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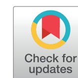
Optical anisotropy modulation in nonpolar *a*-plane AlGaN by manipulating the anisotropic in-plane strains through SiN_x interlayers engineering



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ABSTRACT

The achievement of high polarization nonpolar III-nitrides has been hindered by high density of defects, large surface fluctuations and in-plane strains. Here, we show that these critical challenges can be improved by manipulating the growth time of SiN_x interlayers. The discolation density of nonpolar *a*-plane $Al_{0.1}Ga_{0.9}N$ has been effectively reduced by introducing the *in-situ* deposited SiN_x interlayer, as the nonpolar *a*-plane $Al_{0.1}Ga_{0.9}N$ can only nucleate at open pores of SiN_x nanomask. During the growth process, the lateral overgrowth will occur from the openings with the dislocation bending and annihilation. Furthermore, the strip distributed nanomask can change the anisotropy of in-plane strains and optical properties of nonpolar *a*-plane $Al_{0.1}Ga_{0.9}N$. When the growth time of SiN_x increased from 0 min to 2 min, the in-plane strain along *m*-axis and *c*-axis direction changed from −0.2250% and −0.5145% to −0.0638% and −0.2728%, respectively. More importantly, the optical degree of polarization of nonpolar *a*-plane $Al_{0.1}Ga_{0.9}N$ has been changed from 0.4385 to 0.8129 along with the increase SiN_x deposition time by polarized photoluminescence. By simplely changing the SiN_x growth time, high polarization nonpolar *a*-plane devices may be fabricated on this high quality strain-modulatable nonpolar AlGaN template, which is desirable for high efficiency liquid crystal displays or other polarization sensitive photodetectors.

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1. Introduction

III-nitride semiconductors have attracted much attention for their interesting applications in optoelectronics and microelectronics such as light emitting diodes, laser diodes and high power transistors [1–5]. So far, nearly all wurtzite III-nitride-based devices have been grown predominantly in the (0001) or *c*-plane orientation. However, conventional *c*-plane nitrides-based optoelectronic devices suffer from high piezoelectric and spontaneous polarizations because of the non-centro symmetry of *c*-axis Wurtzite crystal structure. This built-in electric field from polarization effects distorts the electronic band structure of the active region and causes separation between the electron and hole wave functions, resulting in poor radiative recombination efficiency. Growing along one of the nonpolar directions (e.g. *a*- and *m*-directions) is expected to eliminate these effects since its polarization fields are normal to the growth direction [6].

In addition, nonpolar III-nitrides also show interesting optoelectronic properties because of the anisotropy in-plane strain caused by different lattice mismatch and expansion coefficient between the substrate and epilayer. This biaxial strain breaks the crystal symmetry and modifies the valence band structures from the unstrained $|X \pm iY\rangle$ heavy hole and light hole states into $|X\rangle$-like and $|Y\rangle$-like states, leading to anisotropic optical properties [7,8]. However, the application for nonpolar III-nitrides is limited by the high densities of dislocation, rough surface morphology and mediocre optical degree of polarization (DOP). In this report, the relationship of the SiN_x growth time, anisotropic in-plane strains and polarized PL properties in nonpolar *a*-$Al_{0.1}Ga_{0.9}N$ is discussed in detail. These results reveal that the growth time of SiN_x interlayer can effectively regulate the in-plane strain states of nonpolar

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a-plane $Al_{0.1}Ga_{0.9}N$, resulting in the difference in optical degree of polarization.

2. Experimental methods

All the *a*-$Al_{0.1}Ga_{0.9}N$ samples were grown directly on *r*-sapphire substrate in a vertical cold wall metalorganic chemical vapor deposition (MOCVD) reactor. During the growth, trimethylaluminium (TMA), trimethylgallium (TMG), silane (SiH_4) and ammonia (NH_3) were used as Al, Ga, Si and N sources, respectively. Hydrogen was used as the carrier gas. The schematic diagram of the sample structure is shown in Fig. 1(a). Prior to growth, the sapphire substrates were annealed at 910 °C for 5 min in the hydrogen atmosphere. After that, a thin HT-AlN buffer layer with ~100 nm thickness was deposited on the *r*-plane nitridated sapphire substrate at 1010 °C, the details of nitridation process were described in our previous study [9]. Subsequently, a 500-nm-thick *a*-plane $Al_{0.1}Ga_{0.9}N$ was grown at 890 °C, the flow rates of the TMGa, TMAl and ammonia were 88 μmol/min, 2 μmol/min and 2000 sccm. Then, a SiN_x intermediate layer was *in-situ* deposited on the $Al_{0.1}Ga_{0.9}N$, using silane (SiH_4) as precursor with a deposition time of 0 min (sample A), 2 min (sample B), 6 min (sample C). Finally, a ~1 *μm* thick undoped *a*-plane $Al_{0.1}Ga_{0.9}N$ layer was grown on the SiN_x with the same growth conditions as the AlGaN grown initially.

HR-XRD 2θ-ω scan and ω scan were performed by PANalytical X'pert PRO MRD Holland to investigate the crystal quality of the samples, respectively. The surface morphologies of all samples were examined by atomic force microscopy (AFM) in contact mode (Veeco NanoScope MultiMode). Raman spectra (Raman) were recorded with a JY-HR800UV Raman spectrometer using a 514.5 nm laser to study the strain states of all samples. PL measurements were performed by using the 4th harmonic of a Q-switched Cr:YAG laser with the wavelength of 266 nm as an excitation laser, and Glan-Taylor prism was used as polarizer to distinguish the light polarized in a specific direction. The polarization angle θ is defined for angle between the polarization direction of the Glan-Taylor Prism as shown in Fig. 1(b). To characterize optical polarization property, DOP is defined as [10].

$$DOP = \frac{I_{E\perp c} - I_{E\parallel c}}{I_{E\perp c} + I_{E\parallel c}} \quad (1)$$

where $I_{E\perp c}$ and $I_{E\parallel c}$ represent the PL intensities polarized perpendicular and parallel to the *c*-axis, respectively. Clearly, the emitting light with the electric vector perpendicular to *c*-axis (E⊥*c*) is TE modes, and the light with electric vector parallel to *c*-axis (E∥*c*) is TM modes, which is same as what is defined in *c*-plane AlGaN and *a*-GaN film [11].

3. Results and discussion

3.1. Surface morphologies and crystal quality of a-plane $Al_{0.1}Ga_{0.9}N$

The surface morphologies of three samples were studied by AFM. As shown in figure 2(a), sample A without SiN_x deposited exhibits a typical stripe feature along the [0001] direction with a RMS roughness of 9.63 nm. This wavy feature is generally considered to be the result of the anisotropic in-plane diffusion length along the *c*-axis and *m*-axis [12]. On the contrary, it can be seen from Fig. 2 (b) that the morphology of the sample has been greatly improved after the insertion of SiN_x layer, and the RMS roughness decreased from 9.63 to 2.67 nm. The difference between both surface morphologies may indicates that SiN_x interlayer can effectively inhibit the growth in the [0001] direction, thereby reducing the in-plane anisotropy. However, the overgrowth of SiN_x interlayer for sample C leads to a poor surface morphology. As can be seen from Fig. 2 (c), when the growth time change from 2 min to 6 min, triangular pits perpendicular to the *c* direction are clearly visible on the surface of sample. Meanwhile, the RMS roughness also changes to 7.52. The appearance of triangular pits may be caused by the long growth time leading to a large number of SiN_x islands, which suppress the subsequent growth of AlGaN and produce a large number of dislocations.

The crystalline properties of the three samples were investigated by high-resolution x-ray diffraction (HR-XRD). Fig. 3(a) shows the typical XRD 2θ-ω scan of the three samples. No clear distinction is observed in this image for sample with different SiN_x growth time. The peak at 57.88° is assigned to *a*-plane $Al_{0.1}Ga_{0.9}N$ $(11\bar{2}0)$ reflection. Fig. 3(b) shows the full widths at half maximum (FWHMs) of the on-axis $(11\bar{2}0)$ XRCs as a function of the azimuth angles. The azimuth angle was defined as zero when the incident beam is parallel to the [0001] direction (the striation direction). The rocking curves exhibit M-shaped dependence with respect to azimuthal angle, which reveals the crystallographic anisotropic in-plane mosaicity of the *a*-plane AlGaN film. Such type of in-plane FWHM dependence is in agreement with the results reported by other groups [13,14]. As indicated in Fig. 3(b), the FWHM values and

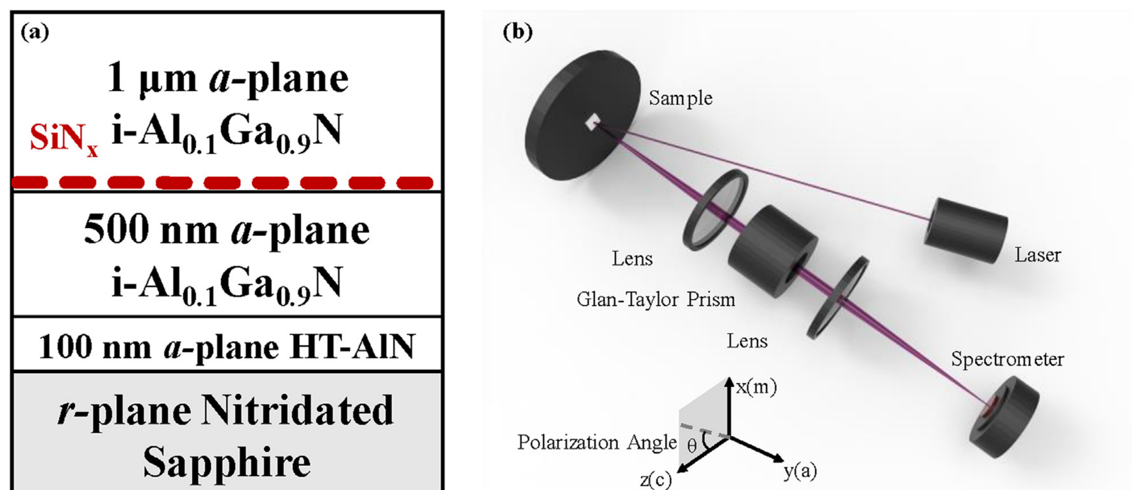

Fig. 1. (a) Schematic structure of samples with SiN_x interlayer. (b) Schematic diagrams of polarization-dependent PL measurement setup in this study.

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