it will lead to the beam bending as Fig. 13(a)–(c). A good process for suspended MEMS structure fabrication is to adjust the low residual stress with a near-zero gradient component by as-deposited process parameters and post annealing.

4. Conclusions

The residual stress of as-deposited PECVD silicon carbide films is affected by total deposition pressures of 80–213 Pa and the substrate temperatures of 250–350 °C, respectively. The lowest stress level of −160 MPa could be obtained at a medium pressure of 147 Pa and at a low substrate temperature of 250 °C. The stress can be further reduced and adjusted by a post rapid thermal annealing. The stress is shifted from the compressive region to the tensile region as the annealing temperature increases. The transition temperature ranged from 400 °C or 500 °C for the variations of the residual stress from the compressive to the tensile stress depending on the deposition substrate temperatures. The higher the deposition temperatures, the higher the annealing temperatures required to relax the compressive stress. The stress relaxation is attributed to the dissociation of hydrogenated bonds and incorporation of hydrogen molecules in the film for the formation of tensile Si–C bonds with large electronegativity difference or the out-diffusion of hydrogen molecules to increase the bond length between atoms for an additional tensile stress. The difference in electronegativities of Si and C atoms is much larger than that in Si and H atoms or in C and H atoms, so the interaction between Si and C atoms is enhanced and the atoms in films are prone to contract, leading to a tensile stress to compensate the compressive stress. In addition, the outdiffusion of hydrogen molecules results in the increasing bond length between atoms and molecules and therefore generates additional tensile stress to compensate the compressive stress.
The residual stress is composed of the uniform stress and the gradient stress. The gradient stress could be displayed from the curvature of cantilevers, depending on the deposition temperature. The larger compressive stress at higher temperature of 350 °C leads to the larger negative curvature or gradient stress. When the deposition temperature is reduced to 250 °C, appreciable deflection is not observed. It implies that the gradient stress is small. In the as-deposited state, the gradient stress is negative and then transformed to positive after annealing at a temperature of 420 °C for 30 s. The difference of stress variation between the free surface and the interface of the film and substrate results in the gradient stress varied from negative to positive. Most stress variation occurring at the interface leads to more positive gradient stress in films while more tensile stress is obtained after annealing. It will be a good process for the flat suspended MEMS structure fabrication by as-deposited process parameters control and post annealing to get a low residual stress with a near-zero gradient component.

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