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Electrical characterization of electron irradiated and annealed lowly-doped 4*H*-SiC

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ABSTRACT

The effect of high energy electron (HEE) irradiation on nickel Schottky contacts fabricated on lowly-doped *n*-type 4*H*-SiC was investigated by deep level transient spectroscopy (DLTS) and high resolution Laplace-DLTS. The Schottky contacts were deposited by resistive evaporation of nickel and were observed to be of good rectification quality from current-voltage measurements. DLTS was performed up to 350K to investigate the presence of defects before and after HEE irradiation. HEE irradiation was observed to induce three deep level defects below 350 K at 0.42 eV, 0.62 eV and 0.76 eV below the conduction band minimum. These deep level defects are labelled $E_{0.42}$, $E_{0.62}$ and $E_{0.76}$. Defects $E_{0.42}$ and $E_{0.76}$ were observed after the same electron fluence and were annealed out at the same temperature, suggesting that the defects could be strongly related. The effect of HEE irradiation and annealing on as-grown defects was also investigated and is reported.

1. Introduction

Silicon carbide is a wide bandgap semiconductor with desirable physical and electronic properties that makes it suitable for fabrication of high frequency, high temperature and high power devices [1–3]. SiC is also a radiation hard material making it a material of choice for devices that can operate in radiation harsh environments and at high temperatures [4]. Such devices include radiation detectors and devices for space applications. As with other semiconducting materials, use of a semiconductor in electronic device fabrication depends on the knowledge of presence of deep levels. This is specially so since defects can either enhance or degrade the devices. Thus, if the presence and effect of defects is well understood, controlled introduction into the material can be performed to enhance performance or the defects can be removed either by annealing or by other methods.

One way of controlled introduction of defects in semiconductors is by particle irradiation. This can be done by varying particle energy and fluence. Irradiation induces vacancies and interstitials in semiconductors, although many other complex defects can also result. Complex defects resulting from particle irradiation include Frenkel pairs, antisites (such as in compound semiconductors), vacancy pairs, etc. Particle irradiation is also used to shed light on the origin and in some instances on the microstructural nature of defects. In SiC for example, low energy particle irradiation can be used to differentiate between carbon and silicon related defects since these elements have a different threshold displacement energy. Irradiation with particles with low energy that is insufficient to produce silicon vacancies or related defects can thus only induce carbon vacancies and related defects.

Defects are described majorly using two parameters; the defect activation energy and the defect capture cross section. The defect activation energy, E_T, refers to the position of the defect level with respect to either the conduction band or the valence band. The capture cross section, σ_n, relates to how effective the defect is in trapping free carriers. The defect concentration is also another parameter used to quantify a defect level.

To the best of our knowledge, the characterization of low doping density ($\sim 4 \times 10^{14}\ \mathrm{cm^{-3}}$) 4*H*-SiC after electron irradiation and thermal annealing has not been studied by high-resolution Laplace deep level transient spectroscopy (Laplace-DLTS). In this study, high energy electron (HEE) irradiation was performed using a strontium source on nitrogen doped 4*H*-SiC with a doping density of $\sim 4 \times 10^{14}\ \mathrm{cm^{-3}}$. Defects present in as-prepared as well as defects introduced by irradiation were characterized by conventional DLTS and Laplace-DLTS. Laplace-DLTS was used to separate defect levels with closely spaced emission rate. Annealing studies were also performed to study the annealing kinetics of the

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irradiation induced-defects as well as the as-prepared and process-induced defects.

2. Experimental procedure

The samples used in this work were epitaxially grown, N-doped, *n*-type 4*H*-SiC wafers supplied by CREE Research Inc. The epilayer was grown by chemical vapour deposition on the Si-face of the SiC substrate, which has a low net doping density of $\sim 4.0 \times 10^{14}$ cm^{-3}. The orientation and thickness of the samples are 8.05° and 19.20 μm, respectively. Samples were degreased by boiling in trichloroethylene, acetone and methanol for 5 min each. This was followed by rinsing in de-ionised water before a 30 s dip in hydrofluoric acid to remove the native oxide layer. Samples were then rinsed in de-ionised water followed by blow drying using N_2 gas. Nickel ohmic contact with a thickness of 300 nm was deposited by joule evaporation onto the highly doped (1×10^{18} cm^{-3}) side of the sample and annealed in flowing argon for 10 min at 950 °C. Prior to Schottky contact fabrication, the same cleaning procedure was repeated except that instead of boiling, 3 min rinsing in each of the three solvents was performed in an ultrasonic bath. Nickel Schottky contacts with a diameter of ~ 0.6 mm and a thickness of 100 nm were then resistively deposited onto the lowly doped side of the samples. DLTS measurements were carried out using a National Instruments Digital Acquisition (DAQ) based Laplace-DLTS system [5]. HEE irradiation of the diodes was performed using a Sr-90 radioactive source. Strontium decays with an emission of ~0.55 MeV into yttrium which then decays zirconium with an emission of ~2.3 MeV. The electrons emitted from the strontium source thus have a continuous energy distribution, with more than 70% having more than 0.25 MeV. Detailed information on the irradiation source can be obtained from Auret et al. [6]. The diodes were annealed after irradiation in steps of 100 °C in flowing argon gas up to 600 °C.

3. Results and discussion

Current–voltage (*I–V*) and capacitance–voltage (*C–V*) measurements were carried out after Schottky contacts deposition to determine the rectification properties of the prepared Schottky contacts. The contacts had excellent rectification properties making it possible for DLTS measurements to be carried out on the samples. *I–V* and *C–V* measurements were also repeated after each irradiation step to determine if the diodes were still suitable for DLTS measurements. Parameters which were monitored from these measurements to determine the suitability of the diode for DLTS measurements include the ideality factor (n), the Schottky barrier height obtained from both *I–V* (Φ_{I-V}) and *C–V* (Φ_{C-V}) and the reverse leakage current measured at −5 V. The ideality factor was obtained assuming the thermionic emission process to be the dominant current transport mechanism across the Schottky barrier. A summary of the obtained *I–V* and *C–V* measurements is presented in Table 1, as earlier reported in Ref. [7]. A more detailed *I–V* and *C–V* study on characteristics of 4H-SiC with HEE irradiation can be obtained from Omotoso et al. [8] It was demonstrated that 4H-SiC retains its rectification properties after fluence of 6×10^{14} cm^{-2} at room temperature [8].

DLTS was carried out on the diodes before HEE irradiation, after HEE irradiation and during annealing of the irradiated diodes. A control measurement was done by annealing as-prepared diodes and taking DLTS measurements. Annealing the as-prepared diodes between 100 °C and 700 °C did not induce new defects. Any defect that may be observed during annealing of irradiated diodes is therefore as a result of irradiation and not annealing.

3.1. Irradiation results

DLTS measurements were carried out in the temperature range 30 K ≤ T ≤ 350 K. The reverse bias was maintained as 5 V while a forward voltage pulse of −1 V with a pulse width of 1 ms was applied. Normalized DLTS spectra obtained using a 2.5 s^{-1} rate window is given in Fig. 1 showing the presence of four defect levels. The energy levels were labelled $E_{0.10}$, $E_{0.12}$, $E_{0.18}$ and $E_{0.69}$ with activation energies of E_C – 0.10 eV, E_C – 0.12 eV, E_C – 0.18 eV and E_C – 0.69 eV. The level $E_{0.10}$ is a shallow nitrogen donor occupying a cubic lattice site [9,10]. When occupying a hexagonal site, nitrogen dopants introduce shallow energy levels with an activation energy of ~0.055 eV [11,12].

Energy level $E_{0.12}$ was observed and has been attributed to a titanium impurity [13,14]. $E_{0.18}$ has also been attributed to a titanium impurity [14,15]. It is possible that these two defect levels are both induced by a titanium impurity but occupying different geometrical positions in SiC. The defect is seen as $E_{0.12}$ when occupying a hexagonal position while it appears as $E_{0.18}$ when occupying a cubic position. A comprehensive study of the two defect levels is provided by Achtziger et al. after implanting 4*H*-SiC with radioactive isotopes of chromium and vanadium [14]. The authors who also observe defect level $E_{0.18}$ however attribute it to a chromium impurity in contrast to Castaldini et al. [15]. Metal impurities in as-grown semiconductors are introduced during growth of the semiconductors.

Defect level $E_{0.69}$ observed in this study is widely believed to be of intrinsic nature. From the literature, this defect has been attributed to a carbon vacancy, V_C [16], a silicon vacancy [17], a silicon-carbon divacancy, $V_{Si} + V_C$, a carbon silicon antisite pair, $Si_C + C_{Si}$ [18]. It has also been claimed that the defect could be a silicon antisite–silicon vacancy complex, $C_{Si}–V_{Si}$ [18] and a nitrogen impurity next to a carbon interstitial I_C + N [19]. The problem of whether the defect level is either a silicon or carbon related defect was resolved by Son et al., and has been identified as double acceptor of an isolated carbon vacancy [20].

The normalized DLTS spectra obtained in this study after HEE irradiation are shown in Fig. 2. The presented spectra were obtained using a reverse bias of 5 V, a forward pulse of −1 V with a pulse width of 1 ms at a 2.5 s^{-1} rate window. The *signatures* (activation energy and apparent capture cross section) of the deep level defects show in Fig. 3 were determined as reported earlier [21].

Using step by step high energy electron irradiation, defect level $E_{0.69}$ was observed to increase in concentration from 2×10^{13} cm^{-3} before irradiation to 9×10^{13} cm^{-3} after a total fluence of 5.4×10^{14} cm^{-2}. Apart from level $E_{0.69}$, another defect level with activation energy of E_C – 0.62 eV was observed as the fluence

Table 1
The diode ideality factor, Schottky barrier height and reverse leakage current obtained before irradiation, after irradiation and after annealing the irradiated diodes at 600 °C.

Fluence	n ± 0.02	Φ_{I-V} (eV) ± 0.02	Φ_{C-V} (eV) ± 0.02	I_L (A) at −5 V
0	1.05	1.62	2.13	1.2×10^{-14}
5.4×10^{14} cm^{-2}	1.04	1.60	2.14	5.2×10^{-13}
5.4×10^{14} cm^{-2} + 600 °C annealing	1.06	1.58	2.20	3.2×10^{-12}