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Enhancement of stiffness of vertically standing Si nanosprings by energetic ions

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In this study, the compressive strength of individual Si nanosprings grown by glancing angle deposition was determined by atomic force microscope based force-distance spectroscopy. Different irradiation conditions were employed to separately investigate the role of inelastic electronic excitations and elastic nuclear stopping on the stiffness of Si nanosprings. This was achieved by using different ion beams at various energies such that the energy deposited per unit volume (ε) ranged from less than 1 eV nm−3 to ∼MeV nm−3. For moderate values of energy densities, a logarithmic dependence of the stiffness change on ε was observed. Interestingly, stiffness of the nanosprings increased up to 32% without any visible deformation and is explained in terms of densification of structures caused via electronic excitations induced by energetic ions. However, for very high energy densities (∼MeV nm−3), an anisotropic deformation of nanosprings initiated by nuclear energy losses occurred predominantly and the stiffness of the nanosprings was observed to increase up to 170%. The present study demonstrates the role of electronic excitations and nuclear stopping in enhancing stiffness of Si nanosprings, and indicates that higher surface-to-volume ratio and shape anisotropy of nanostructures aid in delaying the process of anisotropic deformation during irradiation. © 2010 American Institute of Physics. [doi:10.1063/1.3382913]

I. INTRODUCTION

Recently, the interest in the properties of nanostructures such as nanorods and nanosprings has increased due to their promising application in the field of microelectromechanical/nanoelectromechanical systems.1,2 These fundamental mechanical components can function as resonators, actuators, or cantilevers. Nanosprings in particular have been shown to work as nanoelectromechanical actuators and pressure sensors where their mechanical properties are of prime concern.3,4 In addition, microstructures having helical shapes have applications in the field of optics as circular birefringence and three-dimensional (3D) photonic crystals.5–7 For such applications, it is required to synthesize new materials with desirable characteristics or the existing materials can be tailored to introduce desirable characteristics.

When an energetic ion penetrates a solid, it undergoes collisions with the electrons and atoms of the target material. These collisions cause the incident ion to lose its energy either via inelastic collisions or via elastic collisions. The former process is known as electronic energy loss, ɛe, and is dominant in the million electron volt energy range of ions. The latter process, known as the nuclear energy loss ɛn, is dominant in the kilo electron volt energy range. To understand the ion-solid interaction, a target is considered to comprise of an electronic and an atomic subsystem. Energetic ions first excite and ionize the electrons in time scale of less than 10−16 s. In this case, the stopping cross-section of ions is smaller. Due to the large mass difference between the ion and the electrons, the ion track suffers almost no deviation from its straight line path after collision from electrons. Therefore, electronic collisions involve much smaller energy losses per collision, negligible deflection of the ion trajectory, and negligible lattice disorder giving rise to a cylindrical spike shape.8 The local thermalization in the electronic subsystem is completed in about 10−14 s after which the energy is transferred to the atomic subsystem. As the incident ion slows down, it possesses lower kinetic energy and suffers head-on collisions with the target atoms, undergoes large Rutherford scattering, and gives rise to an increased scattering cross-section.9 The nuclear collisions thus involve large discrete energy losses and significant angular scattering. This forward scattering by the incident particles when averaged over all angles in 3D gives rise to a spherical cross-section. Collectively, during both the processes the energetic ions deposit some amount of energy per unit volume (ε) in the target, which can be estimated as (ɛe+ɛn)Φ, where Φ is the ion fluence. Recently, we have reported a decrease in the stiffness of slanted Si nanorods by 100 MeV Ag5+ ions due to bending of nanorods and the transformation of pristine nanocrystalline rods to completely amorphous Si (a-Si) nanorods.10 An increase in mechanical stiffness of nanosprings by 1.2 MeV Ar8+ ion irradiation due to the anisotropic deformation of the nanosprings has also been studied by us.11 In this paper, we report the effect of ion irradiation on the compressive strength of Si nanosprings grown by glancing angle deposition (GLAD) method. The role of electronic excitations and nuclear stopping are investigated sepa-
rately. By choosing two ion beams of different energies the $\varepsilon$ values were varied from $1.2 \times 10^2$ to $1.3 \times 10^5$ eV nm$^{-2}$. The stiffness of the nanosprings was increased by as much as 32% post irradiation without deformation and by 170% after their anisotropic deformation. The study indicates that nuclear energy losses and surface-to-volume ratio of nanostructures are parameters of prime concern for affecting anisotropic deformation. The dependence of stiffness of nanosprings on $\varepsilon$ is also discussed.

II. EXPERIMENTAL

A. Synthesis, irradiation, and characterization of Si nanosprings

GLAD is a physical vapor deposition technique where the size, shape, and geometry of the nanostructures can be modified by an accurate control of substrate orientation and rotation. The deposition is carried out at small glancing angles so that the atoms deposited by the incoming vapor flux create atomic shadows on the substrate. In addition to substrate orientation, substrate rotation can be controlled to grow various kinds of nanostructures. Two sets of columnar films comprising of five-armed silicon nanosprings were grown by glancing angle ion beam sputter deposition on flat Si substrates. The first set of samples was grown on bare Si(100) substrate and labeled as set S-I. The other set was grown on Si(111) substrate prepatterned with 370 nm diameter silica spheres (labeled as set S-II). The samples were grown by a multiphase substrate rotation technique. In this scheme, one complete rotation of the substrate comprised of four sectors in which the rotation speed and the angle at which the flux was incident were varied. Details of growth of nanosprings have been described in an earlier paper. While S-I samples were subjected to swift heavy ion irradiation by 100 MeV Au$^{8+}$ ion beam at Inter-University Accelerator Centre (IUAC), New Delhi, India, the S-II samples were irradiated by 1.2 MeV Ar$^{8+}$ ion beam using the Low Energy Ion Beam Facility at IUAC. Both irradiation experiments were performed at liquid nitrogen (LN$_2$) temperature. Identical samples from each set were mounted on the topmost part of the LN$_2$ cooled ladder to serve as pristine samples. A high vacuum better than $10^{-6}$ mbar was maintained during irradiation. The samples of S-I set were irradiated such that the atoms deposited by the incoming vapor flux create atomic shadows on the substrate. In total, 10$^{12}$, 10$^{13}$, 5 $\times$ 10$^{13}$, and 10$^{14}$ ions cm$^{-2}$ were employed. Similarly, for S-II set the dose varied from $1.2 \times 10^{12}$ to $1.2 \times 10^4$ eV nm$^{-3}$. To achieve this variation, ion fluences of $10^{12}$, $10^{13}$, $5 \times 10^{13}$, and $10^{14}$ ions cm$^{-2}$ were employed. Similarly, S-II set was irradiated such that $\varepsilon$ varied from $1.3 \times 10^{-1}$ to $1.3 \times 10^5$ eV nm$^{-2}$ by employing ion fluences of $10^{14}$, $10^{15}$, and $10^{16}$ ions cm$^{-2}$. It may be mentioned that $10^{14}$ ions cm$^{-2}$ is the upper limit of ion fluence available for high energy ion irradiation. All the samples have been named according to their sets followed by the ion fluence values. For example, “S-II (10$^{16}$ ions cm$^{-2}$)” refers to the S-II set of nanosprings irradiated at ion fluence of $10^{16}$ ions cm$^{-2}$. The suffix “pristine” has been reserved for pristine samples. The direction of the ion beam was kept normal to the substrate plane for all irradiation experiments. All the samples were characterized by scanning electron microscope (SEM) in both cross-section and top-view with 2.5 kV acceleration voltage. A micro-Raman spectroscope (LA-BRAM HR-800 Jobin Yvon Horiba equipped with an Olympus BX-41 microscope and a charge coupled device detector) with an Ar ion laser of 488 nm wavelength was used at 5 mW power to record the Raman spectra for all the samples in a backscattering geometry. An integration time of 40 s was used to record the spectra. The samples for high-resolution transmission electron microscopy studies were prepared by scratching Si nanosprings off the Si substrate and transferring them onto the carbon coated Cu grids.

B. Force-distance spectroscopy on Si nanosprings

To determine the mechanical stiffness of Si nanosprings, atomic force microscope (AFM, Nanoscope IIIa, Veeco Instruments) based force-distance ($F$-$d$) measurements were performed. In $F$-$d$ spectroscopy, the cantilever deflection ($y$) is measured indirectly as a function of the scanner distance ($d$) in terms of a parameter known as sensitivity. To calibrate the sensitivity in an AFM, $F$-$d$ spectroscopy is first performed against a hard surface like sapphire. Since the AFM cantilever cannot penetrate this hard surface, the distance moved by the piezoscanner gets transferred to the flexible AFM cantilever and appears as the cantilever deflection. The slope of this deflection-distance plot yields the sensitivity value and is used to convert the units of cantilever deflection recorded in volts by the position-sensitive-photodetector (PSPD) to nanometer. Sensitivity of an AFM cantilever is susceptible to the position of the laser spot falling onto the cantilever, and must be accurately determined whenever a tip is mounted/adjusted in the tip holder. The force $F$ applied by the cantilever is given by $k_{CL} \times y$, where $k_{CL}$ is the force constant of the cantilever being used. The accuracy in $F$ thus depends upon the accuracy with which $k_{CL}$ can be measured. The force constant of the AFM cantilevers were calibrated against commercially available reference cantilevers (CLFC-NOBO, Veeco Instruments) by following the method proposed by Torii. In this method, $F$-$d$ data is acquired over sapphire and the slope ($\delta$) of $y$ versus $d$ plot yields $\delta_{tot}$, i.e., the total deflection of the cantilever under the load $F$. Next, sapphire was replaced by a reference cantilever of known force constant ($k_{cel}$). The slope of $y$ versus $d$ in this case yields the deflection in cantilever $\delta_{CL}$ when it is in contact with the reference cantilever. The force constant of the AFM cantilever can then be evaluated as $k_{CL} = k_{cel} \times (\delta_{tot} - \delta_{CL} / \delta_{CL})$. Therefore, the cantilever is completely calibrated. The force constant of AFM cantilevers used for $F$-$d$ measurements on S-I nanosprings was determined to be $59.63 \pm 0.46$ N m$^{-1}$ and that for S-II nanosprings was $74.95 \pm 0.51$ N m$^{-1}$. It is mentioned that the nanosprings were loaded only in the elastic regime. The onset of nonlinearity in an $F$-$d$ plot is considered as the upper limit of the load that can be applied for studying the elastic behavior of the nanosprings. This corresponds to forces less than 3000 nN for both the sample sets. The typical distance covered by the piezoscanner during the loading-unloading cycles was about 150 nm. To determine the stiffness of the nanosprings, $F$-$d$ measurements were carried out on individual nanosprings. The stiffness reported for each of the
samples in this study is an average carried over more than 20 nanosprings for each case.

III. RESULTS AND DISCUSSION

A. Effect of irradiation on dimensions of Si nanosprings

Two different sets of substrates, viz., bare Si(100) and silica nanosphere patterned Si(111) substrates were used for deposition of Si nanosprings. The silica nanospheres self-assemble in a closed hexagonal pattern when dispersed from their colloidal solution on Si substrate. Figures 1(a) and 1(b) show the cross-sectional SEM micrographs of S-I (pristine) and S-II (pristine) samples. The nonuniformity in the structures grown on bare substrates arises due to random nucleation on the substrate. On the other hand, prepatterned substrates provide artificial nucleation sites on the substrate which result in a relatively regular growth of the structures. Micro-Raman studies on S-I (pristine) and S-II (pristine) (not shown here) samples were carried out. In both the cases, a broad band at 470 cm$^{-1}$ appeared confirming the presence of a-Si.24

A broad range of $\varepsilon$ values was investigated by choosing two widely different ion beam energies. The corresponding values of $S_e$, $S_n$, and $\varepsilon$ were computed by the Monte Carlo simulation code “The Stopping and Range of Ions in Matter” (SRIM).25 For 100 MeV Au$^{+}$ ions, the $S_e$ and $S_n$ values in a-Si target are 11.72 keV nm$^{-1}$ and 0.25 keV nm$^{-1}$, respectively. The corresponding values of $S_e$ and $S_n$ losses for 1.2 MeV Ar$^{8+}$ ions in a-Si are 1.17 keV nm$^{-1}$ and 0.14 keV nm$^{-1}$, respectively. The fluence values were chosen such that the total energies deposited per unit volume by the ions in both cases are comparable. The value of $\varepsilon$ in case of 100 MeV Au$^{8+}$ ions in Si at a fluence of $10^{15}$ ions cm$^{-2}$ is $1.2 \times 10^3$ eV nm$^{-3}$. This is comparable to the value $1.3 \times 10^3$ eV nm$^{-3}$ for 1.2 MeV Ar$^{8+}$ ion irradiation at a fluence of $10^{14}$ ions cm$^{-2}$. For S-I set nanosprings, the ion fluences were varied from $10^{12}$ to $10^{14}$ ions cm$^{-2}$ (100 MeV Au$^{8+}$) and for S-II set the ion fluence was varied from $10^{14}$ to $10^{16}$ ions cm$^{-2}$ (1.2 MeV Ar$^{8+}$). The details of the irradiation parameters described above have been summarized in Table I. Figures 2(a)–(d) show the cross-sectional SEM micrographs of S-I ($10^{12}$ ions cm$^{-2}$), S-I ($10^{14}$ ions cm$^{-2}$), S-II ($10^{15}$ ions cm$^{-2}$), and S-II ($10^{16}$ ions cm$^{-2}$), respectively. The SEM micrograph of S-II ($10^{16}$ ions cm$^{-2}$) [Fig. 2(d)] depicts a visible deformation in the nanosprings. The change in the dimensions of the irradiated nanosprings was determined using cross-sectional and top-view SEM micrographs and was compared to the dimensions of the pristine nanosprings. The total length of the pristine and irradiated nanosprings is denoted by $L_0$ [see Fig. 1(a)] and $L_{\text{irr}}$, respectively. The change in the length of the nanosprings is calculated as $[(L_{\text{irr}}-L_0)/L_0] \times 100$, and has been plotted against ion fluence in Fig. 3(a) for both S-I and S-II sets of nanosprings. The data point corresponding to ion fluence of $10^{17}$ ions cm$^{-2}$ has been included from an earlier study on Si nanosprings irradiated by 1.2 MeV Ar$^{8+}$ ions.11 The change in the length of the S-I nanosprings was found to be insignificant within the measurement error. However, for low energy ion irradiation, this change is significant. The silica nanospheres used for patterning also exhibit anisotropic deformation. They shrink along the ion beam direction and expand in lateral directions. The diameter of silica spheres along the ion beam direction is observed to decrease from 370 nm to about 250 nm for an ion fluence of $10^{16}$ ions cm$^{-2}$. At the same ion fluence, the change in the length of S-II set of nanosprings reduces from about 1500 to 1200 nm. Therefore, Si nanosprings exhibit a larger fraction of the total change as compared to silica nanospheres. It is further observed that for low energy ion irradiation the deformation of individual nanospring arms, calculated as $[(L_{\text{irr}}-L_0)/L_n] \times 100$ where the subscript $n$ denotes the arm number, progressively increases from the topmost ($n=1$) to bottommost arm ($n=5$) of the vertically standing Si nanosprings. To understand the variation in the deformation of individual nanospring arms, the variation in $S_e$ and $S_n$ losses with depth is considered for both cases of irradiation by taking into account the porosity of the samples. It may be men-

<table>
<thead>
<tr>
<th>Nanospring set</th>
<th>Ion energy (MeV)</th>
<th>Range of ion fluence (ions cm$^{-2}$)</th>
<th>$\varepsilon$ (eV nm$^{-3}$)</th>
</tr>
</thead>
<tbody>
<tr>
<td>S-I</td>
<td>100 Au$^{8+}$</td>
<td>$10^{12}$–$10^{14}$</td>
<td>$1.2 \times 10^{2}$–$1.2 \times 10^{4}$</td>
</tr>
<tr>
<td>S-II</td>
<td>1.2 Ar$^{8+}$</td>
<td>$10^{14}$–$10^{16}$</td>
<td>$1.3 \times 10^{2}$–$1.3 \times 10^{4}$</td>
</tr>
</tbody>
</table>

FIG. 2. Cross-sectional SEM micrographs showing (a) S-I ($10^{12}$ ions cm$^{-2}$), (b) S-I ($10^{14}$ ions cm$^{-2}$), (c) S-II ($10^{15}$ ions cm$^{-2}$), and (d) S-II ($10^{16}$ ions cm$^{-2}$) nanosprings. The scale bar measures 500 nm. No deformation is evident in S-I nanosprings post irradiation, while S-II nanosprings deform.
tioned that it is the volume of the nanosprings that serves as the medium in which the energetic ions lose their energy. The shape of nanospring arms or their direction does not control the direction of motion of the energetic ions nor their energy loss process. The nanosprings have about 40% void space between neighboring nanosprings. The energetic ions do not lose energy while passing through these void spaces. To account for the voids, SRIM calculations were performed on a simulated target comprising of five alternate layers of Si and N2 gas of varying thickness placed on top of sixth Si layer which modeled the Si substrate. This simulated target is referred to as “layered Si target” in the following discussion, and the values of range and energy losses (both electronic and nuclear) obtained on it were used for further calculations. For 100 MeV Au8+ ions in layered Si target, the S_e losses reduce from a value of 11.7 keV nm⁻¹ at the surface to 10.4 keV nm⁻¹ at a depth of 1.5 μm. The S_n losses, however, increase from 0.09 keV nm⁻¹ on the surface to 0.1 keV nm⁻¹ at a depth of 1.5 μm. On the other hand, the electronic energy loss for 1.2 MeV Ar8+ ions decreases continuously from 1.17 to 0.9 keV nm⁻¹, while the nuclear loss increases from 0.07 to 0.22 keV nm⁻¹. Two important observations are made, (i) a significant increase in the nuclear loss with depth occurs for low energy ion irradiation and (ii) the electronic loss for high energy irradiation is ten times higher than that during low energy irradiation. Further, for comparable values of ε (~10 keV nm⁻³), the deformation of nanosprings occurs for low energy ion irradiation but is negligible for high energy irradiation. Therefore, the observed deformation cannot be attributed to S_e loss. On the other hand, the effect of nuclear loss is only evident near the end-of-range regime of incident ions. It may be noted that the range of ions in case of high energy irradiation was determined to be ~17.1 μm, while that for low energy irradiation is ~1.9 μm, which are widely different. However, the nuclear loss occurring only within the nanospring volume is of interest. As described above, for low energy ion irradiation the nuclear loss as well as the deformation of nanosprings increases with increase in depth. This deformation of individual arms has been plotted against the nuclear loss in Fig. 3(b). It is observed that the anisotropic deformation of arms increases progressively suggesting that nuclear energy loss plays a crucial role in deforming them. This important observation has been overlooked in the past by the authors reporting the anisotropic deformation by ion irradiation due to its restriction in the electronic energy loss regime.

B. Ion beam induced anisotropic deformation

The phenomenon of anisotropic deformation of materials under ion irradiation is best explained by the viscoelastic thermal spike model. According to this model, an energetic ion passing through a solid transfers its energy to the electrons in time less than 10⁻¹⁶ s and to atomic subsystem in about 10⁻¹⁴ s. The relaxation or transfer of excess energy of the excited electrons to the atoms of the target may lead to an increase in the local temperature of the material. Poor electronic mobility of amorphous materials makes it difficult for the systems to lose energy to the surroundings and may allow the energy deposited in the electronic subsystem to be confined long enough to form a transiently heated region. A cylindrical region around the ion trajectory is formed in which the stresses can relax and in this sense it is regarded as a fluid. According to the viscoelastic model, the stress-relaxation in the materials can take place after the temperature of the spike region crosses a particular “flow temperature.” This flow temperature is considered to be the melting temperature for crystalline materials. With time as the viscous track cools, shear stresses are generated. Now, depending upon the nature of the spike i.e., spherical (due to nuclear losses) or cylindrical (due to electronic losses), the process of stress relaxation within the material would be different. In case of spherical spikes, the stresses are hydrostatic in nature giving rise to isotropic stress relaxation in all directions. Therefore, spherical spikes are not usually considered to result in anisotropic deformation of materials. In case of cylindrical spikes, the stress relaxations along the radial and axial directions are different. The stresses along the longer axial direction relax leading to the anisotropic flow of the material laterally. The conservation of volume of the spike material results in an overall contraction along the ion beam direction. These macroscopically visible shape changes are characterized by the rate-of-strain tensor. The tensor includes three terms, which carry contributions from (1) the ion hammering term that quantifies the deformation tensor.
(defined as strain increment per unit ion fluence), (2) Newtonian flow that describes the effective fluidity of the molten material under ion bombardment, and (3) the Hooke’s law that accounts for the change in deformation when the stresses change in time. van Dillen et al.\textsuperscript{32} have studied the effect of 4 MeV Xe ions on spherical silica colloids of radii ranging from 19 to 510 nm. The authors observed that for higher fluences (greater than $2 \times 10^{14}$ ions cm$^{-2}$) the smaller colloids showed less deformation in comparison to larger colloids and that the macroscopic stresses arising from surface curvature of deformed colloids resisted any further deformation by ion bombardment.

C. Role of surface-to-volume ratio of nanostructures on anisotropic deformation

We assume that for nanostructures with high-aspect ratio or those with initial shape anisotropy (like nanosprings in the present case), the contribution from surface stress cannot be neglected. Consequently, the surface-to-volume ratio ($S/V$) of the nanostructures can be considered as a parameter to predict whether the contribution from surface stress would be significant or not. In view of this, $S/V$ ratio of nanosprings studied in the present work was calculated. It was found that for S-I nanosprings the $S/V$ ratio was $\sim 0.016$ nm$^{-1}$, while that of S-II was $\sim 0.004$ nm$^{-1}$. Therefore, the surface stress in both the sets needs to be considered. It would have greater effect on the nanosprings of S-I as compared to S-II set. Following van Dillen et al.\textsuperscript{32} we have modeled individual arms of S-I (pristine) nanosprings as oblates and estimated the surface stresses arising due to the shape anisotropy of nanospring arms. It was found that the surface stresses differ by an order of magnitude along the parallel and perpendicular directions of the ion beam due to nonhydrostatic stress state. In addition, a higher $S/V$ value would oppose the free anisotropic deformation of S-I set of nanosprings as compared to S-II set of nanosprings. Therefore, although the energy delivered per unit volume by 100 MeV Au ions at $10^{14}$ ions cm$^{-2}$ is comparable to that of 1.2 MeV Ar ions at $10^{15}$ ions cm$^{-2}$, the lower $S/V$ ratio and higher nuclear energy losses are favorable conditions for observing anisotropic deformation of S-II nanosprings. For all irradiated samples a broad band at 470 cm$^{-1}$ was observed in the micro-Raman spectra confirming the presence of a-Si (not shown here). Therefore, it is concluded that no structural transformation takes place after ion irradiation.

D. Effect of ion irradiation on the mechanical stiffness of Si nanosprings

As mentioned before, the compressive strength of the nanosprings before and after ion irradiation was determined by AFM-based force-distance spectroscopy. Figures 4(a) and 4(b) show the experimentally obtained $F$-$d$ curves for S-I (pristine), S-I ($10^{14}$ ions cm$^{-2}$) and S-II (pristine), S-II ($10^{16}$ ions cm$^{-2}$) sets of nanosprings. The analysis of the $F$-$d$ curves was carried out by considering the cantilever-nanospring system as a combination of two springs linearly connected in series. When the loading cycle commences, restoring forces appear in the cantilever and the nanospring.
and S-II sets. The percentage change in the stiffness of the nanosprings \((\bar{k}_{\text{irr}} - k_0) / k_0 \times 100\), where \(k_0\) and \(k \) are the average stiffness of irradiated and pristine nanosprings, respectively has been plotted against \(e\). It is observed that the average stiffness of the springs increases by as much as 15% and 32% for energy densities of \(1.2 \times 10^2\) eV nm\(^{-3}\) and \(1.2 \times 10^4\) eV nm\(^{-3}\), respectively. Similarly, in case of low energy irradiation, the stiffness is observed to increase by 25% and 32% respectively for energy densities of \(1.3 \times 10^2\) and \(1.3 \times 10^4\) eV nm\(^{-3}\). The variation in percentage change in stiffness is logarithmic up to \(\sim 10^4\) eV nm\(^{-3}\), beyond which this dependence changes. It is observed that for higher energy densities (>\(10^4\) eV nm\(^{-3}\)), the nanospring arms deform. When nanosprings deform, their length reduces while their lateral dimensions expand giving rise to an increased stiffness of the nanosprings. An increase of about 170% in stiffness of nanosprings has been observed due to this anisotropic deformation. Table II summarizes the dependence of percentage change in stiffness with \(e\). Such changes in physical dimensions of nanosprings were not visible in case of high energy ion irradiation. It is possible that the observed increase in the stiffness of nanosprings irradiated by 100 MeV Au\(^{8+}\) ions is due to the smoothening of nanocolumns present in the individual as-deposited springs into a single column. This is shown in Fig. 6 where the transmission electron micrographs reveal that the granular structures present in the as-deposited nanosprings have completely smoothened out after irradiation at \(10^{14}\) ions cm\(^{-2}\). The smoothening is also observed in case of low energy irradiated nanosprings as shown for S-II (pristine) and S-II \((10^{16}\) ions cm\(^{-2}\)) samples. This increased densification of nanosprings results in an enhanced stiffness of nanosprings. Therefore, the mechanisms, viz., electronic excitations resulting in densification of nanostructures and nuclear losses resulting in their deformation lead to enhanced stiffness of nanosprings. The energy deposited per unit volume is a common parameter that directly influences the stiffness in both cases of ion irradiation. The study indicates that higher surface-to-volume ratio of nanostructures and shape anisotropy can aid in delaying the deformation of nanostructures to some extent.

### IV. SUMMARY AND CONCLUSIONS

Variation in the mechanical stiffness of vertically standing Si nanosprings by ion irradiation has been studied in detail. The stiffness of nanosprings is found to increase logarithmically as a function of energy deposited per unit volume by the energetic ions up to \(\sim 10^4\) eV nm\(^{-3}\). An increase in stiffness of about 30% was observed for energy density value of \(10^4\) eV nm\(^{-3}\) in case of both 100 MeV Au\(^{8+}\) ion irradiation at ion fluence of \(10^{14}\) ions cm\(^{-2}\) and 1.2 MeV Ar\(^{8+}\) ions at ion fluence of \(10^{15}\) ions cm\(^{-2}\). For very high values of energy densities (>\(10^4\) eV nm\(^{-3}\)) anisotropic deformation of nanosprings was observed. The absence of deformation for high energy ion irradiation suggests that nuclear stopping plays a crucial role in the anisotropic deformation. The increase in the stiffness of nanosprings is possibly due to the bundling of nanocolumns present in the as-deposited films. The shape anisotropy and higher surface-to-volume ratio of these nanospring structures possibly counter the ion hammering effect and delay the deformation process for particular ion fluence. We propose that nuclear energy loss and surface-to-volume ratio of the nanostructures are important factors that determine the anisotropic deformation during ion irradiation.

### ACKNOWLEDGMENTS

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**Table II.** The dependence of percentage change in the stiffness of Si nanosprings on \(e\) is shown for both S-I and S-II sets of nanosprings.

<table>
<thead>
<tr>
<th>(e) (eV nm(^{-3}))</th>
<th>Percentage change in nanospring stiffness</th>
<th>S-I</th>
<th>S-II</th>
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<tbody>
<tr>
<td>(10^2)</td>
<td></td>
<td>4</td>
<td>...</td>
</tr>
<tr>
<td>(10^3)</td>
<td></td>
<td>15</td>
<td>25</td>
</tr>
<tr>
<td>(10^4)</td>
<td></td>
<td>32</td>
<td>31</td>
</tr>
<tr>
<td>(10^5)</td>
<td></td>
<td>...</td>
<td>171</td>
</tr>
</tbody>
</table>

**Fig. 6.** Transmission electron micrographs corresponding to (a) S-I (pristine), (b) S-I \((10^{14}\) ions cm\(^{-2}\)), (c) S-II (pristine), and (d) S-II \((10^{16}\) ions cm\(^{-2}\)) nanosprings. Granular structures in the as-deposited nanosprings were not visible in case of high energy ion irradiation. The scale bars correspond to 100 nm.