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Growth of Nitrogen-Polar 2H-AlN on Step-Height-Controlled 6H-SiC (000-1) Substrate by Molecular-Beam Epitaxy

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1. Introduction

AlGaN/GaN high electron mobility transistors (HEMTs) are promising for high-power and high-frequency electronic devices due to their high electron saturation velocity and high breakdown field. Most of AlGaN/GaN HEMTs have been fabricated on the (0001) or metal-polar orientation, which is superior in terms of crystalline quality. Recently, investigations on the $(000\bar{1})$ or nitrogen-polar (N-polar) orientation have attracted a lot of interest due to strong electron confinement and low contact resistance [1]. For the N-polar GaN-based devices, establishment of high-quality growth on a N-polar surface is a key issue.

For the AlGaN/GaN HEMTs, SiC is one of the most suitable substrates due to its high thermal conductivity. Polarity of the GaN layer can be selected by polarity of SiC substrate, i.e., SiC (0001) and $(000\bar{1})$. Compared with the growth of Ga-polar GaN on SiC (0001), the N-polar GaN growth on SiC $(000\bar{1})$ has not been extensively studied. Crystalline quality and surface flatness of the N-polar GaN layers have much room for improvement.

For GaN growth on SiC, an AlN buffer layer is generally used. As one of approaches to improve crystalline quality of the GaN layer, we have investigated high-quality AlN growth exploiting the small lattice mismatch (1%) between AlN and SiC. For the high-quality AlN growth on SiC, issues originating from the polytype difference must be solved; AlN has wurtzite (2H) structure (AB AB…), but the available SiC wafers have a 6H (ABCACB ABCACB…) or 4H (ABCB ABCB…) structure. In 2H-AlN growth on SiC, stacking mismatch boundaries (SMBs) are formed at the step edges of 6H- and 4H-SiC substrates due to the polytype difference [2,3]. Formation of SMBs is expected to be avoided by controlling the step heights of the SiC substrates to full-unit-cell (6 bilayer for 6H-SiC) and realizing layer-by-layer growth mode because the stacking arrangements of the AlN layers on each terrace correspond.

In Al-polar AlN growth on the 6H-SiC (0001) substrates, we have already realized high-quality AlN layers without SMBs due to an enhancement of layer-by-layer growth mode and the step-height control for the substrate [4,5]. Threading dislocation density (TDD) in the Al-polar AlN layer was reduced to 4×10^{8} cm^{-2}. However, there is less report on N-polar AlN growth on SiC [6]. In this study, we investigated N-polar AlN growth on 6H-SiC $(000\bar{1})$ with 6-bilayer-height steps.

2. Experimental Procedures

Carbon-face 6H-SiC $(000\bar{1})$ (vicinal angle 0.2°) substrates were used. The SiC substrates were treated by chemical mechanical polishing (CMP) followed by HCl/H_2 gas etching to control the SiC step height to be 6 bilayers, as shown in Fig. 1(a). The gas etching was carried out at 1300°C and atmosphere pressure for 30 min in a horizontal cold-wall chemical-vapor-deposition (CVD) reactor.

Before growth, the residual oxygen on the substrates was removed by *in-situ* Ga deposition and desorption processes. After that, the SiC surface was exposed to a 2ML-Ga beam as soon as nitrogen plasma ignited [5]. A 200-nm-thick AlN layer was grown at 850°C by using elemental Al and active nitrogen generated by an Veeco Unibulb rf-plasma cell. Nitrogen-plasma power and N_2 flow rate are 160 W and 0.75 sccm, respectively.

Surface morphologies, growth mode, and defect structures of the AlN layer were characterized by atomic force microscopy (AFM), reflection-high energy-electron diffraction (RHEED), and transmission electron microscopy (TEM) with JEM-2100F operated at 200 kV, respectively.

3. Results and Discussion

A surface morphology of 200-nm-thick AlN layer is shown in Fig. 1(b). There were hillocks with density of 5×10^{8} cm^{-2}. The hillocks were resulted from not spiral (screw dislocation) but multiple two-dimensional nucleation. Between the hillocks, parallel and long grooves existed, corresponding to a step direction of the SiC surface before growth. We believe that the multiple nucleation was formed on each terrace of the SiC substrates, resulting in the hillocks and the grooves.

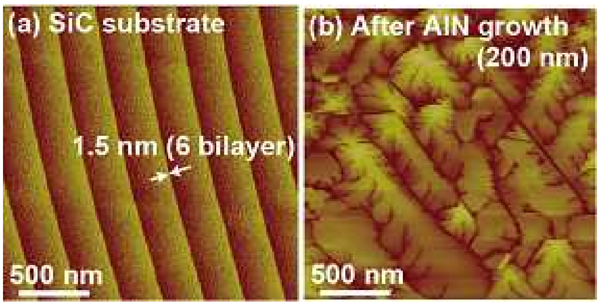

Fig. 1: Surface morphologies of (a) 6H-SiC (000-1) substrate after gas etching and (b) 200-nm thick AlN layer grown on the substrate.

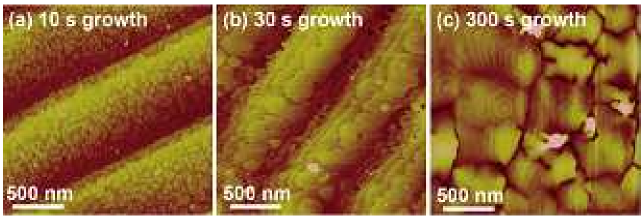

Fig. 2: Surface morphologies of thin AlN grown on 6H-SiC $(000\bar{1})$ substrates for (a) 10 s, (b) 30 s, and (c) 300 s.

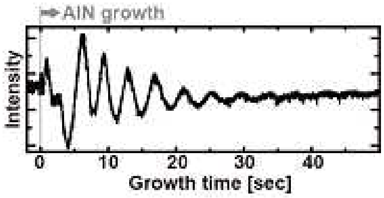

Fig. 3: Growth time dependence of RHEED intensity of AlN layer on 6H-SiC $(000\bar{1})$ substrate.

Surface morphologies of AlN layers at an initial growth stage are shown in Fig. 2. Although AlN layers had layer-by-layer growth mode until 30-seconds growth, vertical growth toward a $[000\bar{1}]$ direction was gradually preferred to lateral growth so that AlN layers formed multiple nucleation or hillocks, as shown in Fig. 2(c). This result corresponds to a RHEED intensity profile at the initial stage of AlN growth, as shown in Fig. 3. The RHEED oscillations were observed for 30 seconds just after growth, indicating layer-by-layer growth mode. After that, the surface flatness and the intensity of RHEED oscillations decreased. We believe that the small Al migration length in the N-polar AlN layers enhances multiple nucleation. To obtain the larger migration length, we tried to increase growth temperature and to reduce supersaturation by decreasing nitrogen-plasma power, but both ways were less effective. Based on the above results, we suggest that AlN should be grown on SiC $(000\bar{1})$ substrates with larger off angle (around 0.5°), of which the terraces are enough narrow for step-flow growth mode, for the further smooth surface.

The FWHMs for (0002) and $(01\bar{1}2)$ ω-scan diffraction peaks of the 200-nm thick AlN layer were 140 arcsec and 550 arcsec, respectively. Effects of Ga pre-deposition on crystalline quality of AlN layers were not observed for the N-polar AlN growth on SiC $(000\bar{1})$, unlike an Al-polar AlN growth on SiC (0001) [5].

The AlN layer was confirmed to have N polarity by convergent-beam electron diffraction (CBED) pattern. A bright-field plan-view TEM image of the N-polar AlN layer is shown in Fig. 4. Dark spots are threading dislocations (TDs). No SMB is obtained indicating step-height control of SiC substrate is effective to eliminate SMB in growth on C-face SiC. The type of the TDs was found to be pure edge by a cross-sectional TEM image with two-beam condition. The TDs are arranged in rows with a separation of around 200 nm. The separation of TD rows closely corresponds to the terrace width of the SiC substrate before growth, indicating that most of the TDs are generated at the step edges of the SiC substrate. The TDD was 2×10^9 cm^{-2}. For further reduction of TDs in N-polar AlN layers, dislocation formation mechanism at the step edges of SiC substrate as well as at the terraces of SiC substrate must be clarified.

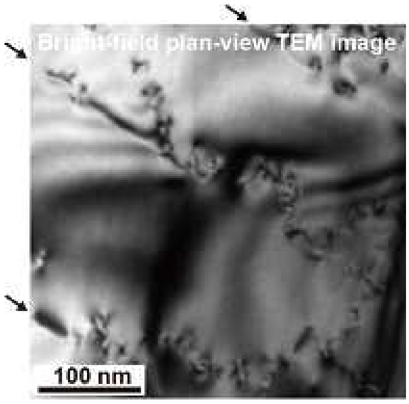

Fig. 4: Bright-field plan-view TEM image (zone axis) of N-polar AlN grown on 6H-SiC $(000\bar{1})$ substrate.

4. Conclusions

N-polar AlN layers were grown on 6H-SiC $(000\bar{1})$ substrates with 6-bilayer-height steps. Formation of SMBs was successfully suppressed by step-height control of SiC substrate and initial layer-by-layer growth. In the N-polar AlN growth, multiple nucleation occurred easily maybe due to the small Al migration length, increasing the surface roughness. TDD in the AlN layers was 2×10^9 cm^{-2}. The most of TDs were generated at the step edges of the SiC surfaces.

Acknowledgements

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