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How nanocavities in amorphous Si shrink under ion beam irradiation: An in situ study

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Nanocavities were formed in Si substrates by conventional H implantation and thermal annealing, after which the samples were amorphized by Si ion irradiation. The size evolution of the nanocavities was monitored in situ during further ion irradiation with Si or As at temperatures of 300 or 600 K. The decrease in nanocavity diameter during ion irradiation depended linearly on the ion fluence. The rate of shrinkage differed according to the ion beam-induced atomic displacement rate and had little or no temperature dependence. These in situ results shed new light on possible ion-beam-induced nanocavity shrinkage mechanisms. © 2002 American Institute of Physics. [DOI: 10.1063/1.1509854]

Nanocavity formation in Si may be important for technological applications [e.g., for the so-called “smart cut®” process in crystalline Si (c-Si) whereby a thin Si layer can be removed from the bulk].1 It has been shown that, like vacancy clusters, nanocavities also act as sinks for interstitials and/or vacancies generated by the ion beam, thus reducing residual disorder after annealing.2,3 The latter is of significance in Si device technology.4,5 Previously, the following results were obtained regarding the influence of nanocavities on irradiation-induced disorder: (1) The defect–nanocavity interaction depends on the sample temperature during irradiation. Specifically, at room temperature nanocavities were preferential sites for Si amorphization through amorphous strain accommodation6–8 whereas at 600 K (a temperature at which Si remains crystalline under irradiation at our conditions), the nanocavities were efficient sinks for interstitials, thus leading to a dislocation-denuded zone in the nanocavity vicinity.4 (2) Conversely, the mechanisms involved in nanocavity reduction differ depending on the (crystalline or amorphous) structure of the surrounding Si host.5,9 In the crystalline phase, the observed decrease in diameter is ascribed to the gettering of interstitials. When the matrix surrounding the cavities is amorphous, nanocavity shrinkage is significantly faster.

The present work aimed at improving our understanding of the shrinkage mechanism in the amorphous phase. The nanocavity shrinkage during Si or As irradiation was monitored by in situ transmission electron microscopy (TEM) observations. In both cases, the nanocavity diameter decreased linearly as the ion fluence increased. The shrinkage rate increased as the ion mass increased, but did not depend on the temperature. The most straightforward description of the nanocavity reduction process at this stage is simply in terms of the ballistic effect associated with the incoming beam: atoms from the host Si surrounding the nanocavity recoil into the latter, thus progressively filling it up. The limited plasticity in the amorphous phase is sufficient to allow adjustment of the nanocavity surroundings to the volume change.

Nanocavities were formed in (100) Si substrates by 50 keV H implantation at a fluence of $3 \times 10^{16}/\text{cm}^2$ and thermal annealing in Ar for 1 h at 1300 K. Irradiation-induced nanocavity evolution was studied in situ with a CM12 Philips electron microscope on line with an ion implanter.10 Cross-sectional TEM samples prepared from bulk c-Si substrates containing nanocavities were amorphized by irradiating both sides at room temperature with a fluence of $3 \times 10^{15}/\text{cm}^2$ Si ions at 50 keV. These amorphous samples, containing 10–25 nm diam nanocavities, were then irradiated in situ in the TEM with either 100 keV Si ions or 300 keV As ions at 300 or 600 K. Irradiation-induced nanocavity evolution during

![FIG. 1. Typical evolution of three nanocavities under As irradiation at room temperature: (a) before irradiation, and (b) after irradiation of 2.1 $\times 10^{14}$ As/cm$^2$. Irradiation at 600 K leads to similar results.](image)
irradiation was thus measurable directly. The ion energies were such that the projected ranges (150 and 180 nm for Si and As, respectively) far exceeded the TEM sample thickness (<70 nm). Hence no impurities were introduced into the sample during the irradiation.

Figures 1 and 2 show the evolution of nanocavities during irradiation. They remained spherical, independent of the temperature or irradiation fluence, as well as of the pre-irradiation nanocavity size or of the irradiating ion mass. Within experimental uncertainties, no significant temperature dependence of the shrinkage rate was observed. On the other hand, as shown in Fig. 2 there was a five fold increase (from 4 to 20 nm per $10^{15}$ ions/cm$^2$) in the shrinkage rate when irradiating with As rather than Si ions. The nanocavity diameter varied linearly with the ion fluence (Fig. 3) in all cases, while the shrinkage rate did not depend on the pre-irradiation nanocavity size (Fig. 4).

The absence of any significant temperature dependence of the nanocavity reduction demonstrates that mechanisms involving diffusion processes, such as vacancy diffusion in the amorphous phase, are not rate limiting. The fact that the shrinkage rate does not depend on the initial nanocavity size indicates that the nanocavity reduction process occurs at or very near the nanocavity/matrix interface as individual collision cascades intersect it. Within the resolution of the TEM, we observed no change in nanocavity shape during irradiation for decreases in diameter as small as 0.5 nm, i.e., corresponding to an additional Si monolayer at the nanocavity internal surface. These results suggest a simple ballistic process at the nanocavity/matrix interface due to Si recoils from incoming collisions with Si or As ions at depths where nanocavities occur. From calculations with the SRIM code, we estimate the corresponding average recoil density (7 and 14 displacements/ion/nm, respectively) and average recoil energy of Si atoms (30 and 100 eV, respectively). Assuming that for Si irradiation only interface atoms can fill the cavities while for As irradiation the recoil energy is sufficient to allow atoms belonging to the second layer from the interface to reach the nanocavity, we can reproduce the observed shrinkage rates to within about 30%. Experiments are in progress to estimate possible additional contributions that involve strain phenomena and/or plastic flow of the amorphous phase into the nanocavity which may guarantee that the nanocavity volume is kinetically free to minimize the internal surface energy. The fact remains that a ballistic model may account for major features observed for nanocavities in the amorphous phase. On the other hand, cavity shrinkage in c-Si is much slower (by a factor larger than 2, presently being measured) than estimated from ballistics. Whether this is due to significant vacancy–interstitial recombination in c-Si, or to ion-induced epitaxial crystallization at the cavity internal surface, remains to be determined.

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