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Photoluminescence Study of Radiation-Enhanced Dislocation Glide in 4H-SiC

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Expansion of Shockley stacking faults in 4H-SiC achieved by the radiation-enhanced glide of 30°-Si(g) partial dislocations has been investigated by photoluminescence experiments. The enhancement rate of the dislocation glide that was induced by light illumination in the present study was found to be governed directly by the light intensity, not by the photogenerated carriers. This fact indicates that the glide enhancement in 4H-SiC is not explained by the widely speculated mechanism that the electronic energy released on recombination of photoexcited electron–hole pairs is utilized to assist the dislocation glide. A photoionization of dislocations is proposed to be the cause of the glide enhancement.

4H-SiC unipolar power devices employing electrons as carriers are now available on the market. However, the commercialization of 4H-SiC bipolar power devices has been hindered by the increase of the forward voltage drop caused by the expansion of Shockley stacking faults when both electrons and holes are injected.[1,2] The expansion of these faults is induced by the glide of the 30° Si core [30°-Si(g)] partial dislocations bounding the stacking faults.[3,4] The dislocation glide induced by irradiation with light and electrons is known as a radiation-enhanced dislocation glide (REDG) and is observed in many covalent semiconductors including 4H-SiC.[5,6] Because of the similarities in experimental features, the cause of the partial dislocation glide in the 4H-SiC bipolar structures has, to date, been attributed to a recombination-enhanced defect reaction (REDR) in which the energy released on nonradiative recombination of electrons and holes via defect levels is converted to kinetic energy for the reaction of the defect.[7,8] However, theory has shown for SiC that the energy barrier for kink migration on 90°-Si(g) partial dislocations can differ depending on the charge state of the kink.[9] Thus, forward current could induce the ionization of the kinks in 4H-SiC, and this could in turn enhance the dislocation glide. So far, however, there has been no experimental proof for any of the above mechanisms. To suppress the anomalous expansion of stacking faults on the sound basis, elucidation of the mechanism of the enhanced dislocation glide is imperative.

To solve this problem, the present work employed light illumination to induce the enhancement of the 30°-Si(g) partial dislocation glide in 4H-SiC. We show that the direct photoionization of the dislocations accounts for the glide enhancement, ruling out the previously speculated REDR mechanism. Generally, the dislocation glide velocity depends on the thermodynamic driving force to exhibit REDG. In the case of 4H-SiC, we proposed that trapping of excitons by quantum-well-like potentials associated with Shockley stacking faults causes the reduction of the effective formation energy of the faults, and hence, facilitates their expansion achieved by the enhanced glide of 30°-Si(g) partial dislocation.[10,11] Such a thermodynamic driving force caused by light illumination may be given by $\{\beta\rho(I) - \gamma_{SF}\}$, where β denotes the efficiency factor, $\rho(I)$ is the density of free excitons trapped by stacking faults, which increases with the excitation intensity I, and γ_{SF} is the thermoequilibrium formation energy of a stacking fault of unit area.[10,11] If we further assume that $\beta\rho(I) \gg \gamma_{SF}$, i.e., $\{\beta\rho(I) - \gamma_{SF}\} \approx \beta\rho(I)$, and consider that $\rho(I) \propto I_{PL}$ where I_{PL} is the PL intensity from the stacking faults, we can resolve the velocity of the 30°-Si(g) partial dislocation glide V into two factors,

$$V \propto \{\beta\rho(I) - \gamma_{SF}\} \times R(f(I)) \approx I_{PL} \times R(f(I)). \quad (1)$$

Here, $R(f(I))$ represents the enhanced rate of dislocation glide, which is also a function of I. The function $f(I)$ would be the photoexcited carrier density if REDR is the dominant origin of the REDG. In contrast, $f(I)$ would be the light intensity if the photoionization of partial dislocations is the direct cause of REDG. In the present study, we conducted photoluminescence (PL) mapping experiments to deduce the relative magnitude of $R(f(I))$ considering $R(f(I)) \propto V/I_{PL}$ from eq. (1).

Dislocations situated within 4 µm from the illuminated surface were examined. The photoexcited carrier density within this range was varied by changing the illumination intensity. However, because the diffusion length of carriers is much longer than the depth range of 4 µm,[12] all the dislocations were surrounded by practically the same photocarrier concentration irrespective of the depth. In contrast, even for the fixed illumination intensity at the surface, the effective light intensity at each dislocation should change substantially as a function of the depth due to the large absorption coefficient. Therefore, while practically keeping the photocarrier concentration constant, the effective light intensity for the direct excitation of dislocation states could be varied with the depth. The present experiments conducted under such conditions have led us to a conclusion that $R(f(I))$ depends more strongly on the effective light intensity than on the photoexcited carrier density.

An 8° off-axis (0001) nitrogen-doped 4H-SiC substrate with a 70-µm-thick epitaxial layer on top was employed in the present study. The nitrogen concentration in the epitaxial layer was $\sim 10^{14}$ cm^{-3}. The sources of dislocations were introduced intentionally by indenting the sample surface with a microhardness tester.

The expansion velocity of Shockley stacking faults was measured by room-temperature PL mapping with the 266 nm line of a CW quadruple Nd:YVO_4 laser as an excitation

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source. The laser was focused down to a spot of $\sim$2 µm diameter on the sample surface using a Cassegrain objective (18×). The sample heating by the laser illumination ($\leq$42 mW) was determined to be negligible based on the fitting of the free-exciton emission spectra to the theoretical Maxwell–Boltzmann shape.[10)]

The stacking-fault-related emission at 2.9 eV was selected using band-pass filters and its intensity mapping on the sample surface was obtained. First, the stacking fault was expanded to $\sim$2 mm in length from the indent by scanning the excitation laser in advance. Then, we repeated one-dimensional lateral sweepings over the sample with the excitation laser spot along the elongated direction of the stacking fault to induce its expansion and obtain a set of PL intensity profiles. From the evolution of these profiles, we obtained the 30°-Si(g) partial dislocation velocity and simultaneously the stacking-fault-related PL intensity. The laser sweeping speed of 100 µm/s was chosen to be sufficiently faster than partial dislocation velocities, so that the effective duration of illumination on the partial dislocation, which could be estimated by dividing the carrier diffusion length by the sweep speed, was practically constant over the investigated illumination intensities. The illumination intensity dependence of $R(I_0)$, where I_0 is the illumination intensity at the sample surface, was thus measured for 30°-Si(g) partial dislocations at various depths.

A schematic of PL mapping of Shockley stacking faults is shown in Fig. 1(a). The depth z of 30°-Si(g) partial dislocation was calculated from the distance x between the intersection of the stacking fault with the sample surface and the laser sweeping path as shown in Fig. 1(a). The surface illumination intensity dependence of $R(I_0)$ for the 30°-Si(g) dislocation was determined using dislocations situated at $z \sim 1.0$, 1.7, and 3.9 µm as shown in Fig. 1(b). A clear variation of $R(I_0)$ was observed among 30°-Si(g) partial dislocations at the different depths.

The experimental results have been analyzed using two models assuming either photoexcited carrier recombination, $f(I) = \Delta n(I_0, z)$, or direct photoionization, $f(I) = I(I_0, z)$.

The first model assumes that the energy released on nonradiative recombination of electrons and holes via dislocation levels is converted to the kinetic energy of the dislocation glide motion.[7,8)] Here, the density of photogenerated electron–hole pairs $\Delta n(I_0, z)$, which would control the recombination rate and therefore the enhanced rate $R(\Delta n(I_0, z))$, is derived from the carrier diffusion equation, and is given by

$$\Delta n(I_0, z) = \frac{I_0 Q \tau}{1-(\alpha\lambda)^2}\left[e^{-\alpha z} - \left(\frac{\tau/\tau_s + \lambda\alpha}{\tau/\tau_s + 1}\right)e^{-z/\lambda}\right], \quad (2)$$

where Q is the electron–hole pair generation rate at the sample surface, τ the carrier lifetime, α the absorption coefficient, and λ the carrier diffusion length. The parameter $\tau_s \equiv \lambda/S$, where S is the surface recombination velocity. Using reported values of $\alpha = 0.833\,\mu\text{m}^{-1}$,[13)] $\lambda = 26\,\mu\text{m}$,[12)] $\tau = 2\,\mu\text{s}$,[14)] and $S = 5 \times 10^3$ cm/s,[14)] we find that the second term in eq. (2) is dominant and practically constant over the depths investigated in the present study.

The second model is based on the direct photoionization, taking into account the light absorption expressed by $I(I_0, z) = I_0 \exp(-\alpha z)$, where $I(I_0, z)$ denotes the effective light intensity at depth z.

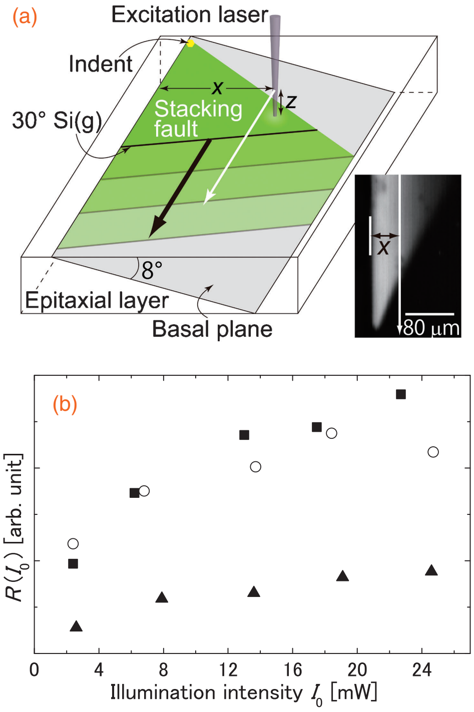

Fig. 1. (a) A schematic illustration of PL profiling experiments to measure the velocity of the 30°-Si(g) partial dislocation bounding a stacking fault. The laser scanning is performed along the white arrow. The depth z of a 30°-Si(g) partial dislocation from the sample surface was calculated from x in the illustration and the PL map, whose example is shown in the inset. The white area in the inset indicates the stacking fault. (b) Enhancement factor of the dislocation glide $R(I_0)$ vs illumination intensity I_0 measured at the sample surface. Data are shown for three different dislocations situated at $z \sim 1.0$ µm (■), $z \sim 1.7$ µm (○), and $z \sim 3.9$ µm (▲) from the sample surface.

Figures 2(a) and 2(b) show the dependences of $R(f(I))$ on the photoexcited carrier density $\Delta n(I_0, z)$ and effective light intensity $I(I_0, z)$, respectively. The plots of $R(\Delta n(I_0, z))$ vs $\Delta n(I_0, z)$ in Fig. 2(a) are not greatly different from those in Fig. 1(b) due to the large carrier diffusion length, which indicates that the enhanced rate $R(f(I))$ is *not* controlled by the photoexcited carrier density and, hence, the recombination-enhancement mechanism. In contrast, the plots of $R(I(I_0, z))$ vs $I(I_0, z)$ shown in Fig. 2(b) share a common curve. This suggests that the glide enhancement of partial dislocations is induced by the direct excitation of dislocations to different charge states. Eberlein and his coauthors investigated in their *ab initio* calculations the effect of charge on the migration barrier of kinks on 90° partial dislocations in 3C-SiC.[9)] They found that the kink migration energy of 90°-Si(g) partials is reduced dramatically by the trapping of holes to the kinks. If there is also a similar effect of charge on kinks in 30°-Si(g) partial dislocations, light illumination may cause photoionization of the kinks to reduce their migration barrier.

This photoionization model can qualitatively account also for the features of the common master curve in Fig. 2(b), which is characterized by the monotonic increase with effective light intensity and a tendency to saturation. In this model, the magnitude of R is explicitly proportional to the temporal fraction of the kinks in the ionized state, which is expressed by

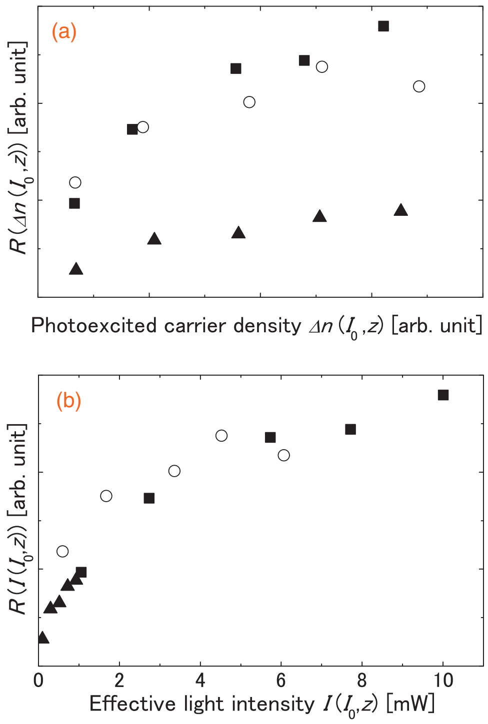

Fig. 2. Enhancement factor of the dislocation glide $R(f(I))$ vs (a) photoexcited carrier density $\Delta n(I_0, z)$ and (b) effective light intensity $I(I_0, z)$ for three different dislocations situated at $z \sim 1.0\,\mu\mathrm{m}$ (■), $z \sim 1.7\,\mu\mathrm{m}$ (○), and $z \sim 3.9\,\mu\mathrm{m}$ (▲) from the sample surface.

$$R(I(I_0, z)) \propto \frac{cI(I_0, z)}{1/\tau_\mathrm{d} + cI(I_0, z)}, \tag{3}$$

where τ_d denotes the neutralization lifetime and c represents the photoionization efficiency. Equation (3) is simplified to $R(I(I_0, z)) \propto I(I_0, z)$ when $I(I_0, z) \ll (1/c\tau_\mathrm{d})$ and becomes constant when $I(I_0, z) \gg (1/c\tau_\mathrm{d})$. This is in good agreement with the experimental results in Fig. 2(b), that is, $R(I(I_0, z))$ linearly increases at low effective light intensities and tends to be saturated at high intensities. Therefore, the photoionization of dislocation is the mechanism more consistent with the REDG in 4H-SiC. Our results also suggest that the effect of dislocation ionization may be responsible for the degradation of 4H-SiC bipolar devices induced by forward biasing. From the viewpoint of basic defect physics, it is important to test whether the ionization mechanism is responsible for REDG in other materials (Si, GaAs, GaN, etc.).

In summary, the effect of light illumination on the radiation-enhanced dislocation glide in 4H-SiC has been investigated by conducting photoluminescence experiments. The results showed that the enhanced rate of a 30°-Si(g) partial dislocation glide is controlled by the effective light intensity at the dislocations rather than by the photoexcited carrier density. These facts disfavor the frequently speculated recombination enhancement mechanism and suggest that the glide enhancement is induced by the photoionization of dislocations.

1) J. P. Bergman, H. Lendenmann, P. Å. Nilsson, U. Lindefelt, and P. Skytt: Mater. Sci. Forum **353–356** (2001) 299.
2) H. Lendenmann, F. Dahlquist, N. Johansson, R. Söderholm, P. Å. Nilsson, J. P. Bergman, and P. Skytt: Mater. Sci. Forum **353–356** (2001) 727.
3) M. Skowronski, J. Q. Liu, W. M. Vetter, M. Dudley, C. Hallin, and H. Lendenmann: J. Appl. Phys. **92** (2002) 4699.
4) R. E. Stahlbush, M. Fatemi, J. B. Fedison, S. D. Arthur, L. B. Rowland, and S. Wang: J. Electron. Mater. **31** (2002) 370.
5) K. Maeda and S. Takeuchi: *Dislocations in Solids* (North-Holland, Amsterdam, 1996) Vol. 10, p. 443.
6) N. Hoshino, M. Tajima, M. Naitoh, E. Okuno, and S. Onda: Mater. Sci. Forum **600–603** (2009) 349.
7) J. D. Weeks, J. C. Tully, and L. C. Kimerling: Phys. Rev. B **12** (1975) 3286.
8) H. Sumi: Phys. Rev. B **29** (1984) 4616.
9) T. A. G. Eberlein, R. Jones, A. T. Blumenau, S. Öberg, and P. R. Briddon: Appl. Phys. Lett. **88** (2006) 082113.
10) R. Hirano, Y. Sato, M. Tajima, K. M. Itoh, and K. Maeda: Mater. Sci. Forum **717–720** (2012) 395.
11) K. Maeda, R. Hirano, Y. Sato, and M. Tajima: Mater. Sci. Forum **725** (2012) 35.
12) Y. Shishkin, R. P. Devaty, and W. J. Choyke: Mater. Sci. Forum **338–342** (2000) 679.
13) S. G. Sridhara, T. J. Eperjesi, R. P. Devaty, and W. J. Choyke: Mater. Sci. Eng. B **61–62** (1999) 229.
14) A. Galeckas, J. Linnros, M. Frischholz, and V. Grivickas: Appl. Phys. Lett. **79** (2001) 365.