

the radiative yield beyond about 10 J cm^{-2} could be attributed to the formation of nonradiative centers. However, increased surface recombination is thought to be a more likely explanation.

The absence of any apparent effect with the use of Q -switched 10^{-8} -s pulses shows the importance of choosing the appropriate combination of intensity and pulse width. Under surface melting conditions there is a substantial loss of As in the melted material,¹³ which could likely result in nonradiative As-vacancy complexes.

In summary, the feasibility of using a pulsed high-power laser to improve the radiative yield in GaAs has been demonstrated. Further experiments are in progress to understand more clearly these results.

We are grateful to Cominco Ltd/Trail for providing us with crystals together with a mass spectroscopic data. This study was supported by NSERC grant No. 67-7804.

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Photoluminescence from hydrogenated ion-implanted crystalline silicon^{a)}

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(Received 20 August 1979; accepted for publication 17 September 1979)

Ion-implanted crystalline silicon annealed in atomic hydrogen photoluminesces at $0.99 \pm 0.01 \text{ eV}$ in a band having a spectral width of 0.1 eV FWHM. The emitted spectrum does not depend on the chemical nature of the implanted ion (Al, As, D, F, H, Ne, P, Si); however, when the surface is amorphized, a spectrum characteristic of hydrogenated amorphous Si is obtained.

PACS numbers: 78.55. — m, 78.55.Ds, 71.55. — i, 81.40.Tv

We have discovered in crystalline silicon new luminescence centers that are due to ion-implantation damage.

We were attempting to produce a layer of hydrogenated amorphous silicon, $a\text{-Si:H}$, on the surface of crystalline Si, $c\text{-Si}$, by a new method that differs from glow-discharge decomposition of silane,¹ sputtering Si in hydrogen ambient,² low-temperature pyrolysis of silane,³ or hydrogenation of evaporated amorphous Si.⁴ The intent of our method was to amorphize the surface layer of $c\text{-Si}$ by ion implantation.⁵ The implantation was done at room temperature using $^{28}\text{Si}^+$ ions in the 40–280 keV range. Then, the damaged layer was hydrogenated by exposure to atomic hydrogen at 300 °C, as described earlier.⁶

The photoluminescence spectrum of the sample was measured at liquid-nitrogen temperature. The luminescence was excited by a chopped 488-nm beam of an Ar laser and detected by a cooled PbS photoconductor followed by lock-in amplification. A typical emission spectrum is shown in Fig. 1 together with emission spectra from $c\text{-Si}$ and $a\text{-Si:H}$. One finds that the new emission peaks at $0.99 \pm 0.01 \text{ eV}$ and has a spectral width of 0.1 eV (FWHM). The highest external efficiency that we have measured at 78 °K is about 1%. The identical spectrum is obtained at all implantation doses up to 10^{14} cm^{-2} , and does not appear to depend on the chemical nature of the ion. Similar spectra were obtained with Al, H, D, P, As, F, or Ne implantation into silicon.

The dependence of the luminescence intensity on the implantation dose is illustrated in Fig. 2. For this run a $10\text{-}\Omega \text{ cm } n\text{-type}$ (111) wafer was implanted at 30 and 60 keV to

^{a)}Research reported herein was supported by the Department of Energy, Division of Solar Technology, under Contract No. EY-76-C-03-1286 and by RCA Laboratories, Princeton, NJ 08540.

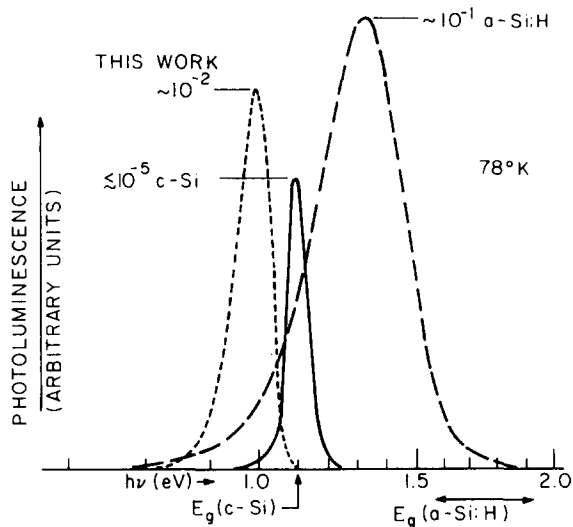


FIG. 1. Emission spectra at 78 °K from hydrogenated ion-implanted Si (this work), from crystalline Si (c-Si), and from hydrogenated amorphous Si (a-Si:H). E_g is the energy gap. The numbers labeling the peaks are the external emission efficiencies.

obtain an approximately uniform depth profile over 10^{-5} cm (within the absorption depth of the exciting light). One sees that the luminescence efficiency rises gradually for doses up to 10^{13} cm $^{-2}$ and then remains nearly constant up to 10^{14} cm $^{-2}$, whereupon the efficiency drops abruptly. Note that below a dose of $\sim 1 \times 10^{14}$ cm $^{-2}$, the material is still a single crystal. At a higher dose rate, 2×10^{14} cm $^{-2}$, a broad emission peak at 1.15 eV is obtained, corresponding to the luminescence of a-Si:H. The gradual rise in efficiency with implantation dose suggests an increase in the number of luminescence centers. The saturation of efficiency implies that these centers are so efficient that a concentration of 10^{13} cm $^{-2}/10^{-5}$ cm = 10^{18} cm $^{-3}$ is sufficient to recombine all the excess carriers. Adding more centers beyond this critical concentration of 10^{18} cm $^{-3}$ cannot increase the efficiency. Above 10^{19} cm $^{-3}$, however, concentration quenching sets in. At this higher density, the defects are on the average 46 Å apart and thus may form centers for nonradiative recombination that compete efficiently with the radiative process.

Although ion implantation of c-Si is now a widely used process for commercial devices, to the best of our knowledge a somewhat similar luminescence has been mentioned only briefly for Ar-implanted Si.⁷ The reason that this center has escaped notice is probably due to the high temperature at which the implantation damage is usually annealed (> 900 °C). There is an extensive body of literature, however, on the photoluminescence of c-Si irradiated by high-energy electrons and annealed at room temperature.⁸ In this case, an entirely different spectrum is observed at low temperatures consisting of emission lines at 0.97 and 0.79 eV; furthermore, these lines are extremely narrow (≤ 5 meV FWHM).

We believe that the luminescence at 0.99 eV is due to a structural damage induced by ion implantation rather than to a chemical doping effect, since the spectrum does not depend on the chemical species of the ion. These centers may be similar to the vacancies induced by 3-MeV electron-beam

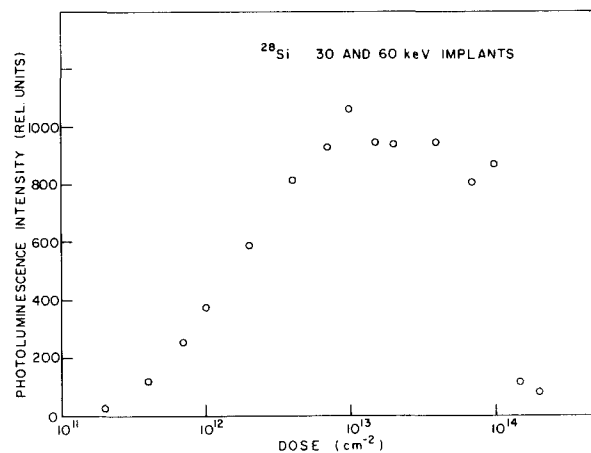


FIG. 2. Dependence of the luminescence intensity at 78 °K on the $^{28}\text{Si}^+$ implantation dose.

irradiation, as reported by Troxell and Watkins⁹ who find donorlike and acceptorlike levels ~ 0.1 eV from the band edges.

The temperature dependence of luminescence from the sample irradiated at 1×10^{13} cm $^{-2}$ with $^{28}\text{Si}^+$ indicates, above ~ 110 °K, an activation energy of 90 meV for the competing nonradiative recombination process—this competing process may be the thermal dissociation of geminate pairs or bound excitons at donorlike or acceptorlike centers. The 0.09-eV value of activation energy is consistent with the results of Troxell and Watkins.⁹ On the other hand, the temperature dependence of luminescence from the amorphized sample (irradiated at 7×10^{14} cm $^{-2}$ with $^{28}\text{Si}^+$) exhibits an activation energy of 0.15 eV for the nonradiative recombina-

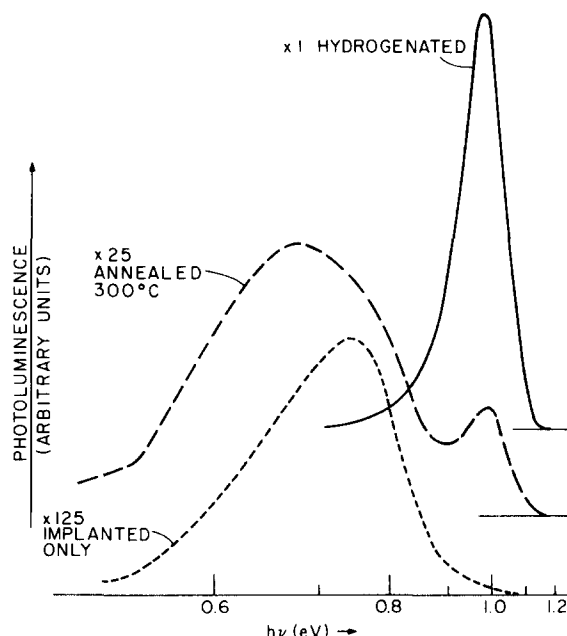


FIG. 3. Emission spectra at 78 °K of ion-implanted Si after implantation, after vacuum annealing at 300 °C for $\frac{1}{2}$ h, and after annealing in atomic hydrogen at 300 °C for $\frac{1}{2}$ h.

tion (compared to the value of 0.13 eV obtained for glow-discharge-deposited *a*-Si : H).¹⁰

Samples of ion-implanted *c*-Si that have not been annealed in atomic hydrogen exhibit a weak broad emission peaking at ~ 0.7 eV (see Fig. 3). Vacuum annealing at 300 °C for 30 min causes the ~ 0.7 -eV peak to grow by a factor of 5 and a contribution at 0.99 eV to appear. Annealing in atomic hydrogen at 300 °C for 30 min greatly enhances the 0.99-eV peak and quenches the 0.7-eV emission.

Ion implantation generates many dangling bonds that form centers for nonradiative recombination. These centers decrease the carrier lifetime and compete effectively with radiative transitions. However, after hydrogenation, since hydrogen ties dangling bonds, the luminescence process becomes more efficient. Furthermore, since Fig. 3 shows that the 0.99-eV emission is obtained even before hydrogen is introduced, the new radiative center may be formed due to residual hydrogen in the *c*-Si that combines with the implantation-induced defect.

In conclusion, ion implantation at doses $< 10^{14} \text{ cm}^{-2}$ introduces in *c*-Si an ion-independent defect that after hydrogenation emits relatively efficiently a characteristic luminescence at 0.99 eV. More work is needed to identify the exact nature of this radiative recombination center.

We are grateful to J.E. Berkeyheiser and F. Kolondra for technical assistance.

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Observation of semiconductor-semimetal transition in InAs-GaSb superlattices^{a)}

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(Received 21 August 1979; accepted for publication 10 September 1979)

The semiconductor-semimetal transition in InAs-GaSb superlattices is observed at a layer thickness in the vicinity of 100 Å. The transition manifests itself in an increase in the measured carrier concentration as a result of electron transfer from GaSb to InAs when ground subbands of electrons and heavy holes cross each other. Shubnikov-de Haas measurements confirm the carrier enhancement in the semimetallic state.

PACS numbers: 64.70. — p, 05.70.Fh, 81.30.Dz

At the interface of an InAs-GaSb superlattice, a periodic structure made of alternating layers of the two semiconductors, the conduction band edge of the former lies below the valence band edge of the latter on the electron energy scale.¹ The most significant consequence of this unusual relationship of band-edge separation is the possibility of achieving a semiconductor-to-semimetal transition.² As the layer thickness is increased, quantum states or subbands associated with electrons can be lowered in energy from above to below those associated with heavy holes, resulting in an electron transfer to InAs leaving behind holes in GaSb. We re-

port the experimental observation of this transition in this letter.

Previous experiments aimed at exploring this feature of band-edge separation have led to interesting observations of Ohmic transport behavior in single heterojunctions and resonant-tunneling characteristics in double barriers.^{3,4} The formation of the superlattice has been established from Shubnikov-de Haas measurements which elucidate the quasi two-dimensional aspect of subbands,⁵ and from optical absorption measurements which give the subband energies.⁶ This series of experiments has aroused great interest in the InAs-GaSb system and motivated various theoretical considerations of the electronic structure of this type of superlattice. Calculations by use of linear combinations of atomic orbitals⁷ yielded essentially the same results as obtained originally by wave function matching.¹ Recently, three-dimen-

^{a)}Sponsored in part under ARO Contract.

^{b)}Part of this work was performed while the authors were guest scientists at the Francis Bitter National Magnet Laboratory, which is supported at Massachusetts Institute of Technology by the National Science Foundation.