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Plasma hydrogenation of strained Si/SiGe/Si heterostructure for layer transfer without ion implantation

Lin Shao,[a] Yuan Lin, J. K. Lee, Q. X. Jia, Yongqiang Wang, and M. Nastasi
Los Alamos National Laboratory, Los Alamos, New Mexico 87545

Phillip E. Thompson
Code 6812, Naval Research Laboratory, Washington, DC 20375-5347

N. David Theodore
Advanced Products Research and Development Laboratory, Freescale Semiconductor Inc., Tempe, Arizona 85284

Paul K. Chu
Department of Physics and Materials Science, City University of Hong Kong, Hong Kong, People's Republic of China

T. L. Alford and J. W. Mayer
Department of Chemical and Materials Engineering, Arizona State University, Tempe, Arizona 85287

Peng Chen and S. S. Lau
University of California at San Diego, San Diego, California 92093

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We have developed an innovative approach without the use of ion implantation to transfer a high-quality thin Si layer for the fabrication of silicon-on-insulator wafers. The technique uses a buried strained SiGe layer, a few nanometers in thickness, to provide H trapping centers. In conjunction with H plasma hydrogenation, lift-off of the top Si layer can be realized with cleavage occurring at the depth of the strained SiGe layer. This technique avoids irradiation damage within the top Si layer that typically results from ion implantation used to create H trapping regions in the conventional ion-cut method. We explain the strain-facilitated layer transfer as being due to preferential vacancy aggregation within the strained layer and subsequent trapping of hydrogen, which lead to cracking in a well controlled manner. © *2005 American Institute of Physics*.
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In the ion-cut technique for layer transfer,[1] implanted hydrogen atoms (at a dose of a few $10^{16}/cm^2$) create hydrogen-terminated cavities upon annealing that finally evolve into cracks parallel to the surface. Growth and joining of these cracks allows the surface layer to become completely separated from the substrate. This technique has become one of the mainstream methods used for the fabrication of silicon-on-insulator wafers.[2,3] The thickness of the transferred layer in the conventional ion-cut technique (Smart-cut™) is controlled by adjusting the energy of the incoming H ions.[1] For transfer of layers of less than 20 nm thickness, which is demanded by the international technology roadmap for the fabrication of fully depleted complementary metal-oxide semiconductor devices,[4] the ion-implantation-based ion-cut technique has difficulty producing effective layer transfer even if low energy hydrogen implantation is used.[5] Various techniques have been developed for the transfer of ultrathin layers.[6–11] Recently, it was shown that thin layer transfer can be realized by plasma hydrogenation in conjunction with a H trapping layer created by heavier ion implantation.[11] However, heavy ion bombardment significantly degrades the Si crystalline quality.

We have developed a thin layer transfer process using a strained SiGe layer as a H trapping layer. The splitting location is controlled by the depth of the strained SiGe layer. Therefore, the region damaged during the introduction of hydrogen is not coupled with the location of the layer splitting. This approach has a significant effect on improving the crystalline quality of the transferred thin Si layer.

Our process began by fabricating a heterostructure with a thin SiGe layer buried inside monocrystalline Si. A 5-nm-thick epitaxial $Si_{0.8}Ge_{0.2}$ layer was grown on a (100) Si substrate and a 190-nm-thick crystalline Si layer was grown on top by molecular beam epitaxy (MBE) at 650 °C. The samples were subsequently hydrogenated using a hydrogen plasma at 250–300 °C for 1 h and then at 300–350 °C for 2 more hours. The applied bias during hydrogenation was −500 V. For the purpose of comparison, a virgin Si wafer without a strained layer was hydrogenated simultaneously.

The strain in the sample was measured using high resolution x-ray diffraction analysis. An $\omega - 2\theta$ scan around Si(004) was collected using a Bede D1 x-ray diffractometer. After hydrogenation, an optical microscope with Nomarski contrast was used to check for surface bubble formation. Infrared (IR) absorption spectroscopy was used to investigate the interaction of hydrogen with defects. The hydrogen atomic profiles were measured with elastic recoil detection (ERD) analysis. ERD analysis was performed using a 3.0 MeV $^4He^+$ analyzing beam. The beam was oriented 75.25° from the sample normal and the detector was positioned 150.5° from the incident beam. To avoid detection of backscattered He particles, a 14-μm-thick Mylar foil was placed

[a]Author to whom correspondence should be addressed; electronic mail: lshao@mailaps.org

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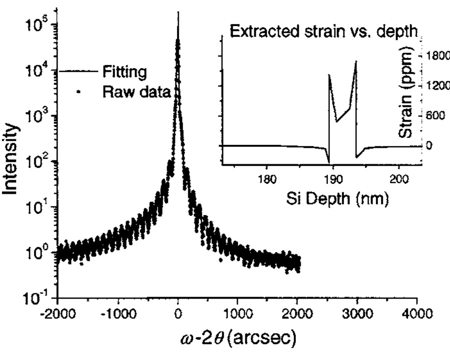

FIG. 1. X-ray ω-2θ scans around the Si(004) diffraction peak, obtained from as-grown MBE Si and the corresponding Bede RADS auto-fit to the data. The inset shows the out-of-plane strain as a function of depth.

between the sample and the hydrogen detector. Transmission electron microscopy (TEM) was used to characterize the structure of the samples.

Figure 1 shows the x-ray diffraction intensity resulting from an $\omega-2\theta$ scan around the Si(004) diffraction peak. The inset shows the strain-depth profiles of the as-grown sample from the Bede RADS auto-fit. The buried SiGe layer has a tensile out-of-plane strain peaking at 1700 ppm. The strains peak at both interfaces of the SiGe/Si. The deposited Si layer and Si substrate have compressive out-of-plane strains near the interfaces with the SiGe layer. The strain in the Si gradually decreases with increasing distance from the SiGe layer, disappearing for distances greater than 8 nm. It is also noted that the thicknesses of the crystalline Si layer and the epitaxial $Si_{0.8}Ge_{0.2}$ extracted from x-ray diffraction are the same as the specifications of the sample fabricated by MBE.

We have observed long range H migration and subsequent H trapping by the buried strained $Si_{0.8}Ge_{0.2}$ layer. ERD H profiles after plasma hydrogenation are shown in Fig. 2. The trapped hydrogen in the MBE sample has a peak concentration of $9.6\times10^{20}/cm^3$ at the depth of the strained $Si_{0.8}Ge_{0.2}$ layer. In the hydrogenated control sample (virgin Si), H is only distributed narrowly in the near surface region. Since H contamination near the surface is present whether or not the sample has been hydrogenated, ERD spectra have been normalized by subtracting the nonhydrogenated spectrum.

Figure 3 shows infrared-absorption measurements from both the hydrogenated MBE sample and the control sample. The MBE sample exhibits enhanced H trapping by defects such as vacancies and internal surfaces, as evidenced by enhanced modes at 2069, 2108–2119, and 2141 cm^{-1}. The modes at 2060–2120 cm^{-1} are related to hydrogen that is trapped at extended internal surfaces, and the mode at 2069 cm^{-1} is close to a value that has previously been assigned to a single hydrogen atom bound to a monovacancy defect.[12] The small features at 2108 and 2119 cm^{-1} are due to symmetric and asymmetric $Si-H_2$ stretching motions on atomically rough (100) internal surfaces.[12] The peak observed at 2141 cm^{-1} is close to the value previously assigned to a vacancy containing two hydrogen atoms within it.[13,14]

Microcracks are present in the MBE Si as revealed by cross-section TEM micrographs. Figure 4(a) shows that the hydrogenated control sample has a damaged layer near the surface due to the dc bias applied to the substrate during hydrogenation. No microcracks were observed in the control

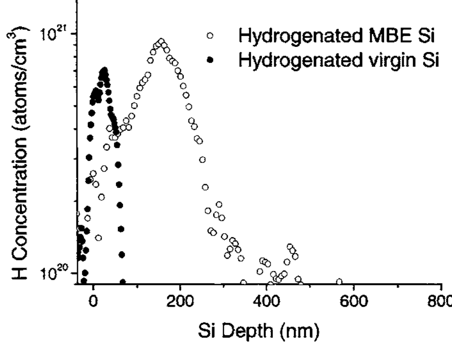

FIG. 2. Hydrogen ERD spectra from a hydrogenated MBE sample (with a strained $Si_{0.8}Ge_{0.2}$ layer buried at a depth of 190 nm) and a hydrogenated control Si sample (without the strained layer). The ERD spectra have been normalized by subtracting the nonhydrogenated spectrum.

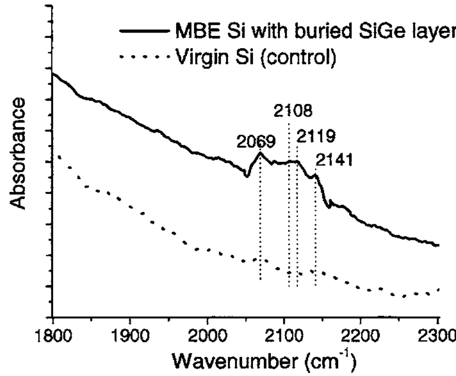

FIG. 3. Infrared absorption spectra from the hydrogenated MBE sample (with a strained layer) and the hydrogenated control Si sample (without a strained layer).

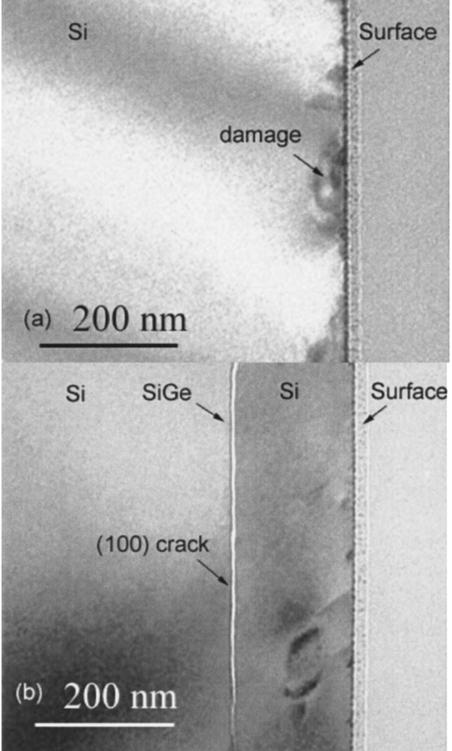

FIG. 4. TEM micrographs obtained from (a) the hydrogenated control Si sample (with a strained layer), and (b) the hydrogenated MBE sample (with a strained layer).

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sample. Figure 4(b) shows a continuous (100) crack, parallel to the Si surface in the hydrogenated MBE sample. It is evident from the TEM micrographs that our approach is capable of controlling the location of the crack by growing a thin strained layer at a desired depth.

It is worth noting that Fig. 2 shows that the splitting occurs for a H peak concentration of $\sim 2\%$, which is much lower than the threshold H concentration ($\sim 5\%$) for splitting in conventional H ion implantation.[5] The high efficiency of splitting in our approach might be explained by the high efficiency of H interaction with vacancy defects (mode $> 2050\ \mathrm{cm}^{-1}$) and the absence of H interaction with interstitial defects (mode $< 2050\ \mathrm{cm}^{-1}$),[14] as evidenced by the IR data in Fig. 3.

The equilibrium vacancy concentration in Si at 650 °C is approximately $1 \times 10^{12}/\mathrm{cm}^3$.[15] Including the effect of strain-induced changes on vacancy formation energy,[16,17] our estimated equilibrium vacancy concentration in the SiGe layer is less than $1 \times 10^{13}/\mathrm{cm}^3$. This value is too low to account for any substantial H trapping. However, gettering sites may also arise from hydrogenation-induced vacancies. Based on our previous studies on the nucleation and growth of platelets in hydrogen-ion-implanted Si,[18] we believe that this strain-facilitated splitting results from two successive steps: (1) preferential vacancy aggregation in the compressively in-plane strained layer and (2) facilitated H platelet formation due to the interaction of H with these vacancy aggregates. Previous studies have shown an outward relaxation of the lattice around a self-interstitial in Si and an inward relaxation of the lattice around a vacancy.[19–21] Accumulation of interstitials within the compressively in-plane-strained SiGe layer is mediated primarily by vacancies diffusing towards the strained layer.[20] Furthermore, clustering of vacancies is enhanced by the higher mobility of the vacancies along the compressive strain layer.[22] High resolution electron microscopy of electron irradiated strained Si has shown preferential aggregation of vacancies on (001) planes within the SiGe layer.[20] Various experiments have provided evidence that vacancies play a central role in H platelet formation.[12,13,23,24] Chemical interaction of H with dangling bonds in vacancy-type defects, and subsequent trapping of hydrogen lead to the formation of H platelets and ultimately to the formation of H_2 gas bubbles. The building up of internal pressure with annealing will essentially lead to cracking. Recent first-principle calculation has proposed that hydrogen-vacancy clusters are the precursors for evolution into internal hydrogenated surfaces.[24]

From the viewpoint of throughput, MBE has limitations for growth on large size wafers. The required heterostructure can be fabricated by chemical vapor deposition, which has been widely used in high-flow processes.

In summary, we have demonstrated a new approach to transfer a thin Si layer by using a buried SiGe layer to provide H trapping centers during hydrogenation. This approach does not require any ion implantation steps. Blistering and subsequent layer separation can be controlled to occur within the SiGe layer. We explain this behavior as being due to preferential vacancy aggregation within the strained layer and subsequent trapping of hydrogen that lead to localized cracking in a well controlled manner.

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