Morphology modification of Si nanopillars under ion irradiation at elevated temperatures: plastic deformation and controlled thinning to 10nm

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Abstract
Si nanopillars of less than 50 nm diameter have been irradiated in a helium ion microscope with a focused Ne$^+$ beam. The morphological changes due to ion beam irradiation at room temperature and elevated temperatures have been studied with the transmission electron microscope. We found that the shape changes of the nanopillars depend on irradiation-induced amorphization and thermally driven dynamic annealing. While at room temperature, the nanopillars evolve to a conical shape due to ion-induced plastic deformation and viscous flow of amorphized Si, simultaneous dynamic annealing during the irradiation at elevated temperatures prevents amorphization which is necessary for the viscous flow. Above the critical temperature of ion-induced amorphization, a steady decrease of the diameter was observed as a result of the dominating forward sputtering process through the nanopillar sidewalls. Under these conditions the nanopillars can be thinned down to a diameter of $\sim$10 nm in a well-controlled manner. A deeper understanding of the pillar thinning process has been achieved by a comparison of experimental results with 3D computer simulations based on the binary collision approximation.

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(Some figures may appear in colour only in the online journal)

1. Introduction

While the development of new device architectures and computing strategies should be disruptive to enable a rapid progress in the field, the underlying fabrication technology ideally builds upon existing knowledge to enable a quick and seamless transition to the new data processing regime. Examples for this approach towards the development of future computing units are new low power devices such as single electron transistors (SETs) or Si-based quantum computing strategies [1, 2]. The existing complementary metal oxide semiconductor (CMOS) technology can handle today’s demands on feature size and shape, but new architectures
require new—ideally CMOS-compatible—fabrication processes to meet the upcoming challenges. Here, we present a CMOS-compatible ion beam based fabrication process for sub 15 nm pillars with the potential for gate-all-around (GAA) SET device architectures [3] and other similar technologies that require three dimensional building blocks with lateral length scales in the few-nm regime. We investigate the achievable diameter reduction for sub-50 nm pillars with a height of 70 nm and also propose a model for the ion beam based mechanism.

The interactions between ion beams and Si-based material systems have been extensively studied in the last decades due to the ubiquitous and manifold applications of ion beams for device fabrication. Examples include the doping of transistors active regions [4, 5] mostly using broad ion beams and milling of nanostructures [6, 7] using focused ion beams (FIBs). Side effects of ion beam processing of Si include defect accumulation and amorphization. Studies—mostly with unstructured bulk materials—to understand the mechanism and to avoid these effects have been carried out with various ion species, energy ranges, target structures etc [8] and are comprehensively summarized in review papers [9, 10] and textbooks [11, 12].

Ion irradiation at elevated temperatures has been considered as a promising technique to mitigate ion beam induced damages and amorphization of the substrate during the irradiation processes [13–15]. For most semiconductors, a finite temperature $T_\text{f}$—lower than the temperature for epitaxial recrystallization [16, 17]—exists at which amorphization is prevented during irradiation. This is related to out-diffusion of vacancies from the ion track region, and is described in the so called out-diffusion theory [18, 19]. According to this theory, at any finite temperature, a thermally driven interstitial-vacancy recombination, competes with the amorphization caused by the incident ion. When the substrate temperature $T \ll T_\text{f}$, the ion damage prevails and Si undergoes continuous amorphization. Once the temperature exceeds $T_\text{f}$ the dynamic annealing will recover the amorphized pocket of each incident ion, thus preventing the amorphization. The process is governed by ion mass, target temperature and—to a lesser extent—flux and ion energy. Here, we utilize this dynamic annealing process at slightly elevated temperatures to prevent the amorphization of nanostructures during ion beam irradiation.

In the case of the nanostructures used in this work the length scale of the ion collision cascade becomes comparable to the structure size which results in unanticipated effects. Such effects can arise from changes in the distribution of the deposited energy. These changes are a result of the truncation of the collision cascade by the nanostructure. This reduces the amount of energy deposited into the nanostructure and changes the distribution of the deposited energy as the final low particle energy part of the cascade is missing. However, this part is characterized by a high relative nuclear energy loss as compared to the part of the cascade closer to the impact point. In addition, new processes such as backside or forward sputtering can change the stability of the nanostructure during ion beam irradiation [20].

In this work we employ FIB and broad-beam irradiation in the few tens of keV range to shrink the diameter of few-tens of nm pillars down to nearly 10 nm. Specifically we use Ne$^+$ irradiation at 25 keV under normal incident in a helium ion microscope (HIM) [21, 22]. In the HIM the sample is heated in situ using a home-built heater stage that can be loaded through the load lock of the Carl Zeiss Orion NanoFAB. Although the HIM has a lateral resolution of less than 2 nm when using Ne$^+$, we scan the beam over a set of nanostructures to emulate a broad beam irradiation, to which we also compare the results at the end of the manuscript. The benefit of this approach is that a few pillars from the same sample chip can be irradiated at different fluxes and/or at different temperatures. Possible morphological or structural changes can than directly be compared in the subsequent HIM imaging step at RT. This characterization has been performed using high resolution HIM and transmission electron microscopy (TEM) to obtain information on the morphology and crystallinity of the obtained nanopillars, respectively.

The recently developed Monte Carlo simulation program TRIDYN [23] is used to perform a fully dynamic 3D simulation of the irradiation process. This is complemented by sputter yields and distributions of sputtered particles extracted from the static collision simulation program TRIDST [20, 24]. The simulated results are compared with experimental findings and help with the understanding of the underlying processes.

To demonstrate the possibility for upscaling of this method we also employ Si$^{++}$ broad beam irradiation. Here, always the entire sample is irradiated and multiple samples have to be used to investigate the influence of temperature and ion flux. Both local and broad beam based irradiation suggest a versatile and CMOS-compatible fabrication method of sub-10 nm diameter vertical Si nanostructures at slightly elevated but still CMOS-compatible temperatures.

2. Methods

Silicon nanopillars have been fabricated via patterning of Si-rich anti-reflective coating (SiARC) and spin-on carbon (SOC) hard mask with an electron beam direct write (EBDW) system (SB3054, VISTEC) and subsequent plasma dry etching (Centum®, Applied Materials) in the 200 nm production line at CEA-Leti. The half-height diameters of nanopillars range from 25 nm to 50 nm and the pitch is larger than 250 nm thus redeposition of sputtered atoms is prevented. The size of a nanopillar array is $5 \mu m \times 5 \mu m$.

The nanopillars are irradiated with a 25 keV focused Ne$^+$ ion beam from the HIM. Scanning the beam over a small area of $5 \mu m \times 2 \mu m$ with a sufficiently small pixel spacing and dwell time emulates the conditions in a broad-beam implanter. The beam current is restricted under 500 fA using a 20 $\mu m$ Au aperture. Under this condition the time interval between two arriving Ne ions is longer than 320 ns compared to the

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3 For Si this is in the range of 550 °C to 600 °C.
irradiated and HT-irradiated nanopillars are shown in addition, broad beam irradiation has been carried out using surface normal with a probe size smaller than 0.5 nm. In with 25 keV Ne\(^+\) (room temperature) Selected silicon nanopillars are irradiated from the top at both the HIM a home-built heater stage with a tungsten filament capable of reaching \(\sim 500^\circ\)C is employed. The temperatures of the sample and the stage are monitored with type K thermocouples. Unless noted otherwise the most important remaining HIM Ne\(^+\) irradiation conditions are: 1 nm pixel spacing and 0.1 \(\mu\)s dwell time. Images of the nanopillars are taken using a 25 keV He\(^+\) beam at 85° tilt angle to sample surface normal with a probe size smaller than 0.5 nm. In addition, broad beam irradiation has been carried out using 50 keV Si\(^+\) from a 200 keV ion implanter (Danysik Model 1090) at the Ion Beam Center (IBC) of the Helmholtz-Zentrum Dresden-Rossendorf (HZDR)

Cross-sectional samples to analyse the lateral shape of selected Si nanopillars were obtained by FIB (Zeiss NVision 40) milling and lift-out of 1 \(\mu\)m \(\times\) 50 \(\mu\)m TEM lamellae (about 50 nm in thickness). Images have been acquired with an FEI Titan 80–300 microscope using bright-field transmission electron microscopy (BF-TEM) with a contrast-enhancing objective aperture and energy-filtered transmission electron microscopy (EFTEM) using the plasmon-loss peak at 17 eV to show the contrast of Si abundance. These data allow to analyze the crystallinity of the nanopillars and the dimensions can be measured even when the pillar is amorphized. Nanopillar diameter and height, as well as the volume reconstruction for the sputter yield calculation were achieved with the help of software Fiji [25].

Monte-Carlo (MC) codes based on the binary collision approximation (BCA) have been employed to simulate the ballistic processes during ion irradiation in order to contribute to the interpretation of the experimentally observed phenomena. The static program TRi3DST [20, 26] models the irradiation of three-dimensional (3D) bodies whose surface can be described by analytical functionals. TRi3DYN [23, 24] is employed for fully 3D dynamic simulations of the modification of an irradiated structure during bombardment, with arbitrary bodies being set up in a 3D voxel grid. Both codes assume amorphous materials and do not include collective phenomena such as the viscous flow. Details of the simulations are given in the Supplementary Information is available at stacks.iop.org/SST/35/015021/mmmedia.

3. Results and discussions

3.1. Amorphization of Si nanopillars

Selected silicon nanopillars are irradiated from the top at both room temperature (RT) and high temperature (HT) (400 °C) with 25 keV Ne\(^+\) using a fluence of \(2 \times 10^{16}\) cm\(^{-2}\). HIM images recorded using an 85°-tilt of the non-irradiated, RT-irradiated and HT-irradiated nanopillars are shown in figures 1(a)–(c), respectively. After the irradiation at RT, a strong change in shape of the nanopillars—characterized by a rounding of the upper edge accompanied by a severe loss of the height—can be observed. However, no such shape change is seen in the case of HT irradiation of the nanopillars.

In figure 1(d) identical nanopillars are irradiated with a combined RT and HT irradiation but in a different sequence. The resulting morphology can be compared in a qualitative way in the image. Both pillar fields have received the same total fluence but in reverse orders. The pillar field in the back has first been irradiated at RT with a fluence of only \(2 \times 10^{15}\) cm\(^{-2}\) followed by a fluence of \(18 \times 10^{15}\) cm\(^{-2}\) at an elevated temperature of 400 °C. The fluence of \(2 \times 10^{15}\) cm\(^{-2}\) for the first irradiation step has been chosen so that it leads to an amorphization of the pillars. To ensure a complete amorphization of the pillars the chosen fluence is about 5 times higher than what has been reported for the amorphization of silicon by few 10 keV Ne\(^+\) irradiation [19]. However, the fluence is small enough to not lead to observable changes in the morphology of the pillars. The pillars in the foreground of figure 1(d) were first irradiated at 400 °C with a fluence of \(18 \times 10^{15}\) cm\(^{-2}\) followed by \(2 \times 10^{15}\) cm\(^{-2}\) after being cooled down to RT. As a result both pillar fields have received a total fluence of \(2 \times 10^{16}\) cm\(^{-2}\). However, as is clear from figure 1(d) the sequence of RT versus HT irradiation plays an important role for the final morphology. The amorphous pillars in the background—irradiated first at RT—are nearly completely removed. A closer look also reveals that the diameter at the foot of the former pillars has increased. The pillars in the foreground—irradiated first at HT—still show their pristine shape with nearly no observable change in

![Figure 1. Si nanopillars with an original diameter of 25 nm have been irradiated with 25 keV Ne\(^+\) and subsequently imaged at 85° with He\(^+\) from a HIM. Image (a) shows the nanopillars before irradiation while (b) and (c) show the nanopillars after 2 \(\times\) 10\(^{16}\) cm\(^{-2}\) Ne\(^+\) irradiation at RT and 400 °C, respectively. Image (d) shows the direct comparison of nanopillars after having received 2 \(\times\) 10\(^{15}\) cm\(^{-2}\) at room-temperature and 1.8 \(\times\) 10\(^{16}\) cm\(^{-2}\) at 400 °C but in different order. Scale bars in all images indicate 100 nm.](image-url)
displacements created by 25 keV Ne\(^+\) nanopillars. The black dashed lines indicate the outlines of the actual Si amorphized, only top segment amorphized and entirely crystalline showing the cases of the nanopillar being almost entirely amorphous regions after an irradiation at temperatures lower than \(T_c\). The pro-

With the MC simulation program TRI3DYN, the dynamic evolution of a Si nanopillar under ion irradiation can be visualized. In figure 3(d) the accumulated displacements per atom (dpa) of the central slice of a nanopillar after 25 keV Ne\(^+\) irradiation with a fluence of \(2 \times 10^{16} \text{ cm}^{-2}\) is shown. In this particular case, the comparison between the simulation and experiments is a simplification from the complex interplay leading to amorphization, including the original dpa created by the ion trajectory, temperature-dependent dynamic annealing, overlapping defect pockets as well as deformation due to viscous flow. BCA type MC simulation programs like TRI3DYN usually only treat amorphous samples, and effects such as dynamic annealing and viscous flow are not considered.

The obtained simulation results can well explain the amorphization profiles obtained by BF-TEM presented in figures 2(a), (b). In these cases, only the top of the pillar is amorphized which corresponds to the region with the highest defect density according to the simulation result presented in figure 2(d). Furthermore, from the BF-TEM micrographs in figures 2(a), (b) one can see that the interface between the amorphous and crystalline part of the Si nanopillar tends to bend towards the bottom of the pillar. This also agrees well with the simulated result which shows that in the lower part of the nanopillar the displacement density is higher close to the sidewalls than in the centre. This is attributed to the slightly

morphology. The final irradiation of the foreground pillars at RT ensures they receive the same amount of amorphization as the background pillars received initially. Again, this does not lead to a noticeable change of the pillar shape.

A detailed investigation of the temperature dependence of the ion beam induced amorphization is carried out on similar structures. A series of nanopillars is irradiated with Ne\(^+\) ions using a fluence of \(2 \times 10^{16} \text{ cm}^{-2}\) at various temperatures from 250 °C up to 350 °C in 25 °C steps. In figures 2(a)–(c) selected BF-TEM micrographs of typical structures are presented (TEM images for the other temperatures can be found in the supporting information). In BF-TEM micrographs only the crystalline part of the nanopillar is visible, and it is evident that after low temperature irradiation the pillar becomes amorphized. The black dashed lines in figures 2(a)–(c) outline the border of the actual pillar as extracted from corresponding EFTEM micrographs (not shown here; please see the supporting information for examples). With increasing irradiation temperature, at the same fluence an increasing part of the pillar stays crystalline and only a small part of the pillar becomes amorphous. Finally, for an irradiation temperature of 350 °C no amorphization can be detected from the BF-TEM images. From the investigation of the amorphization behaviour at additional temperatures (see the supporting information), we conclude that the critical temperature of amorphization (\(T_c\)) during 25 keV Ne\(^+\) ion irradiation for the presented Si nanopillars lies between 325 °C and 350 °C.

Figure 2. BF-TEM micrographs of Si nanopillars irradiated with \(2 \times 10^{16} \text{ cm}^{-2}\), 25 keV Ne\(^+\) at (a) 250 °C, (b) 300 °C and (c) 350 °C showing the cases of the nanopillar being almost entirely amorphized, only top segment amorphized and entirely crystalline. The black dashed lines indicate the outlines of the actual Si amorphized, only top segment amorphized and entirely crystalline. The profile has to be compared to the extent of the crystalline and amorphous regions after an irradiation at temperatures lower than \(T_c\).
nanopillars, it can not recover the crystal structure of an initially amorphous nanopillar by epitaxial recrystallization. As a consequence, surface tension induced by ion irradiation leads to viscous behavior of amorphized nanostructures and the resulting plastic deformation visible in figure 1(b) and in the back of figure 1(d), as shown in previous studies—albeit mainly at ion energies from 100 keV to a few MeV—with various ion species, energies and target materials [27–29].

To summarize, we showed that above the critical temperature of amorphization $T_c$ dynamic annealing prevents the amorphization of the crystalline Si nanopillars. From the TEM analysis we find out that this temperature corresponds to 325 °C to 350 °C for irradiation with 25 keV Ne$^+$. This is in agreement with the results from HIM imaging where we find severe deformation of initially crystalline nanopillars only after irradiation at temperatures below $T_c$. This deformation is a result of viscous flow of the amorphous material due to capillary forces acting on the nanoscale pillars. Initially amorphous pillars experience dramatic shape changes even if irradiated at HT, as the temperature for epitaxial recrystallization is still 200 °C higher and dynamic annealing is not sufficient to recrystallize the nanopillars.

### 3.2. Thinning of the nanopillars

Si nanopillars were irradiated at a temperature well above $T_c$ to fully exclude the influence of ion-induced amorphization, and the applied fluence has been increased up to $8 \times 10^{16} \text{ cm}^{-2}$. The nanopillars studied here have a diameter at half-height of 47 nm and height of 71 nm. After the irradiation, the sample was immediately imaged with the HIM at a tilt angle of 85°. In figure 3(a) the height and diameter of the nanopillars with the same original diameter after different irradiation fluences are plotted. With the increase of the fluence, a steady decrease of the diameter has been observed while the height remains almost unchanged. Linear fitting of the diameter reduction shows a slope, i.e. the reduction rate of $-3.3 \pm 0.1 \text{ nm} / (1 \times 10^{16} \text{ cm}^{-2})$. In the same diagram the simulated results of the thinning process from the program TRI3DY are shown. Snapshots of the central 3 nm slice of the nanopillar after different irradiation fluences are presented in figure 3(b). While the simulation qualitatively fits with the experimental data it slightly overestimates the sputter yield for smaller diameters. Such a tendency was also observed in previous work [29] where the ion irradiation was performed at 45° incidence relative to the axis of a Si nanowire. The smaller the pillar diameter is, or when the ion impact position is close to the nanopillar rim, the higher the chance will be that a recoiled atom will leave the structure with a high kinetic energy. This reduces the energy deposited inside the pillar and subsequently leads to less energy deposited per incident ion as compared to a larger diameter pillar or a bulk system.

To further analyze the mechanism of the decrease of the nanopillar diameter, the average sputter yield during the pillar thinning process has been measured using the incident ion current and the lateral dimensions of the pillars and plotted against the diameter of the pillars. Detailed description of calculating the sputter yield from TEM micrographs is

tapered sidewall of the nanopillars (typically 7°) which leads to high angle sputtering, creating high amount of displacements close to the pillar surface. From the TRI3DY result presented in figure 2 we conclude that a fluence of $2 \times 10^{16} \text{ cm}^{-2}$ at RT results in a damage of at least 6 dpa. This is more than sufficient to amorphize the entire nanopillar in the absence of any dynamic annealing processes.

According to the defect out-diffusing model of amorphization [18], when the temperature of the crystalline silicon is higher than the amorphization critical temperature $T_c$, no matter how high the fluence is during an irradiation the originally crystalline structure will not be amorphized. From figures 1(c), (d) and 2(c) one can see that during irradiation at temperatures higher than $T_c$ the pillars stay crystalline and no morphological changes can be observed. This temperature of 325 °C to 350 °C is still 200 °C lower than the regime where epitaxial recrystallization could recover a large-area amorphous structure [16]. The nanopillars irradiated with 25 keV Ne$^+$ at RT become amorphous already after a very low fluence, similar to the amorphization fluence in bulk Si which is between $1 \times 10^{15} \text{ cm}^{-2}$ and $2 \times 10^{15} \text{ cm}^{-2}$ [19]. While the dynamic annealing by the ion beam at HT above $T_c$ is sufficient to prevent amorphization of the initially crystalline
collision cascade has a high probability to overlap with the surface. When a recoiled atom has a kinetic energy higher than the surface binding energy for Si of 4.7 eV [20], it will leave the nanopillar as a forward sputtered atom. Detailed analysis from the simulation results in figure 3(b) shows that at the initial diameter approximately 74% of the sputtering can be attributed to the forward sputtering from the ions penetrating into the top of the nanopillar.

As the irradiation proceeds, the two contributions evolve in different manners. The high-angle sputtering on the sidewalls of the nanopillars slightly decreases with the decreasing diameter and unchanged sidewall tapering (see figure 3(b) and the insets in figure 3(a)). On the other hand, forward sputtering from ions hitting the top of the pillar will be more severe due to a larger surface to volume ratio.

Results from simulations, using TRIM3DST, of focused-beam irradiation onto a Si nanopillar are shown in figures 4(b)–(d). The ions hit the top surface of the nanopillar at (b) 5 nm from the rim and (c), (d) in the centre. For statistical reasons, an incidence of 1000 ions was simulated in all cases. In figures 4(b)–(d), blue dots indicate the position on the nanopillar sidewall where a sputter event occurred. It is clear that for every single incident ion, the smaller the diameter of the nanopillar is, or the closer the incident spot is located to the rim, the higher the amount of sputtered atoms will be. This behavior results in a steady decrease of the nanopillar diameter, mostly due to the enhanced forward sputtering. In this model, two distinctive stages will be observed for such a thinning process. When the diameter of the nanopillars is significantly larger than the lateral range of the collision cascade, the incident ions that create forward sputtering near the rim, such as in the case of figure 4(b), are unlikely to sputter from the other side of the nanopillar. Incident ions closer to the centre of the nanopillar, as shown in figure 4(c), would only induce a small amount of forward sputtering as the collision cascade barely reaches the pillar surface. As a consequence, the decrease of the pillar diameter has a linear dependency on the fluence. When the diameter of the nanopillars is comparable to or smaller than the size of the collision cascade, which is 18.6 nm in the case of 25 keV Ne⁺, the chance that a recoil atom leaves the nanopillar via forward sputtering is high, independent from the ion impact position.

However, with decreasing pillar diameter, the size and shape of the collision cascade will also change. This is due to the limited size of the nanopillar, which reduces the actual average lateral range of the ions to a value smaller than the diameter but also reduces the range of the ions as they are more likely to leave the target structure. This situation is depicted in the inset of figure 4(a). The lower part of the bulk collisions cascade is missing in the collision cascade inside the nanopillar. This also result in a redistribution of the energy deposition and the related defect density. In the nanopillar more energy is deposited close to the top which leads to more sputtering at the top and eventually a reduction of the pillar height.

To demonstrate the potential for upscaling to industrial applications, we also performed Si⁺ broad beam irradiation of

Figure 5. EFTEM micrographs show the nanopillars after HT-irradiation with (a) Ne⁺ and (b) Si⁺ of 6 × 10¹⁶ cm⁻² fluence. Fast Fourier transform (FFT) shows in both cases the top segment of the nanopillar is partially amorphized while the lower part remains crystalline. (c) As measured from 85°-tilted HIM and EFTEM images, the thinning from Ne⁺ in Si nanopillars is limited when the diameter gets smaller than 16 nm while using Si⁺ for the same purpose could reduce the diameter to 11 nm.

 included in the Supporting Materials. As shown in figure 4(a), the average sputter yield is approximately constant as long as the diameter of the nanopillars is significantly larger than the lateral range of the Ne⁺ ions of ~19 nm. The sputter yield measured for the nanopillars is at least 3 times higher than what is expected for sputtering at normal incidence on bulk Si surfaces as obtained from TRIM3DST or SRIM.

Two factors are contributing to such a high sputter yield. First, due to the approximately 7° tapering, the sidewalls of the nanopillar are also exposed to the irradiation at a high glancing angle which results in a strongly enhanced sputter yield. This contribution is hard to quantify due to the additional radial curvature of the nanopillar sidewalk. The second contribution comes from is the forward sputtering from ions hitting at normal incidence on the top of the nanopillar. From the schematic presented as an inset in figure 4(a) one can see that in contrast to the case of a bulk structure (indicated as light red) where the collision cascade is fully embedded inside the sample, in the case of a nanopillar, the lateral extent of the irradiation with Ne⁺ of 8 × 10¹⁶ cm⁻² fluence.
nanopillar arrays in a standard ion implanter. To be able to compare the results to the previous Ne⁺ irradiation, we used 50 keV Si⁺, which has the similar dpa as 25 keV Ne⁺, to irradiate the nanopillars at 400 °C. In figures 5(a) and (b) EFTEM micrographs of nanopillars after a fluence of $6 \times 10^{16}$ cm⁻² with focused Ne⁺ and broad-beam Si⁺ are shown. These experimental conditions were chosen as they represent the highest fluxes applied in both irradiation conditions that do not lead to a strong decrease of the nanopillar height. Both cases indicate the experimental limit for possible lateral pillar shrinking. Fast Fourier transform (FFT) results of the upper and lower part of an irradiated nanopillar are shown as insets in figure 5(a). They reveal that while the lower part of the Si nanopillar remains single crystalline the top part is characterized by a co-existence of crystalline and amorphous areas. As we further increase the irradiation fluence, the height of the nanopillar start to decrease and the interface between the crystalline and amorphous parts moves towards the bottom. In the case of Si nanopillars irradiated with Si⁺ of $6 \times 10^{16}$ cm⁻² fluence, the majority of the nanopillar also remains crystalline despite the higher density of defects in the vicinity of the top surface, indicating a better structural integrity than the one irradiated with Ne⁺.

In figure 5(c) the average height and diameter of the nanopillars during the thinning processes via focused Ne⁺ and broad-beam Si⁺ are plotted. The dashed lines are guides to the eyes to allow a clear comparison of the trends between irradiation with the two ion species. In the case of Si⁺ irradiated nanopillars, the diameter was measured via 85°-tilted HIM imaging subtracting the thickness of native oxide on the nanopillar sidewalls. While in both cases the curves eventually turn steep as the height starts to decrease and the thinning process slows down, allowing for a 15% reduction of height, the achievable diameters for Ne⁺ and Si⁺ thinned nanopillars are 14 nm and 11 nm, respectively.

The difference in final diameter may be attributed to the following factors. First, with the projected ranges $R_p$ of 73.2 nm for 50 keV Si⁺ and 56.9 nm for 25 keV Ne⁺, the Si⁺ better fits the height of the original nanopillars. A higher energy of the incident ion would result in a deeper and shallower energy deposition thus postponing the defect accumulation at the top segment of the nanopillars. Second, Si self-irradiation leads to a more forward directed collision cascade due the optimal energy transfer between the incident Si and the target Si atoms having the same mass. As a consequence Si irradiation results in more homogeneous sputtering of the pillar and a slower reduction of the longitudinal range due cascade truncation by the pillar.

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