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<td>Author(s)</td>
<td>Kawahara, Koutarou; Suda, Jun; Kimoto, Tsunenobu</td>
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Analytical model for reduction of deep levels in SiC by thermal oxidation
Koutarou Kawahara, Jun Suda, and Tsunenobu Kimoto

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Analytical model for reduction of deep levels in SiC by thermal oxidation

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Two trap-reduction processes, thermal oxidation and C⁺ implantation followed by Ar annealing, have been discovered, being effective ways for reducing the Z₁/₂ center (E_C – 0.67 eV), which is a lifetime killer in n-type 4H-SiC. In this study, it is shown that new deep levels are generated by the trap-reduction processes in parallel with the reduction of the Z₁/₂ center. A comparison of defect behaviors (reduction, generation, and change of the depth profile) for the two trap-reduction processes shows that the reduction of deep levels by thermal oxidation can be explained by an interstitial diffusion model. Prediction of the defect distributions after oxidation was achieved by a numerical calculation based on a diffusion equation, in which interstitials generated at the SiO₂/SiC interface diffuse to the SiC bulk and occupy vacancies related to the origin of the Z₁/₂ center. The prediction based on the proposed analytical model is mostly valid for SiC after oxidation at any temperature, for any oxidation time, and any initial Z₁/₂ concentration. Based on the results, the authors experimentally achieved the elimination of the Z₁/₂ center to a depth of about 90 μm in the sample with a relatively high initial-Z₁/₂-concentration of 10¹³ cm⁻³ by thermal oxidation at 1400°C for 16.5 h. Furthermore, prediction of carrier lifetimes in SiC from the Z₁/₂ profiles was realized through calculation based on a diffusion equation, which considers excited-carrier diffusion and recombination in the epilayer, in the substrate, and at the surface. © 2012 American Institute of Physics. [http://dx.doi.org/10.1063/1.3692766]

I. INTRODUCTION

SiC has attracted increasing attention as a promising wide-bandgap semiconductor for realizing high-power, high-temperature, and high-frequency devices. One of the dominant obstacles to realizing high-performance SiC devices is existence of deep levels in SiC epilayers. Deep levels are generated during epitaxial growth and device fabrication steps such as ion implantation and reactive ion etching. The deep levels work as recombination centers resulting in reduction of the carrier lifetime and also work as carrier traps, leading to reduction of the conductivity. Therefore, deep levels, especially the Z₁/₂ center, a well known deep level that is a lifetime killer in n-type 4H-SiC, must be controlled. The typical carrier lifetime for commercially available SiC epilayers is only 0.6-1 μs due to the high Z₁/₂ concentration of about 10¹³ cm⁻³, while over 5 μs lifetime is required for 10 kV PiN-diodes.

The origin of Z₁/₂ center has been extensively investigated. The authors speculate that Z₁/₂ center is a carbon-vacancy (V_C)-related defect because of the following three experimental results. First, the Z₁/₂ center is generated by low-energy (100–200 keV) electron irradiation, which displaces only C atoms, without subsequent annealing. Second, the Z₁/₂ concentration is lower in the samples grown at 2 concentration is lower in the samples grown at lower C/Si ratio. Third, the Z₁/₂ is a thermally stable defect, showing no change of concentration up to 1600°C, at which temperature interstitials can easily diffuse. Additionally, it has been reported that the energy levels and the negative U nature originating from V_C obtained by ab initio calculation (density functional theory) are in agreement with the properties of the Z₁/₂ center detected by deep level transient spectroscopy (DLTS). There are two effective methods to reduce the Z₁/₂ center: (i) C⁺ implantation followed by Ar annealing and (ii) thermal oxidation. In the case of the C⁺-implantation process, implanted excess carbon atoms should diffuse to the deeper region of an SiC epilayer and occupy vacancies, resulting in reduction of the Z₁/₂ concentration. In the case of thermal oxidation, the Z₁/₂ reduction may be ascribed to diffusion of interstitials generated at the SiO₂/SiC interface during oxidation, as shown in Fig. 1. There are several reports indicating the generation of interstitials at the SiO₂/SiC interface, which support this trap reduction model. (i) Experimental oxidation rate of SiC can be well simulated by considering silicon and carbon emission from the oxidation interface. (ii) An increase of carbon atom concentration at the SiO₂/SiC interface was observed by electron energy loss spectroscopy. (iii) It has been predicted from ab initio calculation that carbon clusters are formed at the SiO₂/SiC interface during oxidation. The trap reduction mechanism by thermal oxidation, however, has not been fully understood.

The purposes of this study are to ascertain the trap-reduction model shown in Fig. 1 and to enable prediction of the depth profiles of deep levels after oxidation in order to control deep levels by thermal oxidation. For these purposes, the authors compare deep levels in the SiC epilayers after the C⁺-implantation process and those after thermal oxidation, and investigate the oxidation-temperature dependence.
oxidation-time dependence, and initial-$Z_{1/2}$-concentration dependence of the defect reduction in both n-type and p-type 4H-SiC epilayers. Based on these data, an analytical model of defect reduction as well as the most effective way to reduce the $Z_{1/2}$ center are proposed. In addition, the effect of the $Z_{1/2}$ reduction on the carrier lifetime is investigated through the relation between the carrier lifetime and the $Z_{1/2}$ depth-profile after the oxidation process under various conditions.

II. EXPERIMENTS

The starting materials were n-type ($N_d : 10^{14} - 10^{15}$ cm$^{-3}$) and p-type ($N_d : 10^{15} - 10^{16}$ cm$^{-3}$) 4H-SiC (0001) epilayers. A series of samples was oxidized at different temperatures (1150–1400°C) for various times (1.3–16.5 h) in 100% oxygen ambient, while the other set of samples was implanted with 10–50 keV carbon ions with a total dose of $1 \times 10^{13}$ cm$^{-2}$ or $1 \times 10^{14}$ cm$^{-2}$ (implanted atom concentration: $1 \times 10^{18}$ cm$^{-3}$ or $1 \times 10^{19}$ cm$^{-3}$), forming a 140-nm-box-profile. The $C^+$-implanted samples were annealed in Ar ambient at various temperatures (1000–1800°C) for 20 min. For DLTS measurements, Ni and Ti were employed as Schottky contacts on n-type and p-type samples, respectively. The typical diameter of Schottky contacts was 1 mm. A Ti/Al/Ni (20 nm/100 nm/80 nm) layer annealed at 1000°C for 2 min was employed as backside ohmic contacts for p-type materials. A period width of 0.205 s was employed for all DLTS measurements performed in this study. The depth profiles of deep-level concentrations until 10-$\mu$m depth were measured by changing the reverse bias voltage up to 100 V in DLTS measurements. To monitor deeper regions (over 10 $\mu$m), the samples were mechanically polished from the surfaces, and DLTS measurements were repeated. It was confirmed that additional deep levels did not appear after the polishing. The carrier lifetime was measured at room temperature by microwave photoconductance decay (µ-PCD) equipped with an yttrium lithium fluoride-third harmonic generation laser ($\lambda = 349$ nm) as an excitation source.

III. RESULTS AND DISCUSSION

A. Defect distributions after trap reduction processes

At first, deep levels after the two trap-reduction processes are compared. Figure 2 and 3 shows the DLTS spectra of the n-type/p-type samples after dry oxidation at 1300°C for 1.3 h, as well as the samples after $C^+$ implantation (dose: $1 \times 10^{14}$ cm$^{-2}$) and subsequent Ar-annealing at 1800°C/1300°C for 20 min. Here, the signal $b_1$ is the coefficient of the first sine term in the Fourier series of deep level transient Fourier spectroscopy. After the $C^+$-implantation process (dashed-dotted lines), the new peaks, ON1 ($E_C – 0.84$ eV), ON2 ($E_C – 1.1$ eV), and ON3 ($E_C – 1.6$ eV), are reduced in n-type SiC (Fig. 2), and HK0 (Refs. 4 and 26) ($E_V + 0.79$ eV) in p-type SiC (Fig. 3), appeared, while the same four peaks are also observed after thermal oxidation (dashed lines). The ON1 and ON2 centers should correspond to the deep levels reported as “new traps” in SiC after $C^+$-implantation process. On the other hand, two major deep levels, $Z_{1/2}$ ($E_C – 0.67$ eV) and EH$_6/7$ ($E_C – 1.6$ eV), are reduced in n-type SiC after thermal oxidation as shown by the dashed line in Fig. 2. The $Z_{1/2}$ and EH$_6/7$ centers are also reduced to below the detection limit by $C^+$ implantation followed by Ar annealing at 1500°C (not shown), while they are regenerated by high-temperature (over 1700°C) annealing as shown by the dashed-dotted line in Fig. 2. The reduced deep levels as well as generated levels by the trap-reduction processes are summarized in Table I. The defect behaviors (generation and reduction) for thermal oxidation clearly agree with those for the $C^+$-implantation process, which indicates that similar phenomena (such as interstitial diffusion) may occur in the two processes.
TABLE I. Reduced and generated defects in SiC by C\textsuperscript{+}-implantation process and by thermal oxidation. The conduction types of the samples where each defect is observed are shown in parentheses.

<table>
<thead>
<tr>
<th>C\textsuperscript{+} implantation + Ar annealing</th>
<th>Thermal oxidation</th>
</tr>
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<tbody>
<tr>
<td>Reduced defects</td>
<td>Generated defects</td>
</tr>
<tr>
<td>Z\textsubscript{1/2} (n-type)</td>
<td>ON\textsubscript{1} (n-type)</td>
</tr>
<tr>
<td>EH\textsubscript{0}/7 (n-type)</td>
<td>ON\textsubscript{2} (n-type)</td>
</tr>
<tr>
<td>HK0 (n-type)</td>
<td>ON\textsubscript{3} (n-type)</td>
</tr>
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</table>

It is important to investigate the depth profiles of generated and reduced defects in order to understand what phenomena occur during the trap-reduction processes. Figure 4 shows the depth profiles of ON\textsubscript{1} center (generated defect) after oxidation at various temperatures for 1.3 h. With increasing oxidation temperature, the ON\textsubscript{1} concentration increases and is distributed to a deeper region, suggesting that the ON\textsubscript{1} center is related to the atoms, most likely interstitials, diffused from the SiO\textsubscript{2}/SiC interface. The ON\textsubscript{2} and HK0 centers (generated defects) showed similar behavior (not shown). Figure 5 shows the depth profiles of Z\textsubscript{1/2} center (reduced defect) after oxidation at various temperatures for 1.3 h. In this particular case, the initial Z\textsubscript{1/2} concentration was increased to $1.7 \times 10^{14}$ cm\textsuperscript{-3} by electron irradiation (energy: 150 keV, fluence: $1 \times 10^{17}$ cm\textsuperscript{-2}) in order to investigate trap reduction in the sample with high initial-Z\textsubscript{1/2}-concentration. With increasing oxidation temperature, the Z\textsubscript{1/2} concentration decreases and is eliminated to a deeper region, suggesting that the Z\textsubscript{1/2} center is related to the (carbon) vacancies, which are occupied by diffused interstitials during oxidation. In addition, the depth of the Z\textsubscript{1/2}-elimination region is proportional to the square root of the oxidation time ($t^{1/2}$). These results clearly suggest that the diffusion phenomena is taking place in the SiC bulk region during thermal oxidation.

FIG. 4. (Color online) Depth profiles of ON\textsubscript{1} center after oxidation at various temperatures for 1.3 h. Each symbol indicates the experimental data and each line indicates the calculated $n\textsubscript{1}$ distribution obtained from Eqs. (1)–(8).

FIG. 5. (Color online) Depth profiles of Z\textsubscript{1/2} center after oxidation at various temperatures for 1.3 h. The initial Z\textsubscript{1/2}-concentration is $1.7 \times 10^{14}$ cm\textsuperscript{-3}. Each symbol indicates the experimental data and each line indicates the calculated $n\textsubscript{V}$ distribution obtained from Eqs. (1)–(8).

B. Analytical model for defect distributions after thermal oxidation

As discussed above, the reduction of the Z\textsubscript{1/2} center by oxidation can be explained by diffusion of interstitials and recombination with vacancies as shown in Fig. 1. Therefore, to predict defect distributions after thermal oxidation, the following diffusion equations are solved.

Evolution equations

$$\frac{\partial m}{\partial t} = D \cdot \frac{\partial^2 m}{\partial x^2} - \gamma \cdot m \cdot n\textsubscript{V},$$

(1)

$$\frac{\partial n\textsubscript{V}}{\partial t} = -\gamma \cdot n\textsubscript{1} \cdot n\textsubscript{V},$$

(2)

Boundary and initial conditions

$$-D \cdot \frac{\partial n\textsubscript{1}}{\partial x}\big|_{x=0} = F_0 \cdot r^{-2}(t \neq 0),$$

(3)

$$n\textsubscript{1}\big|_{x=0} = 0,$$

(4)

$$n\textsubscript{V}\big|_{x=0} = n\textsubscript{V0}.$$  

(5)

Fitting parameters

$$D = D\textsubscript{\infty} \cdot \exp\left(\frac{-E_D}{kT}\right),$$

(6)

$$F_0 = F_0\textsubscript{\infty} \cdot \exp\left(\frac{-E_R}{kT}\right),$$

(7)

$$\gamma = \gamma\textsubscript{\infty} \cdot \exp\left(\frac{-E_R}{kT}\right),$$

(8)

where $n\textsubscript{1}$ and $n\textsubscript{V}$ are the concentrations of interstitials and vacancies, latter of which is related to the origin of Z\textsubscript{1/2} center. $D$ denotes the diffusion coefficient of the interstitials, and $\gamma$ is the recombination coefficient between an interstitial and a vacancy. In this model, vacancies are assumed to be
The slope of the plot is described as a proportional to \( t_{ox} \), where \( t_{ox} = 1 \) s (ref. 22). Since the oxidation rate becomes slow with the time, the gradual decrease in flux of emitted interstitials as the oxidation (time) proceeds is considered by introducing a slowdown coefficient \( x \). The flux of interstitials should be in proportion to the oxidation rate. The slowdown of the oxidation reaction at the interface can be estimated from Eq. 6, which shows the dependence of the oxide growth rate on oxidation time at different oxidation temperatures.

**FIG. 6.** (Color online) Dependence of the oxide growth rate on oxidation time at different oxidation temperatures. A result (1100 °C) reported by Hijikata et al. (Ref. 22) is also plotted in the same figure. When the slope of the plot is described as \(-x\), the oxidation rate is proportional to \( t_{ox}^{-x} \). Therefore, interstitial emission can be assumed to decrease in proportion to \( t_{ox}^{-x} \) as shown in Eq. (3). For simplicity, the time-dependent oxidation-rate is expressed in two stages: high-oxidation-rate stage \((t_{ox} < 0.8\) h\), where \( x \) is unity for 1150–1300 °C oxidation and 0.48 for 1400 °C oxidation; and low-oxidation-rate stage \((0.8 < t_{ox})\), where \( x \) is 0.23 for 1150–1300 °C oxidation and 0.48 for 1400 °C oxidation. It is assumed that \( n_1 \) before oxidation is negligible compared with that after oxidation as described in Eq. (4). A parameter \( n_{V_{ini}} \) in Eq. (5) denotes the initial vacancy distribution before oxidation. As shown in Eqs. (6)–(8), each parameter is described as a function of temperature. The parameters, \( E_{aD} \), \( E_{aF} \), and \( E_{aV} \), indicate the activation energy corresponding to the energy barrier for migration of interstitials, that for generation of interstitials, and that for recombination of interstitials with vacancies, respectively.

**C. Comparison between experimental \( Z_{1/2} \) profile and calculated vacancy profile after thermal oxidation**

Based on Eqs. (1)–(8), the depth profiles of \( n_1 \) and \( n_V \) can be calculated, which are dependent on six fitting parameters: \( E_{aD}, D_{\infty}, E_{aF}, F_0, E_{aV}, \) and \( \gamma_{\infty} \). As mentioned, the authors experimentally obtained the depth profiles of \( Z_{1/2} \) concentration after oxidation at different temperatures for several samples with different initial \( Z_{1/2} \) concentrations.

**FIG. 7.** (Color online) Depth profiles of \( Z_{1/2} \) center after oxidation at various temperatures for 1.3 h. The initial \( Z_{1/2} \) concentration is \( 2 \times 10^{13} \) cm\(^{-3}\). Each symbol indicates the experimental data and each line indicates the calculated \( n_V \) distribution obtained from Eqs. (1)–(8).

**FIG. 8.** (Color online) Effects of changing parameters, \( D, F_0, \) and \( \gamma \) on calculated \( n_V \) profile. Higher \( D/F_0/\gamma \) is used in the calculation for dashed/dotted/dashed-dotted line compared to the calculation for reference (solid line).
concentration after oxidation at different temperature. The calculated $n_V$ profiles after the fitting are shown as curve lines in Figs. 5 and 7. The obtained values of fitting parameters are summarized in Table II, which are used in all cases in this study. The activation energy for diffusion coefficient ($E_{\alpha D}$) of interstitials reducing the $Z_{1/2}$ center was determined as 0.6 eV in this study. On the other hand, the migration barrier for carbon/silicon interstitials in n-type SiC has been reported to be (0.5–0.7) eV/(1.4–1.5) eV by Bockstedte et al. based on theoretical calculation using density functional theory in the local density approximation, and by Gao et al. using molecular dynamics simulations. The activation energy for the diffusion coefficient of interstitials (0.5–0.7 eV), which obtained in this study (0.6 eV) agrees with the reported activation energy for the diffusion coefficient of interstitials ($E_{\alpha D}$) of interstitials reducing the $Z_{1/2}$ center. To minimize the oxidation time, the authors propose three approaches: (i) removing the oxide layer during oxidation, (ii) high-temperature annealing after oxidation, and (iii) higher-temperature oxidation.

(i) Removing the oxide layer during oxidation: In the initial oxidation stage, the interstitial emission rate from the SiO$_2$/SiC interface is high, as shown in Fig. 6. Therefore, removing the oxide layer during oxidation should promote the interstitial emission and thereby $Z_{1/2}$ reduction. Figure 12 shows the depth profiles of $Z_{1/2}$ center after oxidation at 1300 °C for 15.9 h. Each line indicates an $n_V$ profile calculated with the parameters shown in the Table II, and each symbol indicates experimental data. The rhombuses denote the result for continuous 15.9-h oxidation, and reverse

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### Table II. Parameter values obtained by fitting results of calculated $n_V$ profiles based on Eqs. (1)–(8) and experimental $Z_{1/2}$ profiles shown in Fig. 5. The top row indicates the "X" in the first column.

<table>
<thead>
<tr>
<th>$D$</th>
<th>$F_0$</th>
<th>$\gamma$</th>
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<tbody>
<tr>
<td>0.6 eV</td>
<td>1.4 eV</td>
<td>2.1 eV</td>
</tr>
<tr>
<td>$9.7 \times 10^{-9} \text{ cm}^2\text{s}^{-1}$</td>
<td>$4.4 \times 10^{14} \text{ cm}^{-2} \text{s}^{-1}$</td>
<td>$1.4 \times 10^{-10} \text{ cm}^2\text{s}^{-1}$</td>
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**FIG. 9.** (Color online) Depth profiles of $Z_{1/2}$ center (initial $Z_{1/2}$ concentration: $1.3 \times 10^{13} \text{ cm}^{-3}$) after oxidation at 1300 °C for 1.3-15.9 h. Each symbol indicates the experimental data and each line indicates the calculated $n_V$ distribution.

**FIG. 10.** (Color online) Depth profiles of $Z_{1/2}$ center (initial $Z_{1/2}$ concentration: $2 \times 10^{12} \text{ cm}^{-3}$) after oxidation at 1300 °C for 5.3-15.9 h. Each symbol indicates the experimental data and each line indicates the calculated $n_V$ distribution.
triangles for 15.9-h oxidation with removing the oxide layer after every 5.3-h oxidation. Removing the oxide layer is clearly effective for enhancing the reduction of the Z_{1/2} center. In the calculation, the effect of removing oxide layer is included by resetting the flux of emitted interstitials, which decreases as the oxidation proceeds, to the initial value after every (in this case 5.3 h) oxidation. The good agreement between experimental data and calculated results again supports the analytical model for trap reduction proposed in this study.

(ii) High-temperature annealing after oxidation: Dif-
 fused interstitials should remain in an epilayer after oxidation. Therefore, subsequent high-temperature annealing should enhance diffusion of the residual interstitials to the deeper region and promote Z_{1/2} reduction. Figure 13 shows the Z_{1/2} profiles after oxidation as well as after oxidation and subsequent high-temperature (1500 °C) annealing. The solid line indicates the calculated n_{V} profile just after oxidation at 1300 °C for 15.9 h, and the dashed line denotes that after oxidation and subsequent annealing at 1500 °C for 2 h in Ar ambient. To calculate the effects of Ar annealing at 1500 °C, the same parameters obtained in this study are used except that F_{0} = 0 (no additional emission of interstitials during the Ar annealing). As shown in Fig. 13, the Z_{1/2} center is eliminated to the deeper region by the subsequent annealing. The experimental data shown as symbols (reverse triangles: after oxidation; closed circles: after oxidation followed by Ar annealing at 1500 °C) clearly agree with the predicted lines, which indicates that the present analytical model is useful for predicting trap distributions not only after oxidation but also after subsequent Ar annealing. In addition, this annealing reduces the HK0 center generated in p-type SiC by thermal oxidation, which also means that residual interstitials further diffuse to the deeper region and promote the Z_{1/2} reduction by the subsequent annealing.

(iii) Higher-temperature oxidation: Because all parameters, D, F_{0}, and γ, should increase at higher temperature, oxidation at higher temperature must be effective in reduction of the Z_{1/2} centers. Figure 14 shows the depth profiles of Z_{1/2} center after oxidation at 1400 °C for 5.5 h and 16.5 h. Each line indicates the n_{V} profile calculated with the same parameter values, and each symbol indicates experimental data. The Z_{1/2} centers are eliminated to the depth of about 60 μm after oxidation for 5.5 h, which also agrees with the calculated result. The Z_{1/2} centers could be further reduced but appear to remain at the depth of about 95 μm after oxidation for 16.5 h, while they are completely eliminated in the calculated result. (Calculated n_{V} distribution after 16.5-h oxidation is not shown because the n_{V} is lower than 1 x 10^{10} cm^{-3} in the epilayer.) It should be mentioned that the interface between the epilayer and the substrate is located at the depth of about 96 μm from the surface in this sample. Therefore, the data points near the interface might contain the
substrate information (high concentration of the $Z_{1/2}$ center must exist in the substrate). Regardless, it was clarified that thermal oxidation at 1400°C is a very effective way in accelerating $Z_{1/2}$ reduction.

In addition to the above three approaches, higher-rate-oxidation processes such as wet oxidation and plasma oxidation at high temperature may be effective in reducing required oxidation time.

E. Relation between carrier lifetime and $Z_{1/2}$ profile

As mentioned in the introduction, the $Z_{1/2}$ center acts as a lifetime killer in n-type 4H-SiC. Therefore, the deeper region the $Z_{1/2}$ center is eliminated to, the longer carrier lifetime should be. Figure 15 shows $\mu$-PCD decay curves for the 96-µm-thick n-type SiC epilayers ($N_{d}: 2 \times 10^{15}$ cm$^{-3}$) on the n-type SiC substrates (thickness: ∼350 µm) after different oxidation processes. The oxidation temperature is 1300°C in Fig. 15 except for the signal labeled “1400°C.” The excitation photon density is $1.1 \times 10^{14}$ cm$^{-2}$, which leads to a high carrier injection level of $10^{16}$ cm$^{-3}$. The carrier lifetimes obtained from the decay curves are also described in the figure. The carrier lifetime increased from 0.6 µs (as-grown) to 6.5 µs by oxidation at 1400°C for 16.5 h. The carrier lifetimes are derived from the slopes of the decay curves at the points where the $\mu$-PCD signals decrease to $e^{-3}$ of the initial intensity because the initial fast decay severely suffers from carrier recombination at the surface and in the substrate. Figure 16 indicates the depth profiles of $Z_{1/2}$ concentration in the same samples as in Fig. 15, measured by DLTS. It is obvious that a longer carrier lifetime was obtained in a sample possessing a deeper $Z_{1/2}$-eliminated-region.

Here, the authors tried to quantitatively estimate the relation between the carrier lifetime and the $Z_{1/2}$ profile by a numerical simulation. It has to be noticed that the “measured” carrier lifetime does not represent the true “bulk” carrier lifetime in the epilayer itself because there are other recombination paths of excess carriers. Excess carriers generated by the excitation laser recombine at the surface as well as in the epilayer. The carriers excited in the epilayer also diffuse toward the surface and the substrate due to gradient of the carrier density, which promotes carrier recombination at the surface and in the substrate. Therefore, the measured carrier lifetime contains the effects of carrier diffusion and recombination at the surface, in the epilayer, and in the substrate.
In this study, a simulation model, which considers excited-carrier diffusion and recombination in the epilayer, substrate, and at the surface, was employed to estimate the measured carrier lifetimes. An ambipolar diffusion constant of 4.2 cm²/s (D_{amb}) was assumed for the epilayer, while a standard hole diffusion constant of 0.3 cm²/s (D_{hab}) was assumed for the substrate. The optical absorption coefficient at an excitation laser wavelength (349 nm) was determined as 330 cm⁻¹ from the literature. Figure 17 shows a schematic illustration of an SiC epilayer grown on an SiC substrate after thermal oxidation. In this study, the epilayer was divided into two regions, the Z₁₂-eliminated region (Z₁₂ concentration < 2 × 10¹¹ cm⁻³) and the Z₁₂-remaining region (Z₁₂ concentration = 1.3 × 10¹³ cm⁻³, initial value). Taking account of results in the literature, the bulk carrier lifetime in Z₁₂-eliminated region (τ_{n,2D}) and surface recombination velocity (S₁₂) were assumed to be 50 μs (Ref. 31) and 1000 cm/s (smooth surfaces were assumed), respectively. Because the backside recombination velocity (S₂₂) has very little effect on the effective lifetime due to the thick substrate (over 300 μm) and low carrier lifetime in the substrate (0.04 μs), it was assumed as infinity. From the carrier lifetime measured in the as-grown sample, 0.6 μs, the bulk carrier lifetime in the Z₁₂-remaining region (τ₂₂) can be estimated as 0.7 μs. It is mentioned again that the “measured” carrier lifetime, 0.6 μs, is shorter than the “bulk” carrier lifetime, 0.7 μs, due to the recombination at the surface and in the substrate. The boundaries between the Z₁₂-eliminated region and the Z₁₂-remaining region were defined to be 35 μm for a sample oxidized at 1300°C for 5.3 h, 55 μm for oxidized at 1300°C for 15.9 h, 70 μm for oxidized at 1300°C for 15.9 h followed by Ar annealing, and 96 μm for oxidized at 1400°C for 16.5 h from the surface as shown in Fig. 16. By using the simulation model and parameters, “effective” carrier lifetimes (lifetime obtained from a decay curve) were calculated. Table III shows the comparison between the carrier lifetimes measured by μ-PCD and the effective lifetime predicted by the simulation. The predicted effective lifetimes agree well with the measured carrier lifetimes. In the case of 1400°C oxidation, however, the measured carrier lifetime (6.5 μs) is longer than the predicted carrier lifetime (3.9 μs). The Z₁₂ center in the substrate near the epilayer/substrate interface might be also reduced by the intensive oxidation, which resulted in longer carrier lifetime.

IV. CONCLUSION

To clarify the mechanism of trap reduction by thermal oxidation, the authors investigated deep levels after two trap-reduction processes, thermal oxidation and C⁺⁺ implantation followed by Ar annealing. It was revealed that deep levels generated by thermal oxidation are the same as those generated by C⁺⁺ implantation and subsequent Ar-annealing. In addition, the depth profiles of generated/reduced defects represent the distribution of interstitials/vacancies after diffusion phenomena of interstitials from the surface to the SiC bulk region. These results indicate that the same phenomena, diffusion of interstitials, occur during these processes. Furthermore, the authors proposed an analytical model enabling prediction of Z₁₂ distribution after thermal oxidation using a diffusion equation. This model could reproduce the depth profiles of the Z₁₂ center in the SiC with different initial-Z₁₂-concentration after oxidation at any temperature and for any oxidation time. The results suggested that long-time oxidation is required for the elimination of the Z₁₂ center in the SiC when the initial Z₁₂-concentration is high. Thus, it is important to keep the initial Z₁₂-concentration, which depends on the conditions of epitaxial growth and performed device processes, low (<10¹³ cm⁻³) for achieving long carrier lifetimes. To enhance the Z₁₂ reduction and reduce process time, three methods, removing the oxide layer during oxidation, 1500°C annealing after oxidation, and higher-temperature oxidation, were proposed and experimentally proved to be effective. Especially, increasing oxidation temperature was the most effective for enhancement of the Z₁₂ reduction. Post-oxidation annealing at 1500°C could reduce the HK0 center, which was generated in SiC epilayers by oxidation. Therefore, to achieve a thick Z₁₂-free-region for the SiC epilayer with high initial-Z₁₂-concentration (>10¹³ cm⁻³), 1400°C oxidation followed by Ar annealing at 1500°C is recommended. Creating thick Z₁₂-free-region clearly improved the carrier lifetime, which agrees well with the values obtained by calculation based on a diffusion equation, which considers excited-carrier diffusion and recombination in the epilayer, in the substrate, and at the surface. By using the analytical models for predictions of Z₁₂ profile and carrier lifetime, effective carrier lifetimes can be derived from only the depth profile of initial Z₁₂-concentration and oxidation conditions.

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How to determine the $\tau_Z$: In the case of as-grown samples, the bulk carrier lifetime in the whole region of the epilayer is $\tau_Z$ (there is no Z$_1$-eliminated region). When the bulk carrier lifetime ($\tau_Z$) is assumed to be 0.7 $\mu$s, 0.6 $\mu$s is obtained as calculated effective lifetime. Because this 0.6 $\mu$s is equal to the measured carrier lifetime in as-grown sample, $\tau_Z$ is determined to be 0.7 $\mu$s.