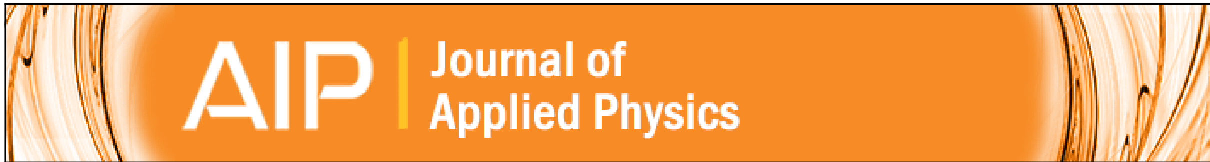


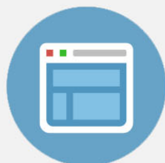
Enhanced electrical activation of indium coimplanted with carbon in a silicon substrate

H. Boudinov, J. P. de Souza, and C. K. Saul

Citation: Journal of Applied Physics **86**, 5909 (1999); doi: 10.1063/1.371611
View online: http://dx.doi.org/10.1063/1.371611
View Table of Contents: http://scitation.aip.org/content/aip/journal/jap/86/10?ver=pdfcov
Published by the AIP Publishing

Re-register for Table of Content Alerts

Create a profile.



Sign up today!

[This article is copyrighted as indicated in the article. Reuse of AIP content is subject to the terms at: http://scitation.aip.org/termsconditions. Downloaded to] IP: 200.130.19.173 On: Wed, 19 Mar 2014 18:27:12

Enhanced electrical activation of indium coimplanted with carbon in a silicon substrate

H. Boudinov, J. P. de Souza,[a] and C. K. Saul[b]
Instituto de Física, Universidade Federal do Rio Grande do Sul, 91501-970 Porto Alegre, R.S., Brazil

(Received 14 April 1999; accepted for publication 18 August 1999)

Electrical activation of In of 18%–52% of the implanted dose ($5\times10^{14}\,\mathrm{cm}^{-2}$) was obtained in Si samples having a C^+ coimplantation after rapid thermal annealing (RTA) at 800–1000 °C for 15 s. This electrical activation yield markedly contrasts with that in samples singly implanted with In in which only ≅0.5% of the dose was activated. The following features were observed in the coimplanted samples: (i) a reverse annealing of the electrical activation in the temperature range of 800–900 °C; (ii) significant reduction of the In profile redistribution during RTA; and (iii) the electrically activated In concentration is substantially higher than the substitutional In concentration. These findings are discussed in terms of the interaction between C atoms and Si self-interstitials (Si_I), strain compensation between C and In atoms in the Si lattice, and formation of stable In substitutional–C substitutional acceptors centers. © *1999 American Institute of Physics.* [S0021-8979(99)06622-0]

The shrinkage of the device geometry to increase the packing level in integrated circuits (ICs) demands reduction of the junction depths according to the scaling rules.[1] Extensive studies of 0.1 μm metal–oxide–semiconductor field-effect transistors (MOSFETs) have been carried out[2] to test their potential applications in ultralarge scale integrated (ULSI) circuits. For a channel length less than 0.1 μm, the source and drain junctions need to be as shallow as 30–40 nm.

Nowadays, a great deal of effort is undertaken in the development of very shallow junction technology. Shallow n^+ doped layers can be conveniently obtained by low energy implantation of group V heavy ions, like As or Sb. However, in the case of p-type dopants a similar approach does not apply, since the column III heavy ions have very low solid solubility limits in silicon crystals.[3] For example, the solid solubility limits for Ga[3] and In[4] at 1100 °C are only 4×10^{19} and $0.4–2\times10^{17}\,\mathrm{cm}^{-3}$. Indium is known to yield a very poor electrical activation after conventional or rapid thermal annealing (RTA). Values as low as 1% were reported[5,6] for implanted doses of $0.5–5\times10^{15}\,\mathrm{cm}^{-2}$. The most convenient p-type dopant is boron. However, shallow boron profiles can be realized only if the channeling[7] and the transient enhanced diffusion (TED)[8] are minimized. These requirements are best accomplished when a preamorphization implant with heavier column IV element ions (Si^+, Ge^+, or Sn^+) is performed prior to B^+ or BF_2^+ implantation.[9]

In recent years considerable attention has been devoted to the behavior of carbon in silicon[10] because it affects various phenomena, such as defect formation kinetics,[8,11] impurity gettering,[12] dopant activation,[13] etc. It is assumed that carbon atoms trap self-interstitial Si atoms (Si_I).[11] The reduction of the Si_I concentration influences the electrical activation and the diffusion of the coimplanted dopant atoms. In the particular case of In doping, an acceptor level related to the presence of carbon was found, with an ionization energy of 0.111 eV. This energy is significantly lower than that of substitutional In in a Si matrix (0.156 eV).[14] Baron *et al.*[15] attributed this acceptor level to a substitutional carbon–substitutional indium (C_S-In_S) pair defect.

In the present study the annealing behavior of samples single implanted with In^+ and dually implanted with In^+ and C^+ are compared. It is demonstrated that a pronounced electrical activation can be obtained when In^+ is coimplanted with C^+.

Silicon wafers of (100) orientation of n-type and resistivity of 4–11 Ω cm were used in the experiments. Square Van der Pauw devices[16] of 6 mm×6 mm were fabricated on the wafers labeled I and II, using a conventional photolitography technique. The p^+ ohmic contacts were prepared with B^+ implantation ($2\times10^{15}\,\mathrm{cm}^{-2}$, 50 keV) followed by drive-in in an oxidizing ambient. Circle windows of 3.0 mm diameter centered in the squares were opened in the SiO_2 layer. $^{12}C^+$ was implanted in wafer II with a dose of $4\times10^{15}\,\mathrm{cm}^{-2}$ and energy of 50 keV. Subsequently, $^{115}In^+$ was implanted in wafers I and II, with a dose of $5\times10^{14}\,\mathrm{cm}^{-2}$ at 50 keV. Both implantations were performed at room temperature (RT). The mean projected range and standard projected range deviation obtained from TRIM[17] simulation are, respectively, 32 and 9 nm for In and 148 and 47 nm for C. The C profile was made deeper than that of In such that the C concentration over the In profile regions is only around 10%–20% of the In peak concentration. The predicted[17] C peak concentration is $3.5\times10^{20}\,\mathrm{cm}^{-3}$.

The implanted devices were individually annealed for 15 s in the temperature range of 800–1000 °C in argon atmosphere using a halogen lamp RTA system. Sheet resistivity and Hall effect measurements were performed in the Van der Pauw devices at RT. Carrier concentrations were determined using a Hall scattering factor of unity ($r=1$).

[a] Electronic mail: souza@if.ufrgs.br
[b] Present address: Departamento de Fisica, UFP, 81531-970, Curitiba, PR, Brazil.

 © 1999 American Institute of Physics

[This article is copyrighted as indicated in the article. Reuse of AIP content is subject to the terms at: http://scitation.aip.org/termsconditions. Downloaded to] IP: 200.130.19.173 On: Wed, 19 Mar 2014 18:27:12

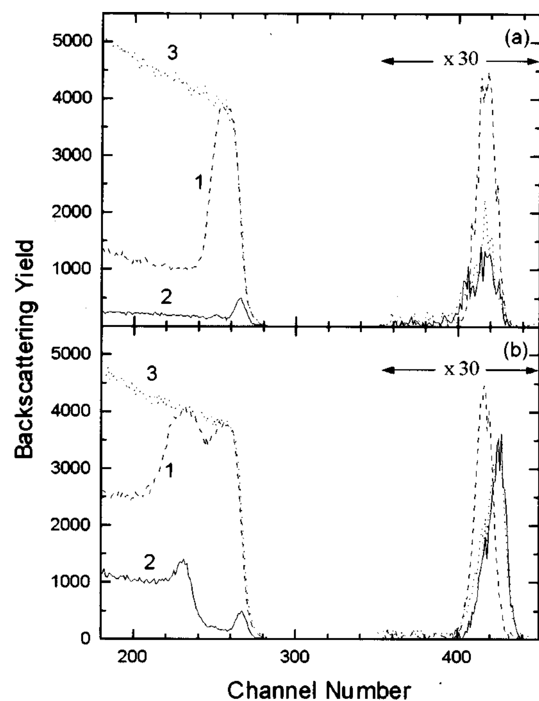


FIG. 1. RBS spectra of sample I (a) and sample II (b) in ⟨100⟩ aligned crystal direction taken before RTA (curves 1) or after RTA (1000 °C/15 s) (curves 2), and in random direction after the same RTA cycle (curves 3). Sample I was single implanted with In^+ ($5\times10^{14}\,cm^{-2}$ at 50 keV). Sample II was dually implanted with C^+ ($4\times10^{15}\,cm^{-2}$ at 50 keV) and In^+ ($5\times10^{14}\,cm^{-2}$ at 50 keV).

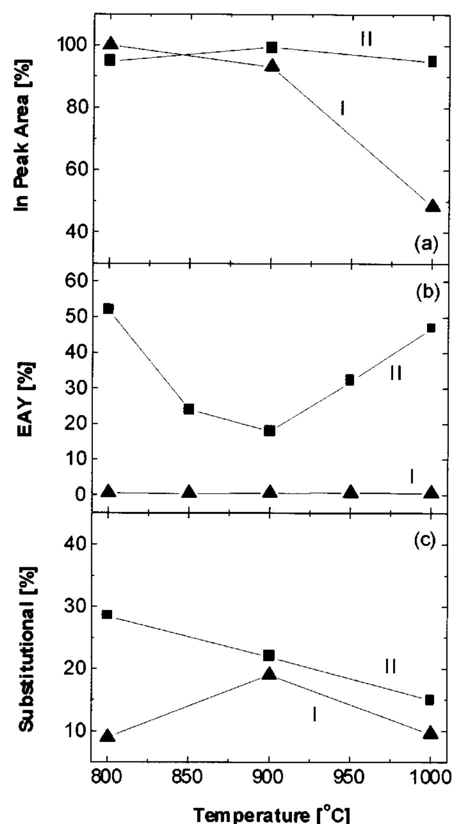


FIG. 2. Remaining fraction of In (a), electrical activation yield after RTA for 15 s (b) and substitutional fraction of In (c) in samples I (single implanted with In^+) and II (dually implanted with C^+ and In^+).

The In depth profiles were analyzed by Rutherford backscattering spectrometry employing a $900\,keV\,He^{++}$ beam aligned with the ⟨100⟩ crystal direction (RBS/C) or in a random incidence direction (RBS). The overall energy resolution of the RBS analysis was 14 keV.

Figures 1(a) and 1(b) show the RBS spectra of samples from wafers I and II, respectively, before and after RTA at 1000 °C. For better visualization of the In backscattering data, after channel 350 the vertical scales were magnified 30 times.

The RBS/C spectrum of an as implanted sample I [curve 1 in Fig. 1(a)] demonstrates that a surface amorphous layer of 56 nm thickness was formed by the In^+ implantation. The calculated In peak concentration is $2.1\times10^{20}\,cm^{-3}$. In Fig. 1(b) the corresponding spectrum for an as implanted sample II indicates the presence of a second amorphous layer beneath the first one. This buried layer was formed by the C^+ implantation and extends to a depth of 147 nm.

The analysis of the channeled [curve 2 in Fig. 1(a)] and random [curve 3 in Fig. 1(a)] spectra taken from sample I after RTA at 1000 °C, discloses the following features: (i) complete recrystallization of the surface amorphous layer with a residual defect density below the detection limit of the RBS technique, (ii) 50% of the In dose has diffused inward as denoted by the reduction of the area under the In backscattering data points, and (iii) a low substitutional fraction (10%) of the remaining In concentration. We have performed integration of the indium peak area in each spectrum. Considering the area of the as-implanted peak as 100% we estimated the substitutional and nonsubstitutional fractions. The difference between the random and channeled spectra fractions gives us an evaluation of the substitutionality, since the substitutional In atoms do not contribute to the backscattered yield in the channeled mode. The in-diffusion of In is also confirmed in spectra of samples annealed at lower temperatures (not shown).

Channeled [curve 2 in Fig. 1(b)] and random [curve 3 in Fig. 1(b)] spectra, regarding sample II after RTA at 1000 °C discloses the following peculiarities: (i) the recrystallization of the surface amorphous layer lead to a residual defect density below the detection limit while that of the buried layer resulted in a defect density of around 20% at the depth of 105 nm, (ii) a reduced loss of implanted In (5%), (iii) movement of the In profile toward the surface, and (iv) low substitutional fraction (15%) of the remaining In concentration. The corresponding spectra taken in samples II annealed at lower temperatures (not shown), also do not evidence any signature of in-diffusion of In.

Figure 2(a) presents the remaining fraction of In in samples I and II after the annealing cycles, evaluated from the RBS and RBS/C spectra. A higher retention of the In during the annealing is clearly apparent in the curve corresponding to sample II in Fig. 2(a). The lack of in-diffusion and hence the higher In retention in the coimplanted samples, very likely results from the trapping of Si_I by the C atoms. In the absence of the C in the single implanted samples, the Si interstitials are probably more mobile and promote enhanced inward diffusion of In.

[This article is copyrighted as indicated in the article. Reuse of AIP content is subject to the terms at: http://scitation.aip.org/termsconditions. Downloaded to] IP: 200.130.19.173 On: Wed, 19 Mar 2014 18:27:12

The electrical activation yield (EAY) of In evaluated from electrical measurements in the Van der Pauw devices from samples I and II are presented in Fig. 2(b). The enhancement of the EAY caused by the C^+ coimplantation is remarkable. While the EAY values in single implanted devices are ≅0.5% irrespectively of the RTA temperature, those in the coimplanted sample are in the range of 18%–52%. It is interesting to note a reverse annealing of the EAY in the coimplanted devices over the temperature range of 800–900 °C. The causes of this reverse annealing are unclear at present. The Hall mobility values varied in the range of 14–83 $cm^2/(V\,s)$ in the coimplanted samples.

Figure 2(c) presents the substitutional fraction of the In concentration after the RTA cycles. This fraction in devices of sample II decreases linearly from 28% at 800 °C to 10% at 1000 °C. Comparing EAY values from Fig. 2(b) and substitutional fraction of In concentration from Fig. 2(c), concerning devices of sample II, one observes that the EAY is always higher than the substitutional fraction. This indicates that most of the electrically activated In atoms reside out of regular substitutional lattice sites. Very likely, they are forming the In–C acceptor centers, consisting of the In atoms, and probably the C atoms, located on near substitutional lattice sites. The formation of this defect pair should be energetically favored, since the stresses introduced by In (covalent radius=0.144 nm) and C (covalent radius=0.077 nm) in the Si lattice (covalent radius of Si=0.117 nm) are alleviated in the In–C pair structure. More investigation needs to be undertaken to corroborate this hypothesis.

The substitutional fraction of In in sample I varied from 9% to 18%. These values are substantially higher than the EAY of ≅0.5%. Consequently, most of the substitutional In atoms are probably complexed with point defects and hence deactivated. Very likely, vacancies should interact with substitutional In atoms to relax the strain field introduced by the larger covalent radius of In (0.144 nm) compared to that of Si (0.117 nm).

In summary, the behavior of samples implanted only with In or dually with In and C during RTA was compared. The observed very poor activation of In in single implanted samples (≅0.5%) agrees quite well with previous published works.[5,6] In addition, a large reduction of the In concentration resulted from a fast inward diffusion by a significant fraction of the implanted dose. This diffusion behavior is suppressed in the C^+ coimplanted samples, in which practically the total implanted dose is retained near the surface. Furthermore, a reduced profile redistribution was observed.

The EAY of In in the C^+ coimplanted samples is enhanced up to two orders of magnitude, compared to that in single implanted samples. RBS/C analysis of coimplanted samples provided evidence that most of the electrically activated In atoms are not on regular substitutional lattice sites. The activation of nonsubstitutional In atoms probably results from the formation of acceptor In–C defects, with In and C atoms residing on near substitutional lattice sites. A reverse annealing of the In activation was disclosed in the temperature range of 800–900 °C.

This work was partially supported by Conselho Nacional de Pesquisas (CNPq) and Financiadora de Estudos e Projetos (FINEP).

[1] K. N. Kwok, S. A. Eshraghi, and T. D. Stanik, IEEE Trans. Electron Devices **ED-40**, 1895 (1993).
[2] M. Saito, T. Yoshitomi, H. Hara, M. Onu, Y. Akasaka, H. Nii, S. Matsuda, H. S. Momose, Y. Katsumata, Y. Ushiku, and H. Iwai, IEEE Trans. Electron Devices **ED-40**, 2264 (1993).
[3] F. A. Trumbore, Bell Syst. Tech. J. **39**, 205 (1960).
[4] A. Mühlbauer, in *Numerical Data and Functional Relationships in Science and Technology*, edited by O. Madelung, M. Schulz, and H. Weiss (Springer, Berlin, 1984), Vol. 17, Chap. 6.
[5] R. D. Pashley, Radiat. Eff. **11**, 1 (1971).
[6] E. C. Jones and E. Ishida, Mater. Sci. Eng., R. **24**, 1 (1998).
[7] M. C. Ozturc, J. J. Wortman, C. M. Osburn, A. Ajmera, E. Frey, G. A. Rozgonyi, W. Chu, and C. Lee, IEEE Trans. Electron Devices **ED-35**, 659 (1988).
[8] S. Nishikawa, A. Tanaka, and T. Yamaji, Appl. Phys. Lett. **60**, 2270 (1992).
[9] J. P. de Souza and D. K. Sadana, in *Handbook of Semiconductors*, edited by T. S. Moss (Elsevier, Amsterdam, 1994), Vol. 3b, Chap. 27.
[10] W. Skorupa and R. A. Yankov, Mater. Chem. Phys. **44**, 101 (1996).
[11] M. Tamura, T. Ando, and K. Ohyu, Nucl. Instrum. Methods Phys. Res. B **59**, 572 (1991).
[12] H. Wong, N. W. Cheung, P. K. Chu, J. Liu, and J. W. Mayer, Appl. Phys. Lett. **52**, 1023 (1988).
[13] J. P. de Souza and H. Boudinov, J. Appl. Phys. **74**, 6599 (1993).
[14] R. Baron, M. H. Young, J. K. Neeland, and O. J. Marsh, Appl. Phys. Lett. **30**, 594 (1977).
[15] R. Baron, J. P. Baukus, S. D. Allen, T. C. McGill, M. H. Young, H. Kimura, H. V. Winston, and O. J. Marsh, Appl. Phys. Lett. **34**, 257 (1979).
[16] L. J. Van der Pauw, Philips Res. Rep. **13**, 1 (1958).
[17] J. F. Ziegler, J. P. Biersack, and U. Littmark, in *The Stopping and Range of Ions in Solids* (Pergamon, New York, 1985).

[This article is copyrighted as indicated in the article. Reuse of AIP content is subject to the terms at: http://scitation.aip.org/termsconditions. Downloaded to] IP: 200.130.19.173 On: Wed, 19 Mar 2014 18:27:12