



2009

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Ni, X., Wu, M., Lee, J., et al. Nonpolar m-plane GaN on patterned Si(112) substrates by metalorganic chemical vapor deposition. Applied Physics Letters, 95, 111102 (2009). Copyright © 2009 AIP Publishing LLC.

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# Nonpolar $m$ -plane GaN on patterned Si(112) substrates by metalorganic chemical vapor deposition

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(Received 30 July 2009; accepted 20 August 2009; published online 14 September 2009)

The concept of nonpolar  $(\bar{1}\bar{1}00)$   $m$ -plane GaN on Si substrates has been demonstrated by initiating growth on the vertical  $(\bar{1}\bar{1}1)$  sidewalls of patterned Si(112) substrates using metalorganic chemical vapor deposition. The Si(112) substrates were wet-etched to expose  $\{111\}$  planes using stripe-patterned  $\text{SiN}_x$  masks oriented along the  $[\bar{1}10]$  direction. Only the vertical Si $(\bar{1}\bar{1}1)$  sidewalls were allowed to participate in GaN growth by masking other Si $\{111\}$  planes using  $\text{SiO}_2$ , which led to  $m$ -plane GaN films. Growth initiating on the Si $(\bar{1}\bar{1}1)$  planes normal to the surface was allowed to advance laterally and also vertically toward full coalescence. InGaN double heterostructure active layers grown on these  $m$ -GaN templates on Si exhibited two times higher internal quantum efficiencies as compared to their  $c$ -plane counterparts at comparable carrier densities. These results demonstrate a promising method to obtain high-quality nonpolar  $m$ -GaN films on large area, inexpensive Si substrates. © 2009 American Institute of Physics. [doi:10.1063/1.3225157]

In  $c$ -plane nitride materials, the polarization-induced electric field causes spatial separation of electron and hole wave functions in quantum wells (QWs) used in light emitting diodes (LEDs) and laser diodes (LDs), thereby reducing quantum efficiency.<sup>1,2</sup> A blueshift in the emission wavelength also becomes inevitable with increasing injection current unless very thin QWs are employed. The use of nonpolar orientations, namely,  $m$ -plane<sup>3</sup> or  $a$ -plane GaN,<sup>4</sup> would overcome this problem. The  $m$ -plane orientation is especially well-suited for optoelectronic applications in that compared to  $c$ -plane it is reported to have a lower valence band density of states and reduced valence band effective mass, leading to higher hole concentrations and larger optical matrix elements which enhance radiative recombination efficiency and improve LED and LD performance.<sup>5</sup> So far,  $m$ -plane GaN has been obtained on  $\gamma$ -LiAlO<sub>2</sub> (100),<sup>6</sup>  $m$ -plane SiC substrates,<sup>7</sup> and  $m$ -plane bulk GaN,<sup>8,9</sup> which all have limited availability and/or high cost.

Silicon substrates are very attractive for the growth of GaN due to their high quality, good thermal conductivity, low cost, availability in large size, and ease with which they can be selectively removed before packaging for better light extraction and heat transfer when needed. Growth of  $c$ -plane GaN on Si has been explored extensively with considerable success; however, studies of nonpolar or semipolar GaN growth on Si have been limited. It is well known that hexagonal GaN(0001) grows on Si(111) with the following epitaxial relationships: GaN $\langle 0001 \rangle \parallel$  Si $\langle 111 \rangle$  and GaN $\langle 2\bar{1}\bar{1}0 \rangle \parallel$  Si $\langle 011 \rangle$ .<sup>10</sup> The growth of  $(11\bar{2}0)$   $a$ -plane GaN on Si(110),<sup>11</sup>  $(11\bar{2}2)$  GaN on Si(113),<sup>12,13</sup> and  $(1\bar{1}01)$  semipolar GaN on Si(001) (Refs. 14 and 15) has been achieved; however, studies of  $m$ -plane nitrides and related LEDs on Si substrates have not yet been reported. According to the epitaxial relationship between GaN and Si, it is possible to align

$m$ -plane GaN( $\bar{1}\bar{1}00$ ) with Si(112). In this case, growth must be initiated on Si(111) facets, which requires patterning of the Si substrate to expose the  $\{111\}$  facets and allow only  $(\bar{1}\bar{1}1)$  facets to participate in growth. In this letter, we describe the epitaxial growth of nonpolar,  $m$ -plane GaN on stripe-patterned Si(112) substrates using metalorganic chemical vapor deposition (MOCVD).

To obtain  $m$ -plane GaN parallel to the substrate Si(112) plane, the  $c$ -axis of GaN and, therefore, select Si $\{111\}$  directions should lie in the substrate plane. Si(112) substrates were therefore patterned prior to growth to expose vertical Si $(\bar{1}\bar{1}1)$  facets, as shown in Fig. 1. First, a 100-nm-thick  $\text{SiN}_x$  layer was grown by plasma-enhanced chemical vapor deposition (PECVD) on a Si(112) substrate,<sup>16</sup> and was then patterned to form a wet-etch mask having periodic stripes (4- $\mu\text{m}$ -wide mask plus 10- $\mu\text{m}$ -wide open window) along the Si $(1\bar{1}0)$  direction by using inductively coupled plasma etching after standard photolithography. Subsequent anisotropic wet-etching of Si was performed in a KOH solution (4.16M) at 40 °C to form Si $\{111\}$  facets. As shown in Fig. 1, after KOH etching two groups of Si $\{111\}$  planes would be formed, one being perpendicular to Si(112) and the other forming an

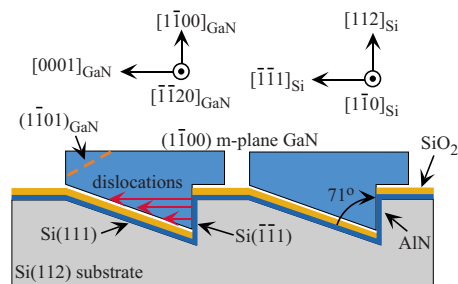


FIG. 1. (Color online) Schematic depiction of selective area  $m$ -plane GaN growth on patterned Si(112) substrates. Since growth is initiated on the vertical Si $(\bar{1}\bar{1}1)$  sidewalls, the threading dislocations propagate along the  $c$ -axis and not toward the surface.

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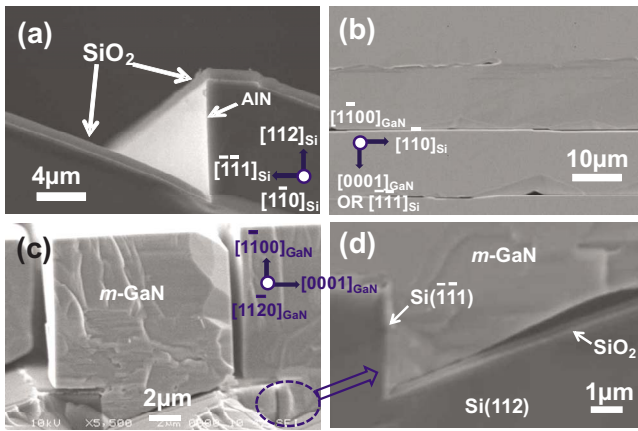


FIG. 2. (Color online) (a) An angled cross-sectional SEM image of a patterned Si(112) substrate ready for GaN epitaxy, where the vertical sidewall is covered by AlN. The top Si(112) plane and the tilted Si(111) plane are both masked with SiO<sub>2</sub>. (b) Plan-view and (c) angled cross-sectional SEM images of an *m*-GaN sample grown on a patterned Si(112) substrate after a 3 h growth. (d) Enlarged cross-sectional SEM image showing the growth of the *m*-GaN initiated at the vertical Si( $\bar{1}\bar{1}1$ ) sidewalls.

angle of 19° with the Si(112) surface. The vertical Si( $\bar{1}\bar{1}1$ ) sidewalls are used for the *m*-GaN growth and have a depth of approximately 4 μm.

Next, a 100-nm-thick AlN layer was grown by MOCVD on the patterned Si substrate at 1050 °C to serve as a seed layer for the GaN growth after the remaining SiN<sub>x</sub> etching mask was removed using buffered-oxide etch (BOE). To initiate growth only on the vertical Si( $\bar{1}\bar{1}1$ ) sidewalls and not on the Si(112) and inclined Si(111) surfaces, a SiO<sub>2</sub> mask was then deposited using PECVD and wet-etched in diluted BOE to expose only the vertical Si( $\bar{1}\bar{1}1$ ) sidewalls. Due to the relatively higher deposition rates of SiO<sub>2</sub> on the (112) and tilted (111) surfaces, only the SiO<sub>2</sub> film on the vertical ( $\bar{1}\bar{1}1$ ) sidewalls was completely etched away after 6 min in dilute BOE, as shown in Fig. 2(a). The Si substrate was then reloaded into the MOCVD chamber for GaN growth, which was initiated on the vertical Si( $\bar{1}\bar{1}1$ ) sidewalls and not on the Si(112) and inclined Si(111) surfaces. Growth advanced laterally first along the GaN [0001] *c*<sup>+</sup> direction and then additionally along the [000 $\bar{1}$ ] *c*<sup>-</sup> direction after the vertical growth advanced above the Si(112) substrate plane, resulting in a coalesced *m*-plane GaN surface after a sufficient amount of growth. The GaN growth temperature was 1030 °C and the chamber pressure was set at 30 Torr. Trimethylgallium (TMGa) and NH<sub>3</sub> flow rates were kept at 117 μmol/min and 550 (SCCM) (SCCM denotes standard cubic centimeters per minute at STP), respectively.

The plan-view and cross-sectional scanning electron microscopy (SEM) images of the *m*-plane GaN after 3 h of growth were shown in Fig. 2. According to the plan-view image in Fig. 2(b), the sample surface was partially coalesced due to an insufficient overgrowth time and also in part due to a relatively low ratio of lateral growth rate to vertical growth rate (1:1 in this case) under the nonoptimized GaN growth conditions employed. From the cross-sectional SEM image in Figs. 2(c) and 2(d), the GaN growth was confirmed to initiate only on the vertical Si( $\bar{1}\bar{1}1$ ) sidewalls. With further growth and coalescence, the facets and trenches observed in Fig. 2(b) at the meeting fronts are expected to gradually dis-

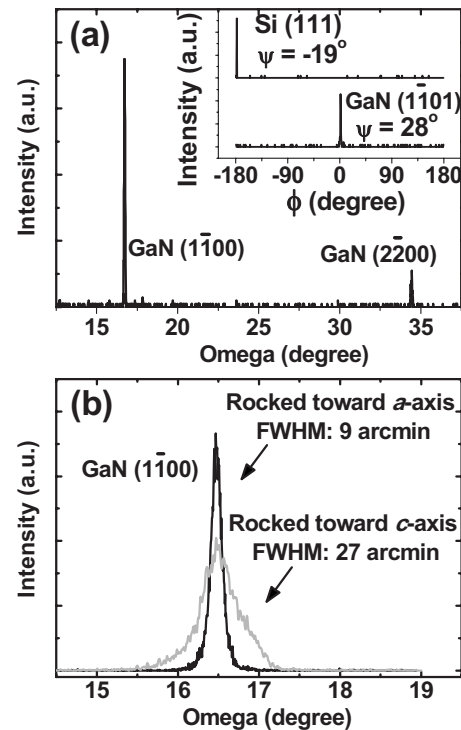


FIG. 3. (a) On-axis XRD  $\omega$ - $2\theta$  scan showing that the GaN has ( $\bar{1}\bar{1}00$ ) *m*-plane orientation. The Si(112) plane does not have a diffraction peak due to diffraction extinction. The inset shows the off-axis scan results, suggesting that GaN( $\bar{1}\bar{1}01$ ) is oriented 180° away from Si(111). (b) XRD rocking curves of GaN( $\bar{1}\bar{1}00$ ) *m*-plane after a 3 h growth, rocking toward the GaN *a*-axis and *c*-axis.

appear and result in a smooth *m*-plane GaN surface. Since the growth is initiated on the vertical Si( $\bar{1}\bar{1}1$ ) sidewalls, the threading dislocations are expected to propagate along the GaN *c*-axis, and not toward the surface.

In order to confirm the GaN orientation, on-axis x-ray diffraction (XRD)  $2\theta$ - $\omega$  scans were performed, as shown in Fig. 3(a). The XRD data show only GaN *m*-plane ( $\bar{1}\bar{1}00$ ) and ( $\bar{2}\bar{2}00$ ) diffraction peaks, while the Si(112) substrate does not appear due to a diffraction extinction effect. To determine the in-plane epitaxial relationship between the GaN and Si substrate, off-axis XRD scans were performed by changing the  $\phi$  and  $\psi$  angles, where  $\phi$  is the angle of rotation about the sample surface normal and  $\psi$  is the angle of tilt about the axis formed by the intersection of the Bragg and scattering planes. The results [see inset of Fig. 3(a)] suggest that GaN( $\bar{1}\bar{1}01$ ) is oriented 180° away from the tilted Si(111) in terms of  $\phi$  angles which conforms to the relationship depicted in the schematic of Fig. 1. To determine the crystalline quality of the grown *m*-plane GaN sample XRD rocking curves were measured. As shown in Fig. 3(b), the rocking curve full width at half maximum (FWHM) value is 9 arcmin when rocked toward the GaN *a*-axis and 27 arcmin when rocked toward the GaN *c*-axis. Because the GaN *c*-axis, which is perpendicular to the stripes, is the laterally advancing direction for the *m*-GaN growth, the larger FWHM value in the case of rocking toward the *c*-axis reflects broadening due to tilting of the advancing wings in the *c*<sup>+</sup> and *c*<sup>-</sup> directions. However, the individual XRD line-widths of the wings were most probably not sufficiently narrow to allow the determination of the tilt angle.

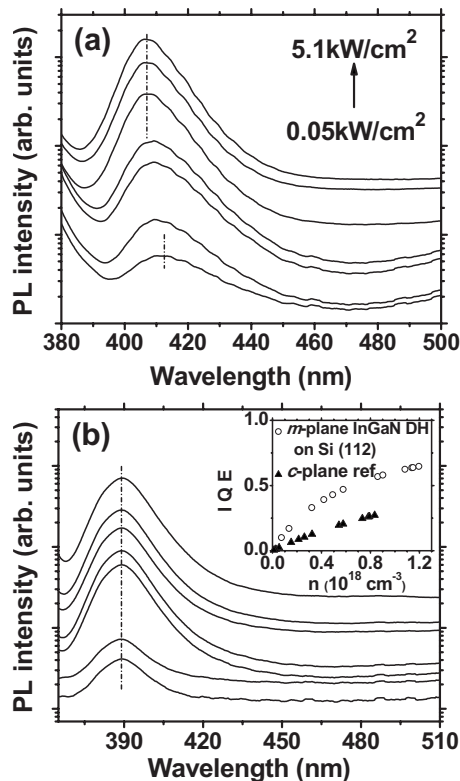


FIG. 4. Room-temperature PL spectra measured with a He–Cd laser at different excitation densities for 6-nm-thick InGaN DH LED active layers on (a) *c*-GaN on sapphire and (b) *m*-GaN on Si. No blueshift is observed in the emission peak of the *m*-plane sample. The excitation power densities in [(a) and (b)] were 0.05, 0.15, 0.52, 1.0, 2.0, 2.5, and 5.1 kW/cm<sup>2</sup>. The inset in (b) shows the IQEs of both samples extracted from the excitation dependence of the PL intensity using a Ti-sapphire laser (370 nm).

It should be mentioned that a very limited number of cracks separated by hundreds of microns were observed on the sample surface along the *c*-axis caused by the large difference of thermal expansion coefficient between GaN and Si. Using low-temperature AlN (Ref. 17) or AlN/GaN superlattices<sup>18</sup> as a buffer layer, the cracking in our case could be avoided. The surface of the *m*-plane GaN sample on Si was very smooth (rms roughness of  $\sim 0.3$  nm over an area of  $2 \times 2 \mu\text{m}^2$ ) and exhibited clear atomic steps, as revealed from atomic force microscopy measurements.

The room temperature band edge photoluminescence (PL) intensity measured using a HeCd laser operating at 325 nm wavelength was approximately 75% of that measured for a *c*-plane epitaxially lateral overgrown (ELO) GaN sample of comparable thickness. This is remarkable considering that the GaN layer is on a Si substrate and *c*-plane GaN ELO samples have undergone years of improvement. For further evaluation of the optical quality, 6-nm-thick InGaN double heterostructure (DH) LED active layers were deposited by MOCVD both on the *m*-GaN sample on Si and on a high quality *c*-GaN template grown on *c*-sapphire. A frequency-doubled femtosecond Ti:sapphire laser emitting at 370 nm was used for excitation dependent PL measurements from which the internal quantum efficiencies (IQEs) were deduced using the method described in Ref. 19, but with Auger effects included. The IQE measurement and results of

the InGaN layers on Si have been described in another report.<sup>20</sup> As shown in Fig. 4, the PL emission peak blueshifts with increasing excitation density for the sample on *c*-plane GaN [Fig. 4(a)], while no shift is observed for the sample on *m*-GaN [Fig. 4(b)], indicating the absence of polarization fields. When compared to its *c*-plane counterpart, the IQE of the InGaN DH layer on *m*-GaN on Si is twice as high [see inset of Fig. 4(b)] and reaches 65% at a steady state carrier density of  $1.2 \times 10^{18} \text{ cm}^{-3}$ , corresponding to the maximum excitation density employed and the radiative recombination coefficient assumed ( $B = 10^{-11} \text{ cm}^3 \text{ s}^{-1}$ ).

In conclusion, the epitaxial growth of *m*-plane GaN films has been achieved on the vertical Si( $\bar{1}\bar{1}\bar{1}$ ) sidewalls of patterned Si(112) substrates, as confirmed by on-axis XRD measurements. Remarkably, IQEs of InGaN DH layers grown on these *m*-GaN films on Si are twice that of *c*-plane varieties, reaching 65% at a moderate carrier concentration of  $1.2 \times 10^{18} \text{ cm}^{-3}$ . The selective area growth of nonpolar, *m*-plane GaN layers on patterned and relatively low cost Si substrates is very promising for high efficiency light emitters.

The work at VCU is funded by the National Science Foundation (DMR, under Dr. Verne Hess). Discussions with Dr. C. Tran of SemiLEDs on the general topic of nonpolar surfaces in conjunction with LEDs are appreciated.

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