MRS Advances © 2017 Materials Research Society
DOI: 10.1557/adv.2017.314

Low Energy Ion Implantation and Annealing of Au/Ni/Ti Contacts to n-SiC

Neelu Shrestha[1], Martyn H. Kibel[2], Patrick W Leech[1], Anthony S Holland[1], Geoffrey K Reeves[1], Mark C Ridgway[3], Phillip Tanner[4]

[1] School of Engineering, RMIT University, Melbourne, Victoria, 3001, Australia
[2] Centre for Materials and Surface Science, Department of Chemistry and Physics, La Trobe University, Melbourne, Victoria, Australia.
[3] Electronic Materials Engineering, Australian National University, Canberra, ACT, Australia.
[4] Griffith University, Queensland Microtechnology Facility, Brisbane, Australia.

ABSTRACT

The electrical characteristics of Au/Ni/Ti/ n-SiC contacts have been examined as a function of implant dose (10^{13}-10^{14} ions/cm^2) at 5 KeV and temperature of annealing (750-1000 °C). Measurements of specific contact resistance, ρ_c, were approximately constant at lower implant doses until increasing at 1 x 10^{15} ions/cm^2 for both C and P ions. Annealing at a temperature of 1000 °C has reduced the value of ρ_c by an order of magnitude to ~1 x 10^{-6} Ω.cm^2 at implant doses of 10^{13}-10^{14} ions/cm^2. Auger Electron Spectroscopy (AES) has shown that annealing at 1000 °C resulted in a strong indiffusion of the metallization layers at the interface.

INTRODUCTION

SiC has become an important semiconductor in devices operating at high power and elevated temperatures. A critical requirement in the realization of these devices has been the formation of ohmic contacts with low specific contact resistance, ρ_c, and strong mechanical adhesion at the metal/ SiC interface. For n-SiC, the most commonly used ohmic contact has been the metal Ni for which annealing at 900-1000 °C was required in order to obtain a minimum in ρ_c [1]. However, the formation of Ni_2Si phase during annealing at 900-1000 °C has been accompanied by a segregation of carbon, the generation of interfacial voids and a roughening of the surface morphology of the metal [2]. These defects have acted to degrade the mechanical adhesion of the metal/n-SiC interface. To reduce the presence of excess carbon, the carbide forming element Ti has been incorporated into Ti/Ni multilayered contacts [3-5]. The reduction in excess carbon through the presence of TiC phase during annealing has resulted in an increased mechanical adhesion in comparison with Ni films [4]. In addition, the formation of $TiSi_2$ at the interface during annealing of the contacts at $\geq$700 °C has resulted in a minimum in ρ_c [6].

An alternative method of forming ohmic contacts to n-SiC as reported by Grodzicki et al. has been the pre-treatment of the substrate with low energy ions [7]. In Cr/ n-SiC contacts, the bombardment of n-SiC with low energy Ar ions has resulted in a transition from a non-linear to a linear current/ voltage response [7]. In another study, the implantation of n-SiC with P or C ions at low energy has resulted in an increase in ρ_c in as-deposited contacts of Ti/Ni/Au [8]. The bombardment of polycrystalline layers of n-SiC with low energy Ar ions has shown a reduction in the surface energy and internal stress in the layers [9]. In this paper, we have examined the combined effect of annealing of Ti/Ni contacts to n-SiC and a pre-implantation using either C or P ions at low energy. The properties of these contacts have been investigated over a range of dose (10^{13}-10^{15} ions/cm^2) and annealing temperature (750-1000 °C).

EXPERIMENTAL DETAILS

The films used in this study were comprised of a 1.0 µm thick layer of n-type 3C-SiC (R_s = 0.001 Ω.cm) deposited on wafers of p-Si <100>. The SiC was grown epitaxially by low pressure chemical vapor deposition (LPCVD) in a hot wall reactor [10]. The wafers with a diameter of 10 cm were diced into square pieces of 1 x 1 cm. The individual pieces of wafer were implanted with either C or P ions at -196 °C using doses in the range 10^{13}-10^{15} ions/cm^2 at 5 keV. On the surface of the implanted wafer was then fabricated an array of two-contact circular transmission line model patterns [11]. In the pattern, the gap spacing between the inner circular and outer ring electrodes was 1.5, 4.5 and 7.5 µm. The patterned samples were dipped in buffered 10% HF for 15 s to remove native oxide from the exposed SiC followed by a rinse in de-ionised water and drying in high purity nitrogen. Layers of Ti (50 nm)/ Ni (50 nm)/ Au (50 nm) were then sequentially sputtered onto the n-SiC layer followed by a process of "lift-off". The samples were Rapid Thermal Annealed (RTA) in nitrogen at temperatures of 750, 900 or 1000 °C for 2 m. The specific contact resistance of the contacts was measured on the CTLM patterns using ultrafine probe tips. In addition, a series of planar samples of Ti/Ni/Au on SiC was prepared to measure the depth profile in the interfacial region using a PHI 710 Scanning Auger Nanoprobe operating at a base pressure of approximately 1 x 10^{-9} torr. Auger depth profiles were obtained using a 10 keV, 1 nA primary electron beam (via a Field Emission tip), operating at a Field of View (FOV) of 50 µm. A 2 keV, 1.3 µA Ar^+ beam was rastered over a 4 mm^2 area to create the required etch crater.

RESULTS AND DISCUSSION

Fig. 1 shows a simulation using TRIM of the vacancy concentration in 3C-SiC versus depth after implantation at 5 keV. A higher peak in vacancy concentration was exhibited by P implanted ions than for C ions. The FWHM of the peak resulting from implantation with P ions was narrower (8 nm) and closer to the surface than the equivalent width of vacancy concentration for C ions (15 nm) at a dose of 1 x 10^{15} ions/cm^2.

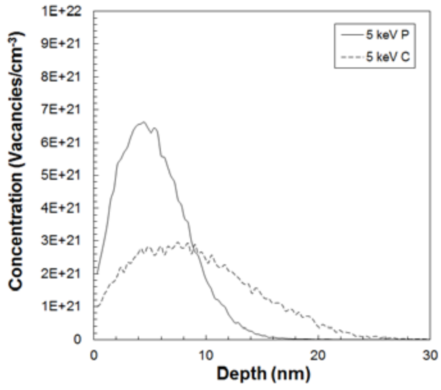

Figure 1. Vacancy concentration after implantation at 5 keV at a dose of 1 x 10^{15} ions/cm^2.

The plots of current (I) versus voltage (V) as measured in the circular TLM patterns were linear for both the as-deposited and annealed Ti/Ni/Au contacts. Fig. 2(a) shows the linear I/V plots for contacts which were annealed at 900° C. A decrease in slope or increase in resistance was evident with increase in the implant dose. In Fig. 2(b), the sheet resistance, R_{sh}, has been plotted as a function of implant dose for both as-deposited and annealed samples (900 °C). Fig. 2(b) shows an increase in R_{sh} with dose for both as-deposited and annealed contacts and a lowering in the value of R_{sh} after annealing at 900 °C. Measurements of ρ_c versus implant dose for Ti/Ni/Au contacts using the different annealing temperatures have been plotted in Fig. 3.

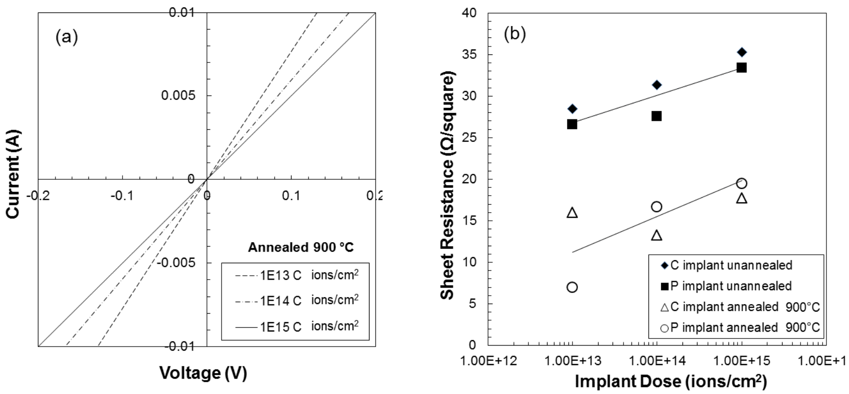

Figure 2. (a) Plots of current/voltage for Ti/Ni/Au contacts on n-SiC after implantation with C ions at 1 x 10^{13}–1 x10^{15} ions/cm^2 and (b) sheet resistance versus implant dose.

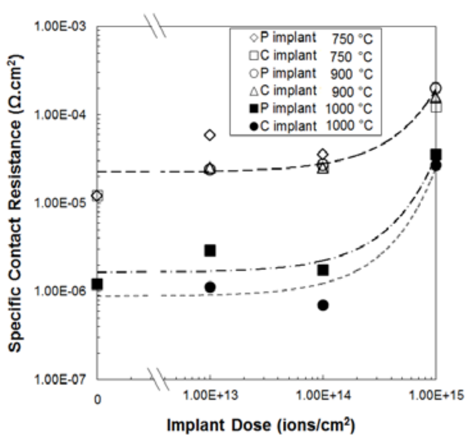

Figure 3. Specific contact resistance versus ion dose for Ti/Ni/Au contacts to n-SiC implanted with P or C ions at 5 keV.

Each of the plots in Fig. 3 was characterized by i) a low dose region (1×10^{13} and 1×10^{14} ions/cm^2) in which there was little or no change in ρ_c and ii) a higher dose region (1×10^{15} ions/cm^2) with an increase in ρ_c. For the contacts implanted at doses of 1×10^{13} and 1×10^{14} ions/cm^2 and annealed at 750 °C or 900 °C, the measurements of ρ_c were approximately constant with ρ_c in the range of 1-5 x 10^{-5} Ω.cm^2. In comparison, at the higher implant dose of 1×10^{15} ions/cm^2 and annealing at 750 °C or 900 °C the value of ρ_c was higher at ~1 x 10^{-4} Ω.cm^2 for both P and C implants. For the samples annealed at 1000 °C, the value of ρ_c for contacts formed at implant doses of 1×10^{13} and 1×10^{14} C ions/cm^2 was reduced by an order of magnitude (0.7-5.0 x 10^{-6} Ω.cm^2) compared with the equivalent contacts annealed at lower temperatures. For the higher implant dose of 1×10^{15} ions/cm^2 in combination with a 1000 °C anneal, the measurements of ρ_c were 2.0 - 4.0 x 10^{-5} Ω.cm^2. These results have indicated no clear dependence of ion species (C or P ions) on ρ_c as a function of dose and annealing temperature.

The AES depth profiles for four samples of layered and treated material are shown in Fig. 4.

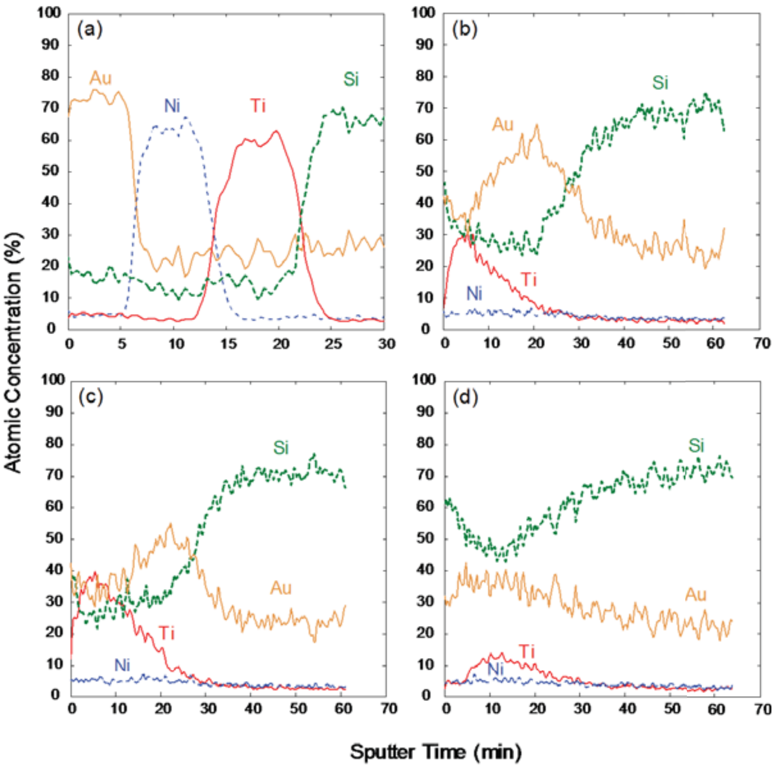

Figure 4. Auger Depth Profiles: (a) as deposited, (b) annealed at 750°C, (c) annealed at 900°C, (d) annealed at 1000°C.

Fig. 4(a), for the as-deposited contacts has indicated the sharp interfaces of the individual layers with minimal overlap. The annealed samples subjected to Auger depth profiling (profiles (b) – (d)) all had a passivation layer of SiO_2 added (100 nm) to protect the underlying layers during heating. Annealing at 750°C (Fig.4 (b)) resulted in a diffusion of the Ti layer out into the Au and SiO_2 outer layers, with the Au layer starting to diffuse towards the SiC layer, as well as outwardly into the layer of SiO_2. The Ni layer has largely diffused throughout all the layers. At an annealing temperature of 900°C (Fig.4(c)), the Ti has remained the same as for the 750°C anneal, while the Au and, in particular, the Ni appear to have inter-diffused further at this higher temperature. Annealing at 1000°C resulted in virtually total inter-diffusion of the Au and Ni, with the Ti (showing a reduced peak intensity) which also appeared to diffuse towards the layer of SiC.

The trends in ρ_c versus implant dose in Fig. 3 have been interpreted in terms of the level of damage generated by the low energy bombardment. Previously, Kameda et al. have identified four types of damage after low energy (20 keV) implantation of 6H-SiC with nitrogen [12]. a) A negligible level of damage below a dose of 5 x 10^{14} ions/cm^2, b) the extensive formation of defect structures at a dose of 5 x 10^{14} - 1 x 10^{15} ions/cm^2, c) an amorphous structure was formed at 1 x 10^{15} - 1 x 10^{17} ions/cm^2 and d) the formation of voids at the interface above 1 x 10^{17} ions/cm^2. In Fig. 3, the zone i) at lower dose with little or no effect of dose or implant species on ρ_c was consistent with the first region identified by Kameda et al [12] associated with negligible damage. In the higher dose region or zone ii) in Fig. 3, the increases in ρ_c and R_{sh} were consistent with the formation of a defect structure and also potentially regions of amorphous SiC of higher resistivity. The range of doses which defined each of the regions of damage was likely to differ slightly between the two studies because of the use a higher energy of 20 KeV by Kamedu [12] in comparison with 5 KeV in the present work.

Another trend of significance in Fig. 3 was the effect of annealing at 1000 °C in reducing ρ_c of the contacts by more than an order of magnitude. For the implant doses of 1 x 10^{13} and 1 x 10^{14} ions/cm^2, the annealing treatment at 1000 °C has resulted in the lowest values of $\rho_c \sim 0.7$-1.0×10^{-6} $\Omega.cm^2$ in this study. The Auger depth profiles have shown that this reduction in ρ_c at 1000 °C was accompanied by a large-scale indiffusion of Ti, Ni and Au into the SiC as seen in Fig. 4(d). Linchao et al have also identified a significant reduction in resistance of Ni/Ti/Ni contacts to n-SiC after annealing at 1000 °C which corresponded to a complete indiffusion of Ni into the SiC [4]. Previous work has shown that in Ni/Ti contacts to SiC, the formation of Ni_2Si during high temperature annealing (~1000 °C) was essential in obtaining ohmic behaviour [4,5]. A reduction in contact resistance during annealed of these contacts has also been attributed to the formation of $TiSi_2$ by the reaction of Ti with SiC [3,6]. In Ni/Ti contacts, the Ti has been shown to react with excess carbon to form TiC [3-5]. Evidently, from the present results, a large amount of indiffusion of interfacial layers at 1000 °C was required in order to obtain the necessary reactions for a lowering in ρ_c

For samples which were implanted at the higher dose of 1 x 10^{15} ions/cm^2, a similar order of magnitude decrease in ρ_c was measured after annealing at 1000 °C although with higher values of $\rho_c \sim 2$-4×10^{-5} $\Omega.cm^2$ than at the lower implant doses. Kamedi has shown a recrystallisation of amorphous regions of SiC at 1100°C but with no alteration in the defect structure in the lattice generated by implantation [12]. Song et al have shown that annealing at 1350 °C was necessary to give a complete recrystalisation of ion implanted damage of dislocation loops in SiC [13]. The decrease in ρ_c with annealing at 1000 °C (1 x 10^{15} ions/cm^2) was therefore attributed to changes in the interfacial structure of the contact during annealing. Despite the differences in vacancy concentration between C and the higher mass P ions as calculated by TRIM, there was little difference in ρ_c for contacts implanted with the two species of ions.

CONCLUSIONS

Au/Ni/Ti/ n-SiC contacts with a prior implantation of the SiC at low energy have exhibited ohmic current-voltage characteristics as-deposited and after annealing at temperatures of 750-1000 °C. Annealing at 1000 °C was required to obtain the minimum in $\rho_c \sim 1.0 \times 10^{-6}$ Ω.cm^2 using implant doses of 10^{13}-10^{14} ions/cm^2. The magnitude of ρ_c was approximately constant with implant dose until increasing at a dose of 1×10^{15} ions/cm^2 for both C and P ions. AES depth profiles have shown that annealing at 1000 °C was correlated with an extensive indiffusion of Ti, Ni and Au into the SiC at the interface. These results have indicated no dependence of ion species (C or P ions) on ρ_c as a function of either dose or annealing temperature.

ACKNOWLEDGEMENTS

This work was performed in part at the Queensland (Griffith) and Victorian (La Trobe) nodes of the Australian National Fabrication Facility (a company established under the National Collaborative Research Infrastructure Strategy to provide nano and microfabrication facilities for Australian Research).

REFERENCES

1. Z. Wang, W. Liu, C. Wang, *J. Electron. Mater.* **45**(1), 267 (2015).
2. Z. Zhang, J. Teng, W.X. Yuan, F.F. Zhang, G.H. Chen, *Appl. Surf. Sci.,* **255**, 6939 (2009).
3. D. Basak, S. Mahanty, *Mater. Sci.Eng.* **B98**, 177 (2003).
4. H. Linchao, S. Huajun, L. Kean, W. Yiyu, T. Yidan, B. Yun, X. Hengyu, W. Yudong, L. Xinyu, *J. Semicond.* **35**(7), 07203-1 (2014).
5. J.W. Lee, B. Angadi, H.C. Park, D.H. Park, J.W. Choi, W.K. Choi, T.W. Kim, *J. Electrochem. Soc.* **154**(10), H849 (2007).
6. F. Touati, K. Takemasa, M. Saji, *IEEE Trans. Electron Dev.* **46**(3), 444, (1999).
7. M. Grodzicki, J. Chrzanowski, P. Mazur, S. Zuber, A. Ciszewski, *Opt. Appl.* **39**(4), 765 (2009).
8. P.W. Leech, A.S. Holland, G.K. Reeves, Y. Pan, M.C. Ridgway, P. Tanner, *Mater. Lett.* **66**, 39 (2016).
9. L. F. Liu, C.H. Li, A.P. Pisano, C. Carraro, R. Maboudia, *JVSTA* **A28**, 1259 (2010).
10. L. Wang, S. Dimitrijev, J. Han, F. Iocopi, L. Hold, P. Tanner, H.B. Harrison, *Thin Solid Films* **519** 6443 (2011).
11. Y. Pan, G. K. Reeves, P. W. Leech, A. S. Holland, *IEEE Trans.Electron Dev.* **60**, 1202, (2013).
12. T. Takeda, A. Tomita, T. Matsui, T. Isshiki, *Mater. Sci. Forum* **778-780**, 350 (2014).
13. X. Song, J. Biscarrat, J-F. Michaud, F. Cayrel, M. Zielinksi, T. Chassagne, M. Portail, E, Collard, D. Alquier, *Nucl. Instr. Meth. Phys. Res.* **B269,** 2020 (2011).