Title
Defect reduction in nonpolar a-plane GaN films using in situ SiN$_x$ nanomask

Permalink
https://escholarship.org/uc/item/8z91c0hc

Journal
Applied Physics Letters, 89(4)

ISSN
0003-6951

Authors
Chakraborty, A
Kim, K C
Wu, F
et al.

Publication Date
2006-07-01

Peer reviewed
Defect reduction in nonpolar a-plane GaN films using in situ SiN<sub>x</sub> nanomask

Electrical and Computer Engineering and Materials Departments, College of Engineering, University of California, Santa Barbara, California 93106

(Received 5 March 2006; accepted 3 June 2006; published online 25 July 2006)

We report on the use of in-situ SiN<sub>x</sub> nanomask for defect reduction in nonpolar a-plane GaN films, grown by metal-organic chemical vapor deposition. High-resolution x-ray diffraction analysis revealed that there was a monotonic reduction in the full width at half maximum, both on-axis and off-axis, with the increase in the SiN<sub>x</sub> thickness. Atomic force microscopy images revealed a significant decrease in the root-mean-square roughness and the density of submicron pits. Cross-section and plan-view transmission electron microscopy on the samples showed that the stacking fault density decreased from 8 × 10<sup>5</sup> to 3 × 10<sup>5</sup> cm<sup>-2</sup> and threading dislocation density decreased from 8 × 10<sup>10</sup> to 9 × 10<sup>9</sup> cm<sup>-2</sup>. Room temperature photoluminescence measurement revealed that the band-edge emission intensity increased with the insertion of the SiN<sub>x</sub> layer, which suggests reduction in the nonradiative recombination centers. © 2006 American Institute of Physics.

[DOI: 10.1063/1.2234841]

Recently, there has been considerable interest in the growth of nonpolar gallium nitride based epitaxial films, heterostructures, and devices.1–5 Unlike conventional c-plane quantum wells, nonpolar (a- and m-plane) quantum wells are free of polarization-related electric fields along the growth direction.6,7 However, heteroepitaxially grown planar a-plane GaN films are characterized by the presence of high dislocation density due to the large lattice mismatch with the available substrates.2 The high defect density in planar films limits device performance because threading dislocations act as nonradiative recombination centers and scattering centers and reduces radiative recombination efficiency, carrier mobility, and effect reliability. Therefore, improving the structural quality of nonpolar GaN films is pivotal to achieving high-performance devices. Lateral epitaxial overgrowth (LEO) techniques have been employed in the past for achieving defect reduction in nonpolar GaN.8–10 However, all of these techniques involve ex-situ processing steps and regrowths. The use of in-situ SiN<sub>x</sub> interlayer has proven to be an effective technique in defect reduction in conventional c-plane GaN.11–13 In this letter, we report on the use of in-situ SiN<sub>x</sub> nanomask for defect reduction in a-plane GaN films. The simplicity of an in-situ defect reduction growth method is highly attractive. Unlike traditional single-step LEO, the reduced feature size of the SiN<sub>x</sub> nanopores facilitates nanometer-scale LEO, thereby resulting in defect reduction uniformly across the wafer.

The a-plane GaN films described herein were grown by metal-organic chemical vapor deposition, and the sample structure consisted of a low temperature GaN nucleation layer grown on m-plane sapphire substrate, followed by the growth of ~0.7 μm high temperature GaN. Trimethylgallium, ammonia (NH<sub>3</sub>), and disilane (Si<sub>2</sub>H<sub>6</sub>) were used as Ga, N, and Si sources, respectively. Then a thin layer of SiN<sub>x</sub> was inserted by flowing disilane and ammonia. The thickness of the SiN<sub>x</sub> interlayer was controlled by varying the growth time. The SiN<sub>x</sub> layer was followed by the growth of a ~0.1 μm thick unintentionally doped GaN layer and capped with a ~2 μm thick Si-doped GaN layer. The Si concentration ([Si] ~ 7 × 10<sup>19</sup> cm<sup>-3</sup>) in the Si-doped GaN layer was much higher than the compensation level (~1 × 10<sup>18</sup> cm<sup>-3</sup>) (Ref. 14). Secondary-ion-mass-spectroscopy (SIMS) measurement revealed that the Si-doping profile in the Si-doped GaN layer was not affected by the thickness of the SiN<sub>x</sub> layer. After the growth, the surface morphology of the as-grown samples was studied by means of Nomarski-mode optical microscopy and atomic force microscopy (AFM). A Digital Instruments D3000 AFM was used in the tapping mode to image the surface of the samples. The crystalline quality and the crystal mosaic of the as-grown films were determined using a Philips four-circle MRD x-ray diffractometer using Cu Kα radiation operating in receiving slit mode with a four bounce Ge (220) monochromator and a 1.2 mm slit on the detector arm. Transmission electron microscopy (TEM) was used to correlate the x-ray measurements to the microstructure of a-plane GaN grown with and without SiN<sub>x</sub> interlayer. [1100] cross-section and plan-view samples were prepared with a FEI focused ion beam instrument (model DB235 dual beam). Two beam diffraction contrast bright field and dark field images were recorded using an FEI Tecnai G2 Sphera Microscope, operated at 200 kV. Room temperature photoluminescence (PL) measurements were carried out using the 325 nm line of a cw He–Cd laser with an excitation power density of ~20 W/cm<sup>2</sup>.

The Nomarski image of a fully coalesced a-plane GaN film with 120 s of SiN<sub>x</sub> interlayer revealed smooth and uniform surface with a few occasional pits formed from the coalescence edge. 5 × 5 μm<sup>2</sup> AFM images of samples with and without SiN<sub>x</sub> nanomask are shown in Fig. 1(a) and 1(b), respectively. There was a significant improvement in the surface morphology after the insertion of SiN<sub>x</sub> interlayer. There was a reduction in the density of submicron pits, and the rms roughness decreased from 2.6 to 0.6 nm.

Omega x-ray rocking curves (XRCs) were measured for both the GaN on-axis (110) and off-axis (100), (101), (201), and (102) reflections. The on-axis and off-axis XRCs of
samples grown with different SiN<sub>x</sub> growth time were measured and the full width at half maximum (FWHM) of the measurements are plotted in Fig. 2. The on-axis FWHMs of φ=0° and φ=90° (c- and m-mosaics, respectively) for GaN template without SiN<sub>x</sub> interlayer were 0.69° (1290°) and 0.36° (2471°), respectively. The (101) off-axis peak, which measures the “twist” mosaic, had a FWHM of 0.64° (2292°). These large FWHM values are in agreement with the high dislocation density typically observed in planar a-plane GaN. It can be seen that the on-axis and off-axis FWHM for all of the reflections decreased with the increase in the SiN<sub>x</sub> deposition time. This decrease signified dislocation reduction with the SiN<sub>x</sub> nanomasking. It was also noted that for the on-axis scan, the ratio of m mosaic to c mosaic approached unity with the increase in the SiN<sub>x</sub> growth time. The minimum XRC FWHMs were obtained for 150 s of SiN<sub>x</sub> deposition, and the on-axis values were 0.29° (1040°) and 0.25° (924°) for φ=0° and φ=90°, respectively. The off-axis values were 0.42° (1508°), 0.38° (1375°), and 0.33° (1208°) for (101), (201), and (102) reflections, respectively. However, the sample grown with 150 s of SiN<sub>x</sub> growth could not be coalesced completely after 2 μm of GaN overgrowth.

TEM was performed on samples with SiN<sub>x</sub> growth time of 0, 120, and 150 s. The cross-section image of the sample with 150 s of SiN<sub>x</sub> interlayer is shown in Fig. 3. From the cross-section image, it was observed that the threading dislocations (TDs) have a common line direction, parallel to the [1120] growth direction, for all the samples. Significant blocking of the extended defects was observed at the GaN–SiN<sub>x</sub> interface. Thus it was evident that dislocation reduction was indeed achieved by the insertion of the SiN<sub>x</sub> interlayer. In addition to the TDs, plan-view TEM on the samples revealed stacking faults (SFs) aligned perpendicular to the c-axis. The TD and the SF density for the samples were determined from the plan-view images, and the values are summarized in Table I. It is evident from the table that both TD and SF density decreased as a result of the SiN<sub>x</sub> nanomasking, which concurs with the XRC findings.

The room temperature PL intensity of the GaN band-edge improved with the SiN<sub>x</sub> nanomasking (shown in Fig. 4). The a-GaN sample without the SiN<sub>x</sub> interlayer did not show band-edge emission. However, with the increase in the SiN<sub>x</sub> thickness, PL emission intensity increased. Also, PL emission from AlGaN/GaN multiple-quantum wells, grown on the samples with different SiN<sub>x</sub> thicknesses, showed a similar reduction was indeed achieved by the insertion of the SiN<sub>x</sub> interlayer.

<table>
<thead>
<tr>
<th>SiN&lt;sub&gt;x&lt;/sub&gt; deposition time (s)</th>
<th>0</th>
<th>120</th>
<th>150</th>
</tr>
</thead>
<tbody>
<tr>
<td>TD Density (cm&lt;sup&gt;-2&lt;/sup&gt;)</td>
<td>6×10&lt;sup&gt;10&lt;/sup&gt;–8×10&lt;sup&gt;10&lt;/sup&gt;</td>
<td>1×10&lt;sup&gt;10&lt;/sup&gt;–3×10&lt;sup&gt;10&lt;/sup&gt;</td>
<td>9×10&lt;sup&gt;9&lt;/sup&gt;</td>
</tr>
<tr>
<td>SF density (cm&lt;sup&gt;-1&lt;/sup&gt;)</td>
<td>6×10&lt;sup&gt;5&lt;/sup&gt;–8×10&lt;sup&gt;5&lt;/sup&gt;</td>
<td>4×10&lt;sup&gt;5&lt;/sup&gt;</td>
<td>3×10&lt;sup&gt;5&lt;/sup&gt;</td>
</tr>
</tbody>
</table>
lar trend. The increased emission intensity is probably a conse-
quense of reduction in the TD density. The much increased
emission intensity from the sample with 150 s of SiN$_x$
is probably due to the increased light extraction from the un-
coalesced facets of the sample.

In conclusion, we have demonstrated the effectiveness of in-situ SiN$_x$ nanomask in defect reduction in $a$-plane GaN films grown on $r$-plane sapphire. The thickness of SiN$_x$ interlayer was varied by changing the growth time. The use of SiN$_x$ resulted in the reduction of on-axis and off-axis XRC FWHMs and also resulted in the reduction of rms roughness in AFM images. Cross-section and plan-view TEM micro-
graphs revealed reduction in defect density—both SFs and TDs. The intensity of band-edge luminescence improved due to the reduction in defect density.

The authors would like to acknowledge the support of SSLDC, JST/ERATO, and the National Science Foundation for use of facilities through the MRSEC program.

3H. M. Ng, Appl. Phys. Lett. 80, 4369 (2002).