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Single Shockley Stacking Fault Expansion from Immobile Basal Plane Dislocations in 4H-SiC

Johji Nishio, Aoi Okada, Chiharu Ota and Ryosuke Iijima

Corporate Research & Development Center, Toshiba Corp.
1 Komukai-Toshiba-cho, Saiwai-ku, Kawasaki 212-8582, Japan
Phone: +81-44-549-2142 E-mail: johji.nishio@toshiba.co.jp

Abstract

Some combinations of partial dislocations that constitute basal plane dislocations have not previously been considered as sources for single Shockley stacking faults because they are regarded as immobile. We show the possibility of immobile C-core partial dislocations being converted to mobile Si-core partial dislocations. A model is proposed from a dynamic viewpoint for interpreting the mechanism by step motion during crystal growth in 4H-SiC.

1. Introduction

Forward voltage degradation observed mainly in bipolar devices and even in MOSFETs fabricated on 4H-SiC which have a *p-n* junction in the structure is widely known to be caused by the expansion of single Shockley stacking faults (1SSFs) in the epitaxial layer which originate from the basal plane dislocations (BPDs). It is thought that this expansion is not possible when the BPDs consist of 90 ^{o} C-core and 30 ^{o} C-core partial dislocations [1].

Evidence for the existence of 90^{o} C-core partial dislocation was found experimentally at a curved BPD by plan-view transmission electron microscope (TEM) and cross-sectional high-angle annular dark-field scanning TEM (HAADF-STEM) imaging [2]. This implies that BPDs that consist of 90 ^{o} C-core and 30 ^{o} C-core partial dislocations might be found around the substrate/epilayer interface. Furthermore, we have also found 1SSFs that have expanded from the curved BPDs in reality.

This presentation provides detailed results of structural analysis of the 1SSFs at their region of origin and attempts to provide an interpretation for why BPDs regarded as immobile were able to expand.

2. Experimental

To obtain electroluminescence (EL) images of 1SSF expansion under forward current injection, PiN diodes were fabricated on a commercially available *n*-type 4H-SiC substrate with 4 ^{o} off-cut toward the $[11\bar{2}0]$ direction. The *p*-type region was formed by aluminum ion implantation. Line-and-space-shaped open window metal contacts were used for the anode electrode. Reliability testing was performed on a solder-mounted directly bonded copper (DBC) plate [3]. After taking the diode chip away from the DBC plate, the electrode metal and passivation film were removed chemically. The 1SSFs in the chip were then analyzed by synchrotron x-ray topography and photoluminescence imaging to obtain the precise shape and the location for focused ion beam (FIB) processing in plan-view TEM sampling. $\boldsymbol{g}\cdot\boldsymbol{b}$ analysis was carried out by using selected-area electron diffraction to obtain dark-field images. Cross-sectional analysis was then made possible using the locational information obtained by the plan-view TEM imaging. The appropriate positions for FIB sampling were chosen accordingly. Cross-sectional BF-STEM imaging was carried out to ensure that the 1SSF edge was captured. Continuous enlargement by HAADF-STEM was performed to enabled high-resolution atomic imaging for analyzing the partial dislocation structure.

3. Results and Discussion

We selected a suitable 1SSF for analysis by the EL experiment. Figure 1 shows a synchrotron x-ray topograph of the 1SSF. The leftmost part was chosen for plan-view TEM

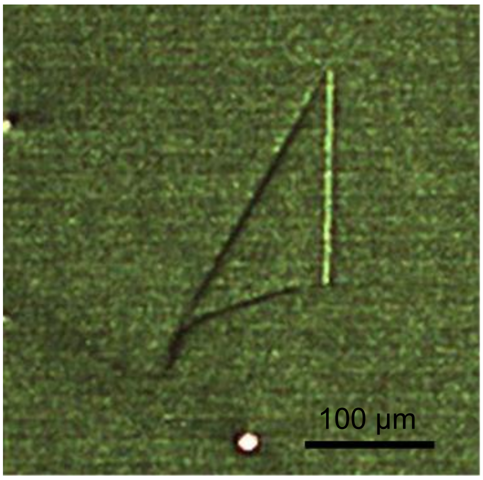

Fig. 1. Synchrotron x-ray topograph taken in Berg-Barrett configuration with $\boldsymbol{g} = 11\bar{2}8$ diffraction.

analysis, with the TEM images shown in Fig. 2. Due to the poor contrast of the image, the broken lines were added to clarify the dislocation lines. A corresponding cross-sectional schematic diagram is also shown in Fig. 2 to indicate the position of the substrate/epilayer interface. From Fig. 2, it should be noted that the partial dislocation lines are almost parallel in the substrate but diverge in the epilayer. Furthermore, in the substrate region, the parallel lines in the $[11\bar{2}0]$ direction are observed to turn toward the $[\bar{1}2\bar{1}0]$ direction. The turning of the dislocation is possible when the movement of two-dimensional nucleation steps formed during the sublimation growth of the substrate is taken into account. From the $\boldsymbol{g}\cdot\boldsymbol{b}$ analysis, the upper and the lower dislocation lines in

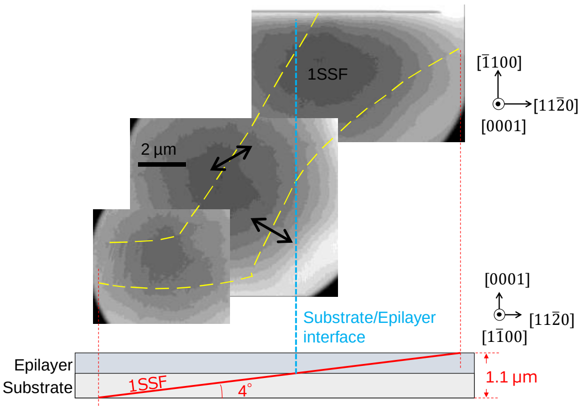

Fig. 2. Plan-view TEM images, the results of $\boldsymbol{g}\cdot\boldsymbol{b}$ analysis, and schematic diagram of the cross-section of the sample.

Fig. 2 were found to have Burgers vectors of $\pm 1/3\ [0\bar{1}10]$ and $\pm 1/3\ [10\bar{1}0]$, respectively. The Burgers vectors indicated by the arrows in Fig. 2 were found to be continuous before and after the inflection points. This implies a single dislocation loop. Figure 3 shows the HAADF-STEM results. The sampling positions and the observation directions are also shown in Fig. 3. Silicon atoms are indicated by the red circles that form a Burgers circuit, and tetrahedrons are depicted by triangles. At Point A in Fig. 3, the number of atoms in the upper line is smaller by one than that in the lower line. This means that an extra-half plane exists in the lower part, and the partial dislocation has a C-core structure. In contrast, at Point B in Fig. 3, the number of atoms in the upper and lower lines is the same. However, tetrahedra are face-to-face at Point A and back-to-back at Point B. According to the recent analysis, when the former is a C-core partial dislocation, it is expected to be a 30 ° dislocation, and the latter corresponds to a 90 ° C-core partial dislocation [4]. With the aid of finish-to-start with right-hand definition in the perfect crystal convention, the Burgers vectors were derived as $1/3[01\bar{1}0]$ or $1/3[1\bar{1}00]$ at Point A, and uniquely $1/3[10\bar{1}0]$ at Point B. Considering also the previous results shown in Fig. 2, the Burgers vectors are determined to be $1/3[01\bar{1}0]$ at Point A, and $1/3[10\bar{1}0]$ at Point B. This suggests that the BPD consisting of these partial dislocations is one that has not been considered to expand [1].

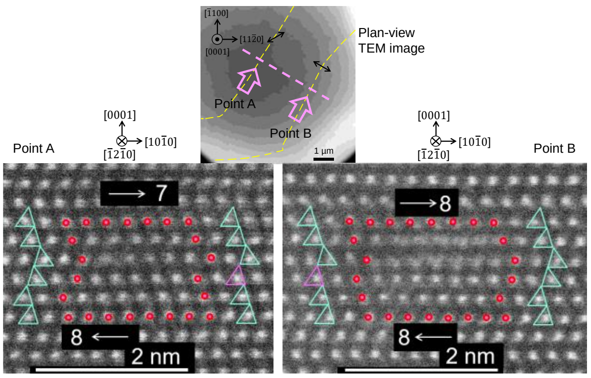

Fig. 3. HAADF-STEM images taken at Point A and Point B.

From the overall estimation of Burgers vectors and core species for the partial dislocations, the results are summarized as shown in Fig.4. The BPD that was regarded immobile expanded in reality, as observed in Fig. 1. This might be explained by the BPD being deep in the substrate, meaning that it is on the left side of the inflection points. The upper partial dislocation is thought to be a 30 ° Si-core, because it has the same Burgers vector as the 30 ° C-core next to it with a direction of [11$\bar{2}$0]. Conversion from a 30 ° Si-core to a 30 ° C-core is possible as explained previously due to the step motion during sublimation growth. Therefore, the model we propose here is a conversion from a 30 ° C-core to a 30 ° Si-core, possibly induced by the [11$\bar{2}$0] step flow in the very early stages of the epitaxial growth process. The inclining feature at the 90 ° C-core partial dislocation after penetration into the epilayer, observed in Fig. 2, may support this idea. The curved C-core partial dislocation was found to be mixed with 30 ° and 90 ° C-core partial dislocations [2].

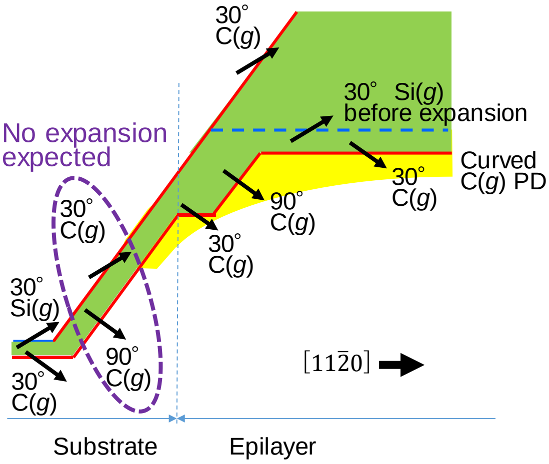

Fig. 4. Summary of the results obtained from the current study and a proposed model of 1SSF expansion from deemed immobile BPD.

3. Conclusions

A triangular shaped 1SSF with a curved bottom line was selected and its initiation region was structurally analyzed. The original BPD was found to be composed of 30 ° and 90 ° C-core partial dislocations that were regarded as immobile. A possible mechanism is proposed based on the conversion of the core species by the step flow during the epitaxial growth of 4H-SiC.

References

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