Chemical and electrical properties of cavities in silicon and germanium

S.M. Myers, D.M. Follstaedt, G.A. Petersen, C.H. Seager, H.J. Stein, and W.R. Wampler

Sandia National Laboratories, Albuquerque, New Mexico 87185-1056 U.S.A.

Abstract

Cavities were formed in Si and Ge by He ion implantation and annealing, and resultant chemical and electrical properties were investigated. The dissociation energies for Si-H and Ge-H surface monohydride bonds were determined, showing that H chemisorption on Si is energetically stable with respect H₂ gas whereas H chemisorption on Ge is not. Cavity walls in Si were found to trap transition metals strongly, suggesting application to impurity gettering in devices. Measurement and modeling of cavity electrical properties elucidated surface electronic states and indicated a potential for controlled electrical isolation in devices.

1. Introduction

Stable faceted cavities can be formed in Si [1-3] or Ge [4,5] by ion-implanting He to form bubbles and then annealing at temperatures ≥ 1000 K to remove damage, promote cavity enlargement and shape equilibration, and induce He permeation from the material. A representative microstructure is shown in Fig. 1. The pristine internal surfaces of such voids have $>10^{14}$ dangling orbitals per cm² after reconstruction, and this, coupled with the near absence of steric hindrance, leads to strong interactions with both solutes and charge carriers.

In this paper we describe quantitative mechanistic studies of the reactions with cavities of hydrogen, the metal solutes Fe, Co, Ni, Cu and Au, and charge carriers. In the case of H, our principal motivation was to determine the strength of H binding on the Si and Ge surfaces, a property important to both passivation [6] and chemical vapor deposition [7] that has nevertheless proved inaccessible to ultrahigh-vacuum (UHV) experiments on external surfaces. Trapping of transition metals by cavities in Si may provide gettering that is superior to presently used methods, which are mostly based on induced precipitation of metal-silicides [8]. The

MASTER

DISTRIBUTION OF THIS DOCUMENT IS UNLIMITED

ED

1995

TI

DISCLAIMER

This report was prepared as an account of work sponsored by an agency of the United States Government. Neither the United States Government nor any agency thereof, nor any of their employees, makes any warranty, express or implied, or assumes any legal liability or responsibility for the accuracy, completeness, or usefulness of any information, apparatus, product, or process disclosed, or represents that its use would not infringe privately owned rights. Reference herein to any specific commercial product, process, or service by trade name, trademark, manufacturer, or otherwise does not necessarily constitute or imply its endorsement, recommendation, or favoring by the United States Government or any agency thereof. The views and opinions of authors expressed herein do not necessarily state or reflect those of the United States Government or any agency thereof.

DISCLAIMER

Portions of this document may be illegible in electronic image products. Images are produced from the best available original document.

electrical activity of cavities will constrain their proximity to devices in gettering applications but in other contexts may provide useful electrical isolation.

2. Mathematical description of solute-binding processes

The quantitative interpretation of experimental data in this work required a mathematical model relating local binding mechanisms to the observed internal redistribution and external release of solute atoms. We constructed such a model as described elsewhere [9] by applying the diffusion equation with source terms to take account of the exchange of solute atoms between bound states and mobile solution sites. Here we omit most mathematical details and simply give equations for the solution concentration in equilibrium with the three types of bound state to be considered. Since equilibrium is closely approached locally in the vicinity of individual binding centers for conditions of interest, these equations convey the essential physical aspects of the model. For precipitation as a solid phase (p), such as a metal silicide,

$$C_{sp} = C_0 p \exp(-Q_p/kT) \quad (1)$$

where the solution concentration C_s is expressed as atomic fraction, C_0 is a temperature-independent prefactor, and Q is the binding energy relative to solution. If the precipitated phase is a diatomic gas (g), such as H_2 ,

$$C_{sg} = C_0 g F^{1/2} \exp(-Q_g/kT) \quad (2)$$

where F is the fugacity of the gas and equal to pressure at lower pressures. The equation for static trap sites (t) that are reversible, saturable, and independently acting is

$$C_{st} = \{\theta/[1 - \theta]\} \exp(-Q_t/kT) \quad (3)$$

where θ is the average fractional occupancy of the traps, and the change in vibrational entropy between solution and the trapped state is neglected.

In these studies we make the approximation that binding sites on the walls of cavities behave independently, so that occupation of one site does not influence the reactive properties of neighbors. Equation (3) then becomes applicable to solute trapping on cavity walls. The

corresponding physical picture lattice-gas chemisorption, as opposed, for example, to two-dimensional ordered islands on the internal surfaces. The approximation of noninteracting surface traps becomes increasingly accurate as 1) solute-solute attraction becomes weak in comparison to the chemisorption energy, 2) fractional surface coverage decreases, and 3) temperature increases.

Equations (1) and (3) differ in ways that have important consequences. For example, C_{sp} is independent of the number of precipitated atoms, being in fact the thermodynamic solid solubility characteristic of the precipitated phase, whereas $C_{st} \propto \theta$ for $\theta \ll 1$. As a result, solid precipitates form only when the concentration of solute atoms exceeds $C_{0p} \exp(-Q_p/kT)$, while trapping persists to arbitrarily small solute concentrations. Furthermore, when traps described by Eq. (3) are present at more than one location within the host, the dependence of C_{st} on θ provides a driving force for diffusive redistribution leading to equalization of θ . In contrast, for precipitation described by Eq. (1) in an isothermal solid, there is no driving force for phase dissolution in one location accompanied by growth in another.

3. Interactions of H with cavities in Si and Ge

We are concerned here with the stability of the most strongly bound, monohydride state of H on the semiconductor surfaces. The local atomic configuration consists of a surface Si or Ge atom attached to one H atom and also to three host atoms of the underlying substrate [6,10]. Conceptually, the dissociation energy of the semiconductor-H bond might be determined in two ways: first, by measurement of H thermal desorption from the surface, yielding a desorption activation energy that is related to the bond energy; and, second, by observation of thermodynamic equilibrium between the chemisorbed state and H_2 gas, whose thermodynamics are well known. Both approaches encounter difficulties, however, when carried out on external surfaces in UHV. For thermal desorption from external surfaces, which occurs through endothermic dissociation of two semiconductor-H bonds and exothermic H-H

recombination, the relationship between the measured activation energy E_D and the bond dissociation energy E_B is

$$E_D = 2E_B - E_R + E_A \quad (1)$$

where E_R is the dissociation energy of H_2 , and E_A is the activation energy for dissociative adsorption. While the adsorption barrier has not been directly measured, the observed difficulty of chemisorption from the molecular gas onto Si and Ge suggests that E_A is not small. Hence, measurement of E_D does not determine E_B . Moreover, the same adsorption barrier has precluded observation of thermodynamic equilibrium between externally chemisorbed H and H_2 gas at the low gas pressures consistent with clean surface conditions.

Both of the above difficulties can be circumvented by using internal cavity surfaces. In the case of thermal release, the rate-controlling processes are now the promotion of H atoms from chemisorption to solution and subsequent diffusion to the surface, which can be modeled in terms of Eq. (3) and the independently known H diffusion coefficient. Moreover, a diffusion-mediated thermodynamic equilibrium between internally chemisorbed H and external H_2 gas can be achieved by using sufficiently high gas pressures at elevated temperatures. We carried out both types of experiments on Si, utilizing nuclear-reaction-analysis (NRA) to monitor the amount of H in the cavity layer and Fourier-transform infrared spectroscopy (FTIR) to observe monohydride bonding on the internal surfaces [11,12]. Thermal release yielded $Q_t = 2.1 \pm 0.2$ eV expressed relative to H in solution in Si, or, equivalently, $E_D = 2.5$ eV when the reference H state is the dissociated atom in vacuum. Equilibration with external gas yielded $Q_t = 2.2 \pm 0.1$ eV or $E_D = 2.6$ eV, consistent with the above result. Since the dissociation energy of the H_2 molecule is 2.26 eV per atom, the chemisorbed state is energetically preferred to the gas.

We have now carried out similar experiments on H in Ge with cavities [5]. Temperature-ramp data obtained using NRA are shown in Fig. 2, where the circles represent specimens with cavities and the triangles are from a reference sample having none. The H was introduced by

ion implantation, and, in the absence of cavities, the observed release reflects defect trapping. When cavities are present, the H is released in a single, sharply defined stage near 900 K, reflecting an interaction of H with the voids.

The model curves in Fig. 2 were calculated [9] under the assumption that the H undergoes release by moving into solution from either H₂ gas in the cavities (Eq. 2) or internal-surface chemisorption (Eq. 3) and diffusing to the external surface. Apart from the trial values of the internal-chemisorption bond dissociation energy listed on the figure, there are no free parameters; transmission electron microscopy (TEM) provided the combined cavity volume (4×10^{10} cm³ per cm² of sample area) and number of chemisorption sites (3×10^{15} per cm² of sample area), and the H diffusion coefficient and solubility in equilibrium with H₂ gas were taken from the literature [13]. The experimental data are seen to be described well by assuming that the bound H exists entirely as encapsulated H₂ gas with no chemisorption occurring on the cavity walls. This is in marked contrast to our results for Si. It implies that $E_D \leq 2.0$ eV.

The predominance of the gas phase in Ge is reinforced by FTIR spectra shown in Fig. 3, which were obtained from specimens treated similarly to those of Fig. 2. In the absence of cavities, Si-H stretching vibrations associated with H-defect centers persist until the H is released from the material. In the presence of cavities, most of the IR absorption intensity is lost below 773 K, even though from Fig. 2 the H remains within the cavity layer until about 900 K. This is explained by H₂ formation within the cavities.

The FTIR data for cavity-containing Ge annealed in the range 773-873 K retain low-amplitude features that correspond to a combination of the spectra reported for monohydrides on well characterized (111) and (100) Ge surfaces [14,15]. This indicates that internal chemisorption is not entirely absent in the cavities; from the area under these spectral features, we estimate that about one chemisorption site in five is occupied. Given this, and taking into account the known total amount of H and the known cavity volume, Eqs. (2) and (3) can be

used to obtain $E_D \approx Q_t = 1.9 \pm 0.2$ eV. (Appreciable redistribution of H between cavity wall and encapsulated gas during cooling prior to FTIR analysis was taken into account.) Thus, chemisorption on the Ge surface is energetically unstable relative to the molecular gas.

4. Gettering of transition metals by cavities in Si

The walls of cavities may bind metal impurities by chemisorption-like trapping involving coverages $\lesssim 1$ monolayer (ML) and also by the nucleation and growth of three-dimensional silicide phases. These two processes have distinctive characteristics that make each of interest for impurity gettering in Si devices, so that the occurrence of both in cavities is fortunate. As discussed with reference to Eqs. (1) and (3), chemisorption trapping is expected to persist to infinitesimal solute concentrations, whereas silicide precipitation requires concentrations above the solid solubility. This is particularly significant for future technologies where impurity tolerances may decrease to 10^9 atoms/cm³ [16]. Moreover, given the propensity of metal impurities to precipitate in undesirable locations such as the gate-oxide interface, the ability of cavity-wall trapping to dissolve pre-existing silicides may prove advantageous. Chemisorption trapping saturates, however, while gettering by silicide precipitation in the cavities does not. As regards binding energy, the results discussed below indicate that chemisorption trapping is stronger than precipitation for some metals and weaker for others. Our studies have concentrated primarily on chemisorption trapping, which is the focus of the present discussion. Previously reported findings for Cu and Ni [17-20] are expanded here and extended to include Co, Fe, and Au. A complementary effort at the Australian National University (ANU) has investigated gettering by cavities formed by H ion implantation and has given particular attention to silicide precipitation into cavities from supersaturated solutions, which we have not examined [21-24].

Trapping effects were separated from silicide formation in the cavities in these studies by using remotely located silicide precipitates as the source of metal atoms. As discussed in

Sect. 2, there is no driving force for silicide dissolution in one region accompanied by precipitation of the same phase in another region. This approach is exemplified in Fig. 4, which plots the accumulation of Cu in cavities during heating at 873 K as measured by Rutherford backscattering spectrometry (RBS). The source of the Cu is a precipitated layer of Cu_3Si , formed on the opposite side of the wafer by ion implantation and annealing and characterized by TEM [19]. The flow of Cu to the cavities initially occurs at a constant rate, as expected for steady-state diffusion over the wafer thickness of 250 μm to a strong sink. Moreover, the diffusivity-solubility product extracted from the model fit in the picture [18,20] is consistent with reported values of the two quantities [25]. Cavity saturation occurs near $5 \times 10^{15} \text{ Cu/cm}^2$, corresponding quantitatively to monolayer coverage of the internal surface area calculated from TEM images. Lattice imaging confirmed the absence of silicide precipitation in the cavities after saturation, as seen in the inset. Gold behaved very similarly to Cu in experiments at 1123 K, diffusing from the precipitated equilibrium eutectic to a cavity layer until saturation occurred near monolayer coverage.

Figures 5 and 6 show depth profiles of Cu, Ni, Co, and Fe in the vicinity of cavity layers after diffusion-mediated equilibration with pre-existing metal silicide on the opposite side of the wafer at a distance of 250 μm . The profile for Au, not shown, is very similar to that for Cu. Profiling was accomplished by RBS for Cu and Ni and by secondary-ion mass spectrometry (SIMS) for Co and Fe. Silicides were formed by implanting the metal and heating. The phase was confirmed by TEM for Cu and Ni, while for Co and Fe we used silicide-formation procedures reported in the literature [26,27]. The width and shape of the metal depth profiles are seen to conform closely to the depth distribution of implanted He calculated from the TRIM Monte-Carlo code [28]. This provides semiquantitative evidence that the metal concentrations vary with depth in proportion to the available cavity surface area, equalizing the fractional site occupation θ as expected from Eq. (3).

The integrated areal densities of bound metal atoms in Figs. 5 and 6 vary over more than two orders of magnitude, and we ascribe this largely to corresponding differences in the stability of the chemisorbed state relative to the silicide phase with which it is equilibrated. In this regard it is noteworthy that the silicide binding energy relative to solution rises by ~ 1 eV on going from Cu and Ni to Co and Fe [25], correlating with a large decrease in the number of metal atoms bound to the cavities.

We quantify these considerations by first estimating the fractional coverage θ in Eq. (3) and then combining Eqs. (1) and (3) to extract Q_t for the trapping, the solubility parameters in Eq. (1) being taken from Ref. 25. To obtain θ we divide the measured areal density of metal atoms in the cavities by the areal density corresponding to saturation. The latter quantity is equated to the product of the cavity surface area per unit sample area obtained from TEM and the atomic density of ordered chemisorption structures on the (111) surface reported from UHV studies on external surfaces. The saturation chemisorption density varies significantly among the metals, being ~ 1 ML for Cu and Au, $1/19$ ML for Ni, $1/7$ ML for Co, and $\sim 1/2$ ML for Fe [29-33]. In the cases of Cu, Au, and Ni, the calculated values of θ are indistinguishable from saturation, being respectively 0.8, 1.1, and 1.5. Hence, the deduced binding energies represent lower bounds: ≥ 1.3 eV for Cu, ≥ 2.0 eV for Au, and ≥ 1.5 eV for Ni. For Co and Fe the calculated values of θ are 0.018 and 0.012, respectively, giving $Q_t \approx 1.7$ and 1.6 eV. In the case of Cu, we also carried out detailed experiments observing the diffusive redistribution of metal atoms between two cavity layers, one initially saturated with Cu ($\theta = 1$) and the other initially unoccupied ($\theta = 0$), leading to a final condition where θ is the same in both layers [18,20]. This permitted us to quantify the cavity-wall binding energy for Cu, giving $Q_t = 2.2 \pm 0.2$ eV.

Summarizing, we find that the solutes Cu, Au, Ni, Co, and Fe undergo chemisorption trapping at cavities even when the metal atoms are initially precipitated as metal silicides in a

remote region of the wafer. In the case of Cu, the deduced binding energy for cavity trapping is about 0.5 eV greater than that in the silicide phase, whereas for Co and Fe, the silicides exhibit the greater binding energy. Cavity trapping occurs even in the latter cases, however. We ascribe this to the persistence of trapping to arbitrarily small solution concentrations that was discussed above. In physical terms, this property arises from the fact that the configurational entropy per trapped atom becomes large when only a few atoms occupy many trap sites; such an effect does not occur for silicide precipitation, where the entropy per atom is independent of the number of bound atoms. When these findings concerning chemisorption trapping are combined with the observation by ANU workers of apparently uninhibited precipitation of silicides in cavities when the Si matrix is supersaturated [21-24], cavities appear particularly promising for gettering.

5. Electrical properties of cavities in Si

Dangling orbitals in Si can have both positive and negative charge states within the bandgap, as seen, for example, from studies of the vacancy and divacancy [34], and this suggests that cavities also are electrically active. We have explored the electrical properties of cavity-containing Si using deep-level transient spectroscopy (DLTS), capacitance versus voltage (C-V), and conductance versus temperature [35]. These studies are relevant to the use of cavities for gettering and electrical isolation and also illuminate fundamental surface properties. The experiments utilized cavity layers formed with the He dose and annealing condition of Fig. 1, but with a He energy of 300 keV that positioned the peak cavity density at a depth of 1.3 μm .

In both p-type and n-type Si, cavities produce a strong barrier to current flow with an activation energy of several tenths of an eV and a capacitively measured thickness of $\sim 1 \mu\text{m}$. This indicates that the cavity layer is positively charged when the Fermi level is close to the valence band and negative when the level approaches the conduction band. A more

quantitative indication of level positions was obtained in DLTS experiments, where we observed capacitance transients associated with thermal emission of electrons and holes from cavity states as a function of temperature under three conditions: 1) n-type Si approaching zero-current equilibrium by electron emission after injection of excess electrons by a voltage pulse; 2) p-type Si approaching equilibrium by hole emission after injection of excess holes; and 3) emission of holes from cavities in p-type Si after reverse biasing of a Schottky diode to deplete carriers in the region of the cavity layer. The observed emission rates are plotted in an Arrhenius format in Fig. 7. We interpret these transitions in terms of positive, neutral, and negative charge states of the cavity dangling orbitals as indicated in the inset.

Knowledge of the cavity-wall electrical levels permitted us to model the electrical behavior of the system, both in equilibrium and under bias, by solving the Poisson equation numerically while taking account of electron and hole capture and emission by the three charge states of the dangling orbitals. In this way we semiquantitatively accounted for a wide range of novel electrical behaviors associated with the cavities [35]. (Since publication the model was refined to treat the strong coulomb repulsion between charges within the same cavity explicitly, rather than assuming that the repulsion prevents more than one charge per cavity. This improved the agreement between experiment and model.) A typical model calculation is represented in Fig. 8, and it illustrates two key points: first, cavity charging reduces carrier density by several orders of magnitude, thereby producing electrical isolation; and, second, carrier redistribution causes the electrical perturbation to extend several Debye lengths beyond the cavity layer.

6. Conclusion

Cavities formed in Si and Ge by He ion implantation combine microstructural stability with high chemical and electrical reactivity, and this provides unique opportunities for the investigation and manipulation of the semiconductor properties. In this paper we have reported

1) first determinations of Si-H and Ge-H surface-bond dissociation energies, 2) strong cavity-wall trapping of transition-metal impurities that may provide superior gettering in Si devices, and 3) characterization of cavity reactions with electrons and holes leading to improved knowledge of Si surface states and the possibility of utilization for electrical isolation in devices.

Acknowledgment

This study was carried out under the the U. S. Dept. of Energy and funded by its Off. of Basic Energy Sciences, Div. of Materials Sciences, under Contract DE-AC04-94AL85000.

References

- [1] C.C. Griffioen, J.H. Evans, P.C. de Jong, and A. Van Veen, Nucl. Instrum. Meth. B 27 (1987) 417.
- [2] D.M. Follstaedt, Appl. Phys. Lett. 62 (1993) 1116.
- [3] D.J. Eaglesham, A.E. White, L.C. Feldman, N. Moriya, and D.C. Jacobson, Phys. Rev. Lett. 70 (1993) 1643.
- [4] D.M. Follstaedt, S.M. Myers, and H.J. Stein, MRS Symp. Proc. 279 (1993) 105.
- [5] S.M. Myers, H.J. Stein, and D.M. Follstaedt, Phys. Rev. B, 1995, in press.
- [6] G.S. Higashi, Y.J. Chabal, G.W. Trucks, and K. Raghavachari, Appl. Phys. Lett. 56 (1990) 656.
- [7] N.M. Russell and W.G. Breiland, J. Appl. Phys. 73 (1993) 3525.
- [8] W. Schröter, M. Seibt, and D. Gilles, in: Materials Science and Technology, vol. 4: Electronic Structure and Properties of Semiconductors, ed. W Schröter (VCH Publ., New York, 1991) pp. 539-589.
- [9] S.M. Myers and D.M. Follstaedt, J. Appl. Phys. 65 (1989) 311.
- [10] J.J. Boland, Phys. Rev. Lett. 65 (1990) 3325.
- [11] S.M. Myers, D.M. Follstaedt, H.J. Stein, and W.R. Wampler, Phys. Rev. B 47 (1993) 13380.

- [12] W.R. Wampler, S.M. Myers, and D.M. Follstaedt, *Phys. Rev. B* 48 (1993) 4492.
- [13] R.C. Frank and J.E. Thomas, Jr., *J. Phys. Chem. Solids* 16 (1960) 144.
- [14] Y.J. Chabal, *Sur. Sci.* 168 (1986) 594.
- [15] J.E. Crowell and G. Lu, *J. Electron Spectros. and Related Phen.* 54-55 (1990) 1045.
- [16] H.R. Huff, in: *Proc. 4th Inter. Symp. on ULSI*, eds. G. Celler, E. Middlesworth, and K. Hoh (ECS, Pennington NJ, 1993) pp. 103-132.
- [17] S.M. Myers, D.M. Bishop, D.M. Follstaedt, H.J. Stein, and W.R. Wampler, *MRS Symp. Proc.* 283 (1993) 549.
- [18] S.M. Myers, D.M. Follstaedt, and D.M. Bishop, *MRS Symp. Proc.* 316 (1994) 33.
- [19] D.M. Follstaedt and S.M. Myers, *MRS Symp. Proc.* 316 (1994) 27.
- [20] S.M. Myers, D.M. Follstaedt, D.M. Bishop, and J.W. Medernach, in: *Semiconductor Silicon 1994*, eds. H.R. Huff, W. Bergholz, and K. Sumino (ECS, Pennington, NJ, 1994) pp.808-819.
- [21] J. Wong-Leung, J.S. Williams, R.G. Elliman, E. Nygren, D.J. Eaglesham, D.C. Jacobson, and J.M. Poate, *Nucl. Instrum. Meth. B*, in press.
- [22] J. Wong-Leung, C. Ascheron, M. Petravic, R.G. Elliman, and J.S. Williams, *Appl. Phys. Lett.*, in press.
- [23] J. Wong-Leung, E. Nygren, J.S. Williams, and D.J. Eaglesham, *MRS Symp. Proc.*, in press.
- [24] J. Wong-Leung and J.S. Williams, private communication.
- [25] E.R. Weber, in: *Properties of Silicon*, EMIS Datareviews Series No. 4 (INSPEC, New York, 1988) pp. 409-452.
- [26] R. Hull, A.E. White, K.T. Short, and J.M. Bonar, *J. Appl. Phys.* 68 (1990) 1629.
- [27] X.W. Lin, J. Desimoni, H. Bernas, Z. Liliental-Weber, and J. Washburn, *MRS Symp. Proc.* 320 (1994) 97.

- [28] J.F. Ziegler, J.P. Biersack, and U. Littmark, *The Stopping and Range of ions in Solids* (Pergamon, New York, 1985).
- [29] J.E. Demuth, U.K. Koehler, R.J. Hamers, and P. Kaplan, *Phys. Rev. Lett.* 62 (1989) 641.
- [30] A. Shibata, Y. Kimura, and K. Takayanagi, *Sur. Sci. Lett.* 273 (1992) L430.
- [31] R.J. Wilson and S. Chiang, *Phys. Rev. Lett.* 58 (1987) 2575.
- [32] P.A. Bennett, M. Copel, D. Cahill, J. Falta, and R.M. Tromp, *Phys. Rev. Lett.* 69 (1992) 1224.
- [33] T. Urano, M. Kaburagi, S. Hongo, and T. Kanaji, *Appl. Sur. Sci.* 41-42 (1989) 103.
- [34] S.J. Pearton, J.W. Corbett, and M. Stavola, *Hydrogen in Crystalline Semiconductors* (Springer-Verlag, Berlin, 1992) pp. 175-182, and citations therein.
- [35] C.H. Seager, S.M. Myers, R.A. Anderson, W.L. Warren, and D.M. Follstaedt, *Phys. Rev. B* 50 (1994) 2458.

Figure captions

1. Transmission electron micrograph of cavities formed in Si by implanting 10^{17} He/cm² at 30 keV and then vacuum annealing at 1173 K for 1 hr.
2. Retention of implanted H in Ge with and without cavities during temperature ramping at 2 K/min.
3. Infrared absorption spectra of ion-implanted H in Si with and without cavities during sequences of 30-min anneals at increasing temperatures.
4. Cross-wafer diffusion of Cu from Cu₃Si precipitates to cavity traps. The TEM lattice image in the inset shows the absence of precipitated silicide in the saturated cavities.
5. Depth profiles of Cu and Ni at cavity layers measured by RBS after diffusion-mediated equilibration with metal-silicide phase on the opposite side of the wafer.
6. Depth profiles of Co and Fe at cavity layers measured by SIMS after diffusion-mediated equilibration with metal-silicide phase on the opposite side of the wafer.

7. Capacitance transients associated with electron and hole emission from cavity layers in Si.

The assumed transitions are depicted on the inset diagram.

8. Calculated depth profiles of charged species in p-type Si containing cavities.

Cavities in He-Implanted Si

[110] cross-section TEM

1×10^{17} He/cm² at 30 keV

1 hour at 900°C

WAVE NUMBER (cm^{-1})