

# Irradiation effects of swift heavy ions on gallium arsenide, silicon and silicon diodes

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**The irradiation effects of high energy lithium, boron, oxygen and silicon ions on crystalline silicon, gallium arsenide, porous silicon and silicon diodes were investigated. The ion energy and fluence were varied over the ranges 30 to 100 MeV and  $10^{11}$  to  $10^{14}$  ions/cm<sup>2</sup> respectively. Semiconductor samples were characterized with the X-ray fluorescence, Photoluminescence, Thermally Stimulated Exo-Electron Emission and Optical Reflectivity techniques. The life-time of minority carriers in crystalline silicon was measured with a pulsed electron beam and the lithium depth distribution in GaAs was measured with the Neutron Depth Profiling technique. The diodes were characterized through electrical measurements. The results of optical reflectivity, life-time of minority carriers and photoluminescence show that swift heavy ions induce defects in the surface region of crystalline silicon. In the ion-irradiated GaAs, migration of silicon, oxygen and lithium atoms from the buried region towards the surface was observed, with orders of magnitude enhancement in the diffusion coefficients. Enhancement in the photoluminescence intensity was observed in the GaAs and porous silicon samples that were irradiated with silicon ions. The trade-off between the turn-off time and the voltage drop in diodes irradiated with different swift heavy ions was also studied.**

## 1. Introduction

EXTENSIVE studies have been carried out in the field of irradiation effects of low and medium energy (keV to a few MeV energy) ions, electrons and neutrons on silicon, germanium and metallic thin films, and a vast amount of data has been gathered by using different techniques. In addition, for semiconductors, attempts have been made to establish a correlation between the damage coefficients for different types of radiations<sup>1-4</sup>. Furthermore, as a part of the space technology mission, the irradiation effects of protons and heavy ions on different types of semiconductor devices have been investigated<sup>5,6</sup>. The energy deposited by swift ions through the process of nuclear collisions has been used for tailoring the switching characteristics of silicon diodes<sup>7,8</sup>.

The irradiation effects of GeV energy heavy ions on metallic and semiconductor targets have been studied earlier<sup>9-11</sup>. In this case, the target thickness was chosen such that the contribution due to nuclear elastic collisions could be neglected. Irradiation effects were studied with techniques such as DLTS (Deep Level Transient Spectroscopy), Hall Voltage, conductivity measurement, etc. However, in these studies, the techniques used were not surface sensitive and thus the structural changes that occurred in the surface region due to ion irradiation could not be investigated. In the case of metallic targets exposed to GeV energy heavy ions, the induced damage was attributed to the energy deposited through electronic energy loss. The Coulomb explosion and heat spike mechanisms were found to play an important role<sup>10-13</sup>. In other experiments carried out on semiconductors using GeV energy heavy ions, it was observed that the energy deposited through the mechanisms of electronic energy loss and inelastic collisions can lead to auto-crystallization in germanium and annealing of defects in GaAs and germanium<sup>9,14</sup>. The irradiation effects of GeV energy uranium ions on *n*-type silicon were studied with the DLTS technique. The results of this study led to the conclusion that the energy deposited was not sufficient to produce extended defects, and therefore silicon is stable against high electronic excitation<sup>13,15,16</sup>. However, in the case of GaAs exposed to GeV energy heavy ions, the annealing of defects through the energy deposited by the process of electronic excitation was observed<sup>14,17</sup>. In these studies, however, the ion-irradiated surface region was not probed with any of the surface sensitive techniques. In view of the data generated with GeV energy ions, it would be interesting to carry out a detailed study on irradiation effects of 30 to 100 MeV heavy ions on semiconductors, and to find out if the energy deposited through electronic energy loss is sufficient for producing defects in the surface region as well as in the bulk of the crystalline silicon, in addition to that produced by nuclear collision. Over this energy range, a large variation in the ratio of the nuclear energy loss to electronic energy loss could be obtained. The Pelletron of the Nuclear Science Center, New Delhi can produce reasonably high particle current in this energy range so that the irradiation period of about one hour is sufficient to ob-

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tain a fluence of  $\sim 5 \times 10^{14}$  ions/cm<sup>2</sup>. In the case of swift ions of lower mass number, the energy deposited in the surface region through the process of electronic loss is greater by almost three orders of magnitude than that deposited through nuclear elastic collisions. In such cases, therefore, the contribution of the nuclear energy loss in inducing defects in the surface region is usually neglected. However, the results of earlier studies indicate that the surface region of silicon is modified even if the nuclear energy loss is extremely small<sup>18</sup>.

To determine whether heavy ions of energies 30 to 100 MeV can induce defects in the surface region of crystalline silicon and GaAs, techniques like X-ray diffraction, X-ray fluorescence and thermally stimulated exo-electron emission, optical reflectivity, etc., are required for probing the surface region. Furthermore, an in-depth understanding of the irradiation effects of swift heavy ions on semiconductor devices is important, since it might lead to some industrial applications<sup>7,19</sup>. The objectives of this work, therefore, were to study the irradiation effects of 30 to 100 MeV heavy ions on (i) the switching characteristics of silicon junction diodes (ii) inducing defects in the surface region of silicon and GaAs (iii) the diffusion coefficients of certain elements in GaAs and (iv) the lifetime of minority carriers in crystalline silicon. In addition, an attempt has been made to improve the photoluminescence of porous silicon by silicon ion irradiation.

## 2. Theory and methodology

### 2.1 Effects of ion-irradiation

*Crystalline silicon.* While passing through a semiconductor, high energy heavy ions deposit a large fraction of energy by the processes of ionization and excitation of electrons (electron energy loss,  $S_e$ ) and a small fraction through elastic collisions with the medium atoms (nuclear energy loss  $S_n$ ). The lattice atoms of the medium are dislodged mostly through the phenomenon of nuclear energy loss. However, studies carried out recently indicate that the energy deposited through  $S_e$  can also induce defects in the surface region of the semiconductor.

By investigating the surface region of the heavy ion-irradiated semiconductor with different surface techniques, it is possible to know whether the atoms of the surface region have been dislodged by the energy deposited through the process of electronic energy loss. For these studies the choice of ions and their energies should be made in such a way that nuclear energy loss in the surface region is very small, below the threshold energy for defect production.

In general, a large number of defects can be produced in a semiconductor through ion irradiation. As a result,

the initial lifetime of the minority carriers,  $\tau_0$  decreases to  $\tau$ , as per the following relation<sup>20</sup>:

$$\frac{1}{\tau} - \frac{1}{\tau_0} = k\phi, \quad (1)$$

where  $k$  is the damage coefficient and  $\phi$  is the particle fluence. The damage coefficient,  $k$  is an indicator of the number of atoms displaced from the lattice, which depends on the ion energy as well as on the semiconductor.

*Silicon diodes.* In the  $n$  region of a silicon diode, the concentration of minority carriers (holes) is maximum near the junction and it decreases exponentially with distance from the junction. As a result, when separation between the junction and the contact is large, the concentration of minority carriers reduces to a negligible value at the end of the  $n$  region, near the contact.

During a forward operation, the number of excess minority carriers (holes) in the  $n$  region is large, and therefore when the applied voltage is suddenly reversed, the magnitude of the reverse current can be equal to or greater than the forward current. Under a reverse biased condition, the concentration of minority carriers decreases continuously with time, and in due course it reaches the distribution corresponding to a reverse bias condition. The time required to remove this stored charge due to the excess minority carriers, is the turn-off time,  $t_{rr}$ . It can be decreased by reducing the concentration of the minority carriers through the process of recombination. On the other hand, if a region of defects is produced away from the junction, where the concentration of the minority carriers is low, then these defect centers are not effective in reducing the  $t_{rr}$ . The location of a region containing defects, relative to the junction, is, therefore important for reducing the turn-off time,  $\tau_{rr}$  in the junction diodes.

In the case of diodes in which the distance between the junction and the  $n$ -side contact is greater than the diffusion length, the turn-off time,  $t_{rr}$ , is related to the life time of minority carriers according to the following relation<sup>21</sup>:

$$t_{rr} = 0.5\tau_p(I_F/I_R), \quad (2)$$

where  $I_F$  and  $I_R$  are respectively the forward and reverse current, and  $\tau_p$  is the lifetime of minority carriers. Due to the displacement of silicon atoms from the lattice, recombination centers are produced and therefore the life time of minority carriers (holes in the  $n$  region)  $\tau$ , decreases as the defect concentration increases. From eq. (2), it is evident that  $t_{rr}$  of a diode can be decreased by decreasing  $\tau_p$ . In the  $n$  region of a diode, the ratio  $I_F/I_R$  may, however, change with defect concentration

and therefore the variation in  $t_{rr}$  with  $\tau_p$  may not be linear. As the region of the induced defect is shifted away from the junction, the number of minority carriers per unit volume decreases with distance, and therefore only a marginal variation in  $t_{rr}$  can be achieved, even if the net decrease in the lifetime of minority carriers in that region is significant. In such a case the main contribution of the induced defects is toward increasing the voltage drop. Since an ion displaces atoms mostly at the end of its trajectory, a region of defects could be produced in the  $n$  region at any desired distance from the junction by varying ion energy.

## 2.2 Thermally stimulated exo-electron emission

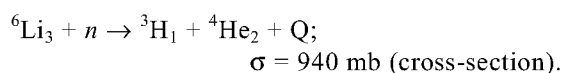
In this thermally stimulated exo-electron emission technique (TSEE)<sup>22</sup>, the sample is heated in vacuum ( $\sim 10^{-6}$  torr) and the electrons emitted from the surface of the sample, called exo-electrons, are collected by a Faraday cup. The sample temperature is increased slowly and the electron current is measured as a function of sample temperature. The rate of temperature rise is kept at  $\sim 100$  K/min. The electronically active trapping centers are identified as peaks in the TSEE spectrum. The energy of the defects can be estimated from the following expression<sup>22</sup>:

$$E_C - E_T = 25 k_B/T_m, \quad (3)$$

where  $E_C$  is the minimum energy of the conduction band,  $E_T$  is the energy of the defects in the surface region,  $k_B$  is the Boltzmann constant and  $T_m$ , the temperature at which peak of the current appears. Using this technique, the energies of defects in GaAs were measured with an accuracy of  $\sim 0.02$  eV. For studying the outdiffusion of the implanted silicon atoms in GaAs, the TSEE spectrum for each GaAs sample was recorded four or five times. However, it was observed that after the second run, there was no change in the nature of the TSEE spectrum, and so only results up to the second run were used for the analysis.

## 2.3 Neutron depth profiling

The depth distribution of lithium atoms in a sample can be measured by the technique of neutron depth profiling. Thermal neutrons, on interacting with lithium, can induce the following nuclear reaction;



The emitted alpha particle  ${}^4\text{He}_2$  and tritium ( ${}^3\text{H}_1$ ) have energies 2.055 MeV and 2.727 MeV respectively. The energies and intensities of  ${}^4\text{He}_2$  and  ${}^3\text{H}_1$  particles, emitted normal to the sample surface, are measured by the

silicon surface barrier detector mounted inside a chamber. The energy loss suffered by these charged particles depends on the depth of the lithium atoms from the sample surface. Hence, the depth distribution of the lithium atoms in a given sample can be obtained from the measured intensity and energy of  ${}^3\text{H}_1$  or  ${}^4\text{He}_2$  particles. The depth information is obtained from the energy loss suffered by the  ${}^3\text{H}_1$  or  ${}^4\text{He}_2$  while exiting the sample. The energy loss per unit length,  $dE/dx$  of these particles in the sample can be estimated by TRIM computer program.

## 3. Experimental

### 3.1 Samples

Silicon samples, each of size  $12\text{ mm} \times 12\text{ mm}$ , were obtained by cutting polished wafers of  $n$ -type silicon (polycrystalline but highly oriented in a  $\langle 111 \rangle$  plane), of thickness  $\sim 350\text{ }\mu\text{m}$  and resistivity in the range of  $130\text{--}150\text{ }\Omega\text{-cm}$ . These samples were chemically cleaned and dried.

Samples of GaAs, each of size  $\sim 10\text{ mm} \times 10\text{ mm} \times 0.2\text{ mm}$ , were obtained by cutting wafers of  $\langle 100 \rangle$   $n$ -type GaAs. These samples were chemically cleaned and dried. The diodes used in this work were made out of  $n$ -type silicon wafers with a thickness of  $\sim 250\text{ }\mu\text{m}$ , diameter  $\sim 100\text{ mm}$  and resistivity  $\sim 60\text{ }\Omega\text{-cm}$ . By diffusing elemental boron in silicon through the thermal process, a  $p$ -type region of thickness  $\sim 60\text{ }\mu\text{m}$  was obtained. For electrical contacts, a thin nickel coating was provided on both the  $n$  and  $p$  sides of each diode. A large number of diodes, each of size  $\sim 2\text{ mm} \times 2\text{ mm}$ , could be made on a  $100\text{ mm}$  diameter silicon wafer. The junction was located at a distance of  $\sim 180\text{ }\mu\text{m}$  from the  $n$ -side contact and  $\sim 60\text{ }\mu\text{m}$  from the  $p$ -side contact. These processed silicon wafers were cut into small square pieces (each of size  $\sim 10\text{ mm} \times 10\text{ mm}$ ) in such a way that each piece had about 16 diodes.

Porous silicon samples used in the work was synthesized in this laboratory.  $P$ -type crystalline silicon with  $0.4\text{ }\Omega\text{-cm}$  resistivity and  $100\text{ }\mu\text{m}$  thick was cut into small pieces ( $10\text{ mm} \times 10\text{ mm}$ ), and porous silicon samples were prepared by the electrochemical method of anodization with ohmic contact on one of its surfaces. The electrolyte solution was a mixture of 48% HF and ethyl alcohol in a 1 : 1 proportion. Anodic current density of  $40\text{ mA cm}^{-2}$  was maintained for a period of 20 min.

### 3.2 Ion irradiation

For ion irradiation, the 15 MV Pelletron facility of Nuclear Science Center, New Delhi, was used. Samples

were mounted on the sample holder of a scattering chamber connected to a beam line. The sample holder consisted of a rectangular rod, 20 cm long and 2 cm wide. At a time, six samples were mounted on each side, with a spacing of about 4 to 5 mm between adjacent samples. This rod was fixed, through a teflon coupler to the electromechanical assembly attached to the top flange of the vacuum chamber. By operating the electromechanical assembly, the rod could be moved up and down as well as rotated about the vertical axis. In this way, any sample could be brought into the path of the ion beam for irradiation. Samples of crystalline silicon, GaAs, silicon diodes and porous silicon were irradiated with lithium, boron, oxygen and silicon ions. The ion energy and fluence were varied in the range of 30 to 100 MeV and  $10^{11}$  to  $10^{14}$  ions/cm<sup>2</sup> respectively. Using an electromagnetic system, the ion beam was scanned to obtain uniform ion intensity over the sample, covering a rectangular area, approximately 12 mm long and 10 mm wide. During ion irradiation, a pressure of  $\sim 10^{-6}$  torr was maintained around the sample. The ion fluence was estimated by measuring the total charge collected by the sample. After irradiating a sample to the desired fluence, the beam was turned off and a new sample was brought into position for irradiation. Details of ions and their energies used are given in Table 1.

### 3.3 Measurements

The lifetime of minority carriers,  $\tau$  (holes in *n*-type silicon) in each silicon sample, before and after ion irradiation, was measured using a pulsed 1 MeV electron beam. The pulse width and the pulse repetition rate of the electron beam were  $\sim 1$   $\mu$ s and 50 pulses per second respectively. The experimental details and the method used have been outlined in our earlier publication<sup>20</sup>. The lifetime of minority carriers,  $\tau$ , could be measured simultaneously in the front side (surface region) and in the rear side of the sample with an accuracy of  $\pm 0.50$   $\mu$ s. In the virgin silicon samples, the lifetime of minority carriers (holes) was around 30  $\mu$ s.

**Table 1.** Details of the ion energies and fluences used for irradiating silicon, gallium-arsenide and silicon diodes

Sample (type)	Ions (type)	Energy (MeV)	Fluence, $\phi$ (cm <sup>-2</sup> s <sup>-1</sup> )
Crystalline silicon	O	30 to 80	$10^{11}$ to $10^{14}$
Silicon diodes	Li	30 to 35	$10^{10}$ to $10^{13}$
Silicon diodes, Si	B	40 to 65	$10^{10}$ to $10^{13}$
Silicon diodes	O	30 to 100	$10^{10}$ to $10^{13}$
GaAs	Li	30 to 50	$10^{12}$ to $10^{14}$
GaAs	O	30 to 50	$10^{12}$ to $10^{14}$
GaAs	Si	50	$10^{12}$ to $10^{14}$
Porous silicon	Si	10	$10^{13}$

These samples were also studied with the technique of optical radiation reflectivity using a polished aluminum sheet as a standard mirror. The reflectivity of 200 to 700 nm radiation from the surface of each ion irradiated silicon sample was measured. The reflectivity of the virgin sample was compared with that of ion-irradiated samples.

The turn-off time,  $t_{tr}$ , of each diode was measured by applying square wave pulses of frequency  $\sim 1$  kHz and of voltage  $\sim 20$  V (peak-to-peak) across the diode<sup>18</sup>. About 180 diode chips, with a rating of 1 kV, 1 A and a turn-off time of  $\sim 1000$  ns, were selected from the lot and arbitrarily divided into four groups. The diodes of the first group were irradiated with 35 MeV lithium ions, the second group with 65 MeV boron ions, and the third group with 100 MeV oxygen ions. The ion fluence was varied from diode to diode. The diodes of the fourth group were irradiated with lithium, boron, oxygen and silicon ions of different energies in the range 30 to 100 MeV. Thus distance  $D$  between the region of defects and the junction could be varied from 10  $\mu$ m to 140  $\mu$ m by irradiating diodes from the *n*-side with ions of different energy. The measured value of  $t_{tr}$  was correlated with the value of  $D$ , which was estimated from the TRIM program<sup>23</sup>.

The surface region of each ion-irradiated GaAs sample was studied with the thermally simulated exo-electron emission (TSEE) technique<sup>22</sup>, Photoluminescence, and X-ray fluorescence techniques. Similarly, another set of GaAs samples was irradiated with 30 MeV lithium ions and the surface of each lithium-irradiated GaAs sample was studied with the neutron depth profiling technique<sup>24,25</sup>. For each virgin and ion-irradiated GaAs sample the variation in the intensity of the exo-electrons with temperature over the range of 300 K to 550 K was studied.

Since the magnitude of the electron current was very low ( $< 10^{-13}$ ), the Faraday cup was replaced by a Channel Electron Multiplier and the output current was measured by an electrometer. A microprocessor-based system was used to control the heater current through which the sample temperature was maintained. During measurements, the sample temperature was increased at a rate of  $\sim 100$  K per minute. Typically the time required to measure this TSEE spectrum for a sample is about 1 h.

All those GaAs samples which were subjected to TSEE measurements (oxygen and silicon ion irradiated) were also studied with photoluminescence (PL) technique at room temperature. An argon-ion laser was used as a source of excitation radiation of wavelength  $\sim 488$  nm, and the photoluminescence spectrum was recorded over a wavelength region from 700 to 950 nm.

To find out if the buried silicon atoms from the irradiated region of the sample had diffused out and reached the surface region, these samples were also

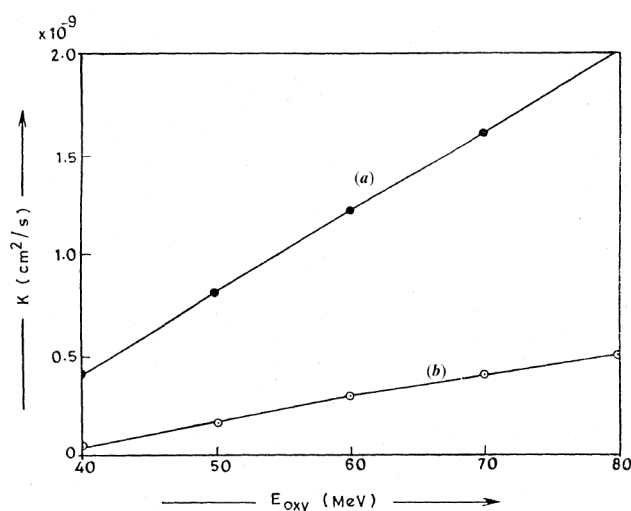
studied with the X-ray fluorescence technique using a RIGAKU-3070 spectrometer. The XRF spectrum was recorded by changing the detector angle ( $2\theta$ ) from  $30^\circ$  to  $120^\circ$ . In the XRF spectrum the peaks were identified and assigned to the corresponding elements such as As, Ga and silicon, etc.

The lithium ion-irradiated GaAs samples were also studied using the neutron depth profiling technique, as with this technique, concentrations of lithium as low as  $4 \times 10^{13}$  atoms/cm<sup>2</sup> can be measured<sup>24</sup>. These measurements were made using the Dhruva reactor at BARC, Mumbai. Thermal neutrons (flux  $\sim 10^6$  n cm<sup>-2</sup> s<sup>-1</sup>), were incident on the sample, mounted at  $45^\circ$  to the beam axis in a vacuum chamber. The intensity and energy of the emitted  $^4\text{He}_2$  and  $^3\text{H}_1$  particles were measured by a surface barrier detector. The details of the experimental setup and the method of analysis are given in our earlier publication<sup>25</sup>. The GaAs samples irradiated with 30 MeV lithium ions, at fluences  $10^{13}$  and  $10^{14}$  ions/cm<sup>2</sup>, were studied and the depth distribution of lithium was obtained.

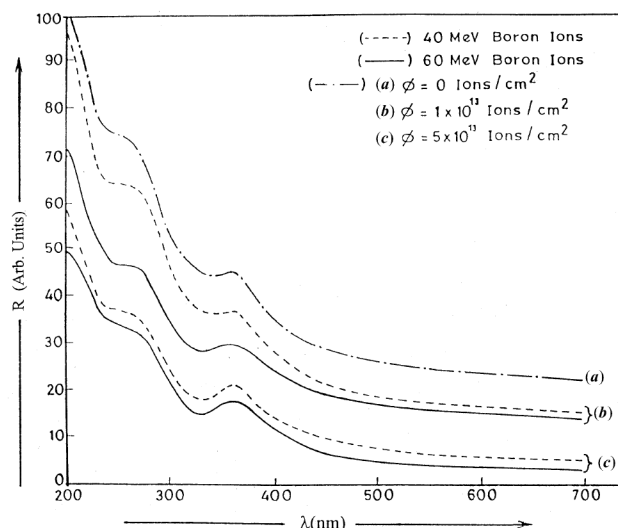
## 4. Results and discussions

### 4.1 Si

The variation in the damage coefficient,  $k$ , with the energy of oxygen ions (40 to 80 MeV) was studied for two fluences of  $10^{13}$  and  $10^{14}$  ions/cm<sup>2</sup>, and the results are shown in Figure 1. It is interesting to note that for any oxygen ion energy, the damage coefficient,  $k$  at a fluence of  $10^{13}$  ions/cm<sup>2</sup> is greater than that at a fluence of  $10^{14}$  ions/cm<sup>2</sup>. This effect may be attributed to the



**Figure 1.** In *n*-type silicon, variations in the damage coefficient,  $k$  with the energy of oxygen ions  $E_{\text{oxy}}$ : (a) fluence  $\sim 10^{13}$  ions/cm<sup>2</sup> and (b) fluence  $\sim 10^{14}$  ions/cm<sup>2</sup>.



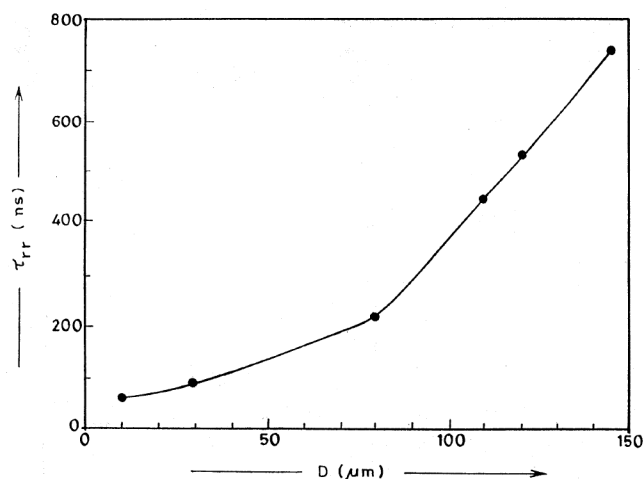
**Figure 2.** Variation in reflectivity,  $R$  with the wavelength,  $\lambda$  of the incident radiation, for crystalline silicon samples exposed to different fluences of 45 and 60 MeV boron ions.

annealing of defects through the energy deposited by the process of electronic energy loss,  $S_e$  in the surface region<sup>14,17</sup>. These results indicate that in the surface region, the number of defects annealed out increases with increasing ion fluence, provided the ion energy is kept constant.

The reflectivity of optical radiation in the range of 200 to 700 nm from the surface of each of the 40 and 60 MeV boron ion-irradiated silicon was measured and the results are shown in Figure 2. One can see from this figure that the reflectivity of optical radiation from a 60 MeV boron ion-irradiated sample is smaller than that for a 40 MeV boron ion-irradiated silicon sample. This indicates that the degree of the induced damage in the surface region of 60 MeV ion irradiated silicon is greater than that of 40 MeV ion irradiated silicon. However, for 60 MeV boron ions, both the nuclear energy loss and the electronic energy loss ( $S_e \sim 31.4$  eV/Å and  $S_n \sim 0.18$  eV/Å) are smaller than the respective values for 40 MeV boron ions ( $S_e \sim 41$  eV/Å and  $S_n \sim 0.26$  eV/Å). The relatively large value of  $S_e$  for 40 MeV boron ions may be responsible for annealing out a fraction of defects in the surface region of silicon<sup>14,17</sup>.

### 4.2 Si diodes

For irradiating diodes, mass number and energy of the ions were chosen in such a way that the ions could induce defects at different depths in the *n* region of diodes. Figure 3 shows that for a given ion fluence ( $10^{13}$  ions/cm<sup>2</sup>) the turn-off time,  $t_{\text{tr}}$ , decreases as the average distance,  $D$ , between the junction and the region of the induced defects is decreased. The distance,  $D$ , was varied by using lithium, boron and oxygen ions of different energies over the range 30 to 100 MeV.

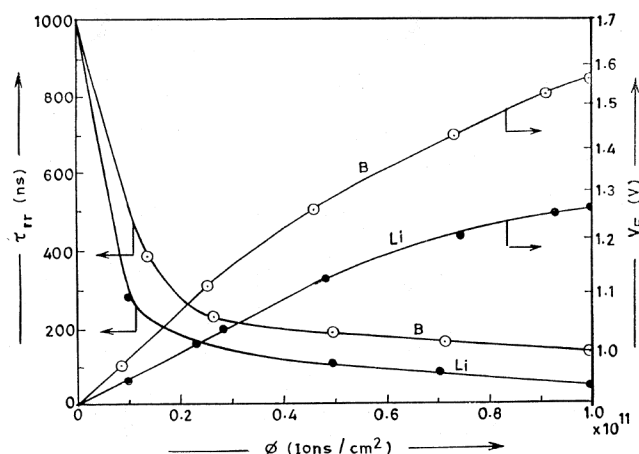


**Figure 3.** Variation in the turn-off time,  $t_{rr}$  with the distance,  $D$  between the region of the induced defects and the junction, in silicon diodes irradiated with different energy heavy ions at a fluence,  $\phi \sim 10^{13}$  ions/cm<sup>2</sup>.

It is seen from Figure 3 that by irradiating diode from the  $n$  side, though it is possible to tailor their turn-off time,  $\tau_{rr}$ , proper choice of ion mass number and energy is necessary for achieving optimum value of turn-off time,  $t_{rr}$ . Furthermore, these results show that a region containing a high concentration of defects (due to oxygen ions) produced away from the junction is less effective than a region with a relatively lower concentration of defects (due to lithium ions) located closer to the junction. These results have provided, for the first time, experimental evidence that in  $n$  region the concentration of the minority carriers decreases with the distance from the junction.

For 35 MeV lithium and 60 MeV boron ion-irradiated silicon diodes, Figure 4 shows variations in  $t_{rr}$  and  $V_F$  with the ion fluence,  $\phi$ . It can be seen that the decrease in  $t_{rr}$  is followed by an increase in the forward voltage drop,  $V_F$  across the diode. One can see also from Figure 4 that upon ion irradiation even though  $t_{rr}$  of the diodes decreases, the undesirable voltage drop,  $V_F$  across the diode increases. In the case of boron ion irradiated diodes, the rate of increase in the voltage drop with the ion fluence is greater than that in the lithium ion irradiated diodes. These results can be explained on the basis of the compensation effect. Boron atoms, being a  $p$ -type impurity, can compensate the effect of the  $n$ -type impurity in silicon. As a result, the resistivity of the local region in which compensation has occurred, increases with the fluence of boron ions. In addition, the resistivity of the  $n$  region also increases as the number of defects increases. On the contrary, the resistivity of the  $n$  region decreases as the number of lithium atoms increases, since lithium is an  $n$ -type impurity.

For the diodes used in the present work, the acceptable limit of the voltage drop is around 1.2 V and therefore  $t_{rr}$  could be brought down from 1000 ns to around

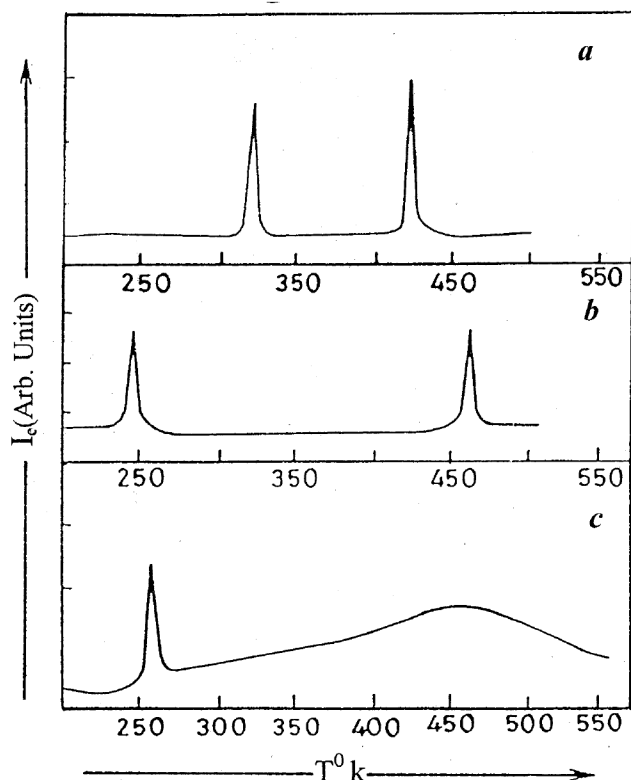


**Figure 4.** For the lithium and boron ion-irradiated diodes, the variations in the turn-off time,  $t_{rr}$  and voltage drop  $V_F$  with the fluence,  $\phi$  of 35 MeV lithium and 60 MeV boron ions.

100 ns with lithium ions only. Results of this study demonstrate that the switching characteristics of silicon diodes can be tailored by swift heavy ions.

#### 4.3 GaAs

Typical TSEE spectra, recorded from virgin GaAs and from silicon ion-irradiated GaAs are shown in Figure 5. Virgin GaAs exhibited [plot *a*] two peaks, one at 321 K and the other at  $\sim 420$  K. During the first TSEE run of the silicon ion-irradiated GaAs [plot *b*], the peak at 321 K of plot *a*, appeared at 246 K in plot *b*, whereas the peak at 420 K of plot *a* appeared at 463 K in plot *b*. The first and the second peaks in plot *b* correspond to the energy levels  $\sim (E_c - 0.53)$  eV and  $(E_c - 0.99)$  eV respectively. The TSEE spectrum of the silicon ion-irradiated GaAs, which had undergone one cycle of annealing and cooling, is as shown in plot *c* of Figure 5. By comparing the spectrum of plot *c* with that of the irradiated GaAs, plot *b*, it can be seen that peak at 463 K, in plot *c*, had broadened whereas the peak at 246 K shifted to  $\sim 260$  K with a marginal change in the intensity. The first peak in plot *b* corresponds to the oxygen-vacancy complex and the second peak is related to the silicon complex. After subsequent annealing and cooling cycles, the increased oxide concentration led to the shifting of the first peak from lower energy to higher energy. The shift in the peak position of the TSEE spectra is related to the new position of the defect level on the surface of GaAs. After ion-irradiation, the surface of GaAs developed defects related to the variations in the space charge concentration and bond angles, and, therefore, stresses are induced in the surface region. The energy levels of the defects get modified after each cycle of annealing due to variation in the stress

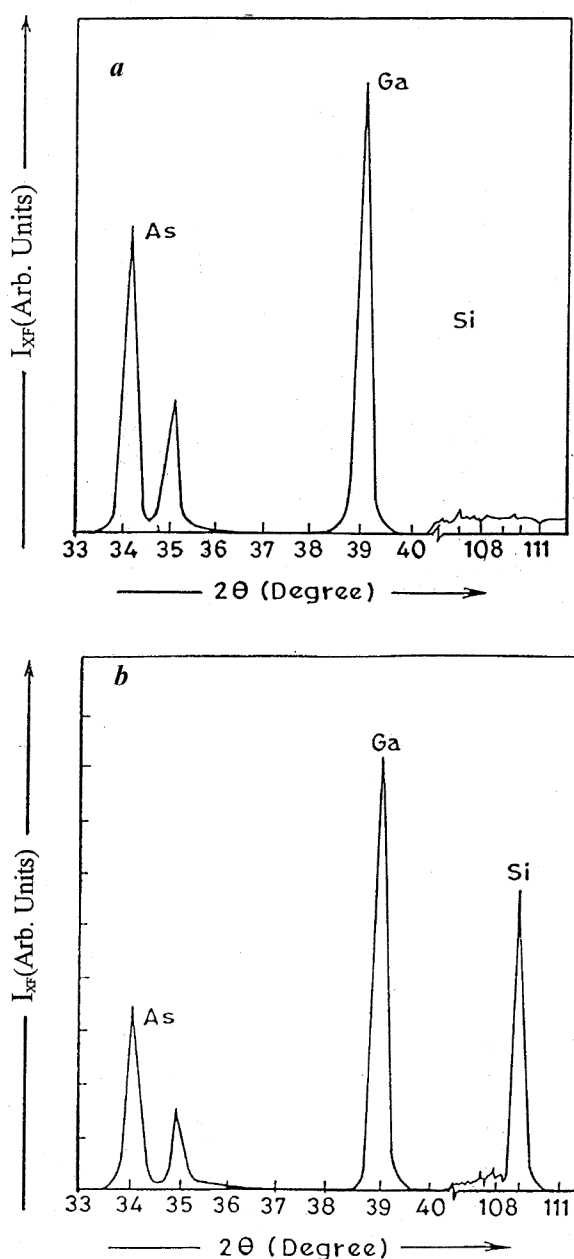


**Figure 5.** The variation in the exo-electron current,  $I_e$  with the sample temperature,  $T$  K, in the TSEE spectra of (a) Virgin GaAs (b) Si irradiated GaAs, the first TSEE run and (c) Si irradiated GaAs, the second TSEE run.

concentration around the defects and also due to the out-diffusion of silicon atoms, and subsequently the formation of silicon oxide in the surface region. The shift in the first peak from 246 K to 260 K can be attributed to the formation of  $\text{SiO}_2$  in the surface region and the second peak in plot *c* is due to anti-site defects. The broadening of the second peak in plot *c* indicates the existence of a group of defects related to silicon and silicon oxide. These results lead to the conclusion that the silicon atoms from the buried layer have diffused to the surface of GaAs. The diffusion of silicon atoms in GaAs at a relatively low temperature [ $\sim (300\text{--}550\text{ K})$ ] indicates that the diffusion coefficient of silicon in GaAs increased by at least 16 orders of magnitude due to the induced defects in GaAs.

The oxygen ion-irradiated GaAs samples were also subjected to TSEE measurements and these results have been reported earlier<sup>26</sup>. These results were similar to that of silicon ion-irradiated GaAs. After a number of annealing and cooling cycles, the implanted oxygen atoms gradually diffused out to the surface, forming an oxygen–vacancy complex. The intensity of the peak due to oxygen–vacancy complex was found to increase with increasing oxygen ion fluence.

These results reveal that both the silicon and oxygen atoms diffuse out from the buried layer to the surface



**Figure 6.** Variation in the XRF intensity,  $I_{XF}$  with the detector angle ( $2\theta$ ) in the XRF spectra of a virgin GaAs and of an annealed Si ion-irradiated GaAs. The peak at  $2\theta \sim 109^\circ$  indicates presence of silicon.

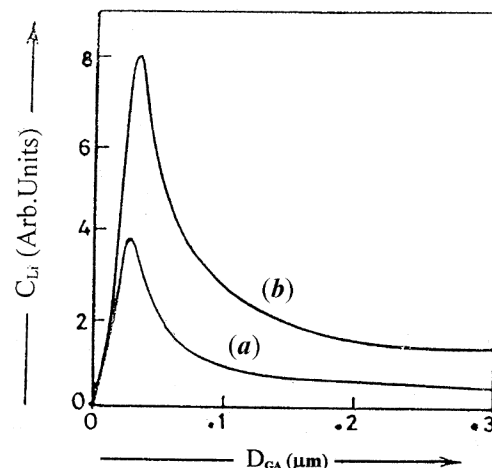
region. It appears that defects are produced throughout the ion trajectories, thus providing an easy path for the diffusion of silicon and oxygen atoms.

Typical X-ray fluorescence spectra recorded from an annealed virgin GaAs and a silicon ion irradiated-GaAs are shown in Figure 6. In this figure, only the XRF peaks corresponding to elemental As, Ga and Si are shown. The XRF peak at  $109^\circ$  ( $2\theta$ ), due to silicon atoms, was present in the spectrum of annealed silicon ion-irradiated GaAs but was absent in the annealed

GaAs sample. These results confirm that silicon atoms were present in the surface region of annealed silicon ion-irradiated GaAs sample.

In Figure 7, plot *a* shows the photoluminescence spectrum of a virgin GaAs sample, over the wavelength range 790 nm to 940 nm, with a peak at around 875 nm. Plot *b* is the PL spectrum of GaAs irradiated with silicon ions at a fluence of  $10^{14}$  ions/cm<sup>2</sup> but not annealed. Plot *c* shows the PL spectrum recorded from the GaAs sample which was irradiated with silicon ions at a fluence of  $10^{14}$  ions/cm<sup>2</sup>, and subjected to two runs of TSEE measurement. The results shown in Figure 7 indicate that due to the diffusion of silicon atoms from the buried layer, the GaAs surface was passivated and therefore the intensity of the PL had increased almost four-fold as compared to that for virgin GaAs. These results support the earlier findings that silicon atoms diffuse from the buried layer to the sample surface through the irradiated region. Furthermore, along with an enhancement in the intensity, the PL peak is also associated with a blue shift. The silicon atoms get oxidized on interaction with the oxygen atoms present in the surface region, and this leads to the formation of new interface states. A recombination of charge carriers through these surface states, related to the oxide layer, is responsible for the blue shift. The increase in the PL intensity is comparable to the enhancement in the PL intensity normally obtained after passivating a GaAs surface with silicon.

Figure 8 shows the depth distribution of lithium atoms in the GaAs samples irradiated with 30 MeV lithium ions at fluences  $10^{13}$  ions/cm<sup>2</sup> [plot *a*] and  $10^{14}$  ions/cm<sup>2</sup> [plot *b*]. The lithium depth distribution was measured by the neutron depth profiling technique. The measurements were carried out after a period of 90 days after irradiation and during this period the samples were kept at room temperature. From Figure 8

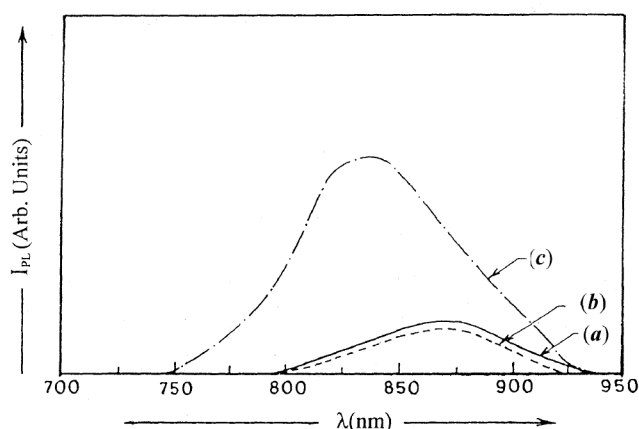


**Figure 8.** Variation in the concentration of lithium,  $C_{Li}$  with depth,  $D_{Ga}$  in GaAs samples irradiated with 30 eV lithium ions, (a)  $\phi = 10^{13}$  ions/cm<sup>2</sup> (b)  $\phi = 10^{14}$  ions/cm<sup>2</sup>.

one can see that the concentration of lithium atoms near the surface is at a maximum. The range of 30 MeV lithium ions in GaAs is  $\sim 88 \mu\text{m}$ , and therefore it is interesting to observe that the implanted lithium atoms had diffused out from the buried layer towards the surface. Furthermore, the number of lithium atoms that have diffused out in GaAs and migrated to the surface region varies with the fluence of lithium ions. Initially, the lithium concentration at the surface was small, at  $\sim 10^{11}$  atoms/cm<sup>3</sup>. However, after storing the GaAs samples for 90 days, the lithium concentration at the surface increased dramatically. A rough estimate indicates that at room temperature the diffusion coefficient of lithium in GaAs is  $\sim 5 \times 10^{-13} \text{ cm}^2 \text{ s}^{-1}$ . This large value for the diffusion coefficient is expected, due to the defects produced in the GaAs samples during ion-irradiation. These results further support the conclusion that defects are produced throughout the ion trajectories.

#### 4.4 Porous Si

As seen from plot *a* in Figure 9, the PL spectrum of porous silicon is broad and centered around 730 nm. Plot *b* in Figure 9 is the PL spectrum of the silicon ion irradiated porous silicon sample. In this spectrum a small blue shift in the peak position was observed, and the new peak appeared at 644 nm (1.83 eV). The 20-fold (approximate) increase in the PL intensity as compared to that of the unirradiated porous silicon samples, indicates that the surface region of porous silicon is altered on ion-irradiation. In this case the blue shift in the peak of the PL spectrum is attributed to the surface recrystallization, leading to formation of smaller size crystallites in the surface region<sup>27</sup>. The increased PL intensity remained stable with respect to time, and was studied over a period of four years.



**Figure 7.** Variations in the intensity of photoluminescence,  $I_{PL}$  with wavelength,  $\lambda$  in the photoluminescence spectra of (a) Virgin GaAs (b) Si irradiated GaAs (c) Annealed Si irradiated GaAs (undergone two TSEE runs).



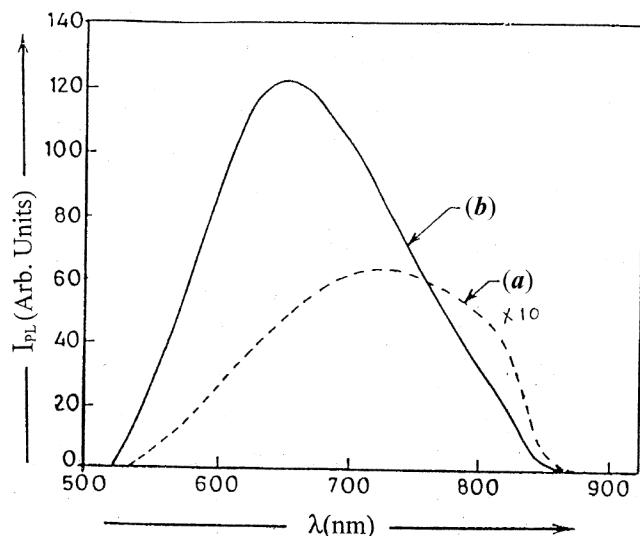


Figure 9. Variations in the intensity of photoluminescence,  $I_{PL}$  with the wavelength,  $\lambda$  in the photoluminescence spectra of porous silicon samples (a) virgin (b) silicon ion-irradiated.

## 5. Conclusions

Damage coefficient measurements and the reflectivity measurements indicate annealing of defects in Si due to high energy heavy ion irradiation. Enhancement in the PL intensity of porous silicon upon exposure to swift silicon ions has been observed. The migration of silicon, oxygen and lithium atoms from the buried layer to the GaAs surface suggests that the displacement-type defects must have been produced throughout the ion path. Analysis of the ion-irradiated silicon diodes provides direct experimental evidence that the concentration of minority carriers varies with the distance from the junction. Furthermore, the results indicate that low mass swift ions are very effective in tailoring the switching characteristics of silicon diodes.

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