Particular electrical quality of *a*-plane GaN films grown on *r*-plane sapphire by metal-organic chemical vapor deposition*

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Abstract: Nonpolar $(11\bar{2}0)$ *a*-plane GaN films have been grown by low-pressure metal-organic vapor deposition on *r*-plane $(1\bar{1}02)$ sapphire substrate. The structural and electrical properties of the *a*-plane GaN films are investigated by high-resolution X-ray diffraction (HRXRD), atomic force microscopy (AFM) and van der Pauw Hall measurement. It is found that the Hall voltage shows more anisotropy than that of the *c*-plane samples; furthermore, the mobility changes with the degree of the van der Pauw square diagonal to the *c* direction, which shows significant electrical anisotropy. Further research indicates that electron mobility is strongly influenced by edge dislocations.

Key words: GaN; anisotropic; HRXRD; nonpolar
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1. Introduction

Recently, considerable attention has been paid to the growth of nonpolar III-nitride films for polarization-free heterostructure optoelectronic and electronic devices[1, 2]. Spontaneous and strain-induced piezoelectronic polarization effects produce strong electric fields at the hetero-interfaces, and the electric fields can cause spatial separation of electrons and holes in quantum wells of light-emitting diodes, reduction of the recombination efficiency in light-emitting devices and a redshift of the emission wavelength. On the other hand, it is difficult to enable normally-off operations on the *c*-plane face[3−5]. To eliminate the internal polarization fields, group III nitride layers have been recently grown on nonpolar planes, such as the *a*-plane.

The anisotropic structure is one of the most remarkable characteristics of $(11\bar{2}0)$ *a*-plane GaN. The structural anisotropy will lead to anisotropic electrical and optical characteristics of *a*-plane GaN. Furthermore, the scattering centers which primarily influence the electron mobility need to be defined on *a*-plane GaN. Actually, electron mobility is very important among the parameters characterizing the material quality of GaN films[6−8]. However, detailed reports and systematic investigations of the electron mobility of *a*-plane GaN are lacking.

2. Experimental

In this paper, *a*-GaN films were grown on *r*-plane $(11\bar{2}0)$ sapphire substrates using a cold-wall shower head MOCVD system. After chemical cleaning, *r*-plane sapphire substrates were loaded into the chamber. A technique which consists of two steps, low temperature AlN and AlN/AlGaN SLs nucleation layer, was utilized before increasing the temperature for film growth. Hydrogen was used as the carrier gas and triethylgallium, trimethylaluminium and ammonia (NH_3) were used as compound sources. HRXRD was performed to obtain the FWHMs of scan rocking curves. The van der Pauw method was employed in the Hall measurements to obtain the electron mobility of the GaN samples. After buffer layer growth, a 1.5-μm-thick *a*-plane GaN epitaxial layer was grown for the samples. The FWHM for the $(11\bar{2}0)$ plane XRD rocking curve is only 662 arcsec, much lower than the previous reports that were grown without using the lateral overgrowth technique[9, 10].

3. Results and discussion

AFM images of the *a*-plane GaN on *r*-sapphire are shown in Fig. 1. The anisotropic growth rate which induces crystallographic anisotropy of the *a*-plane has previously been studied, but the anisotropic electrical property has not been revealed. The Hall measurement is one of the most useful methods for characterizing the material quality of GaN. The combined method of high-resolution X-ray diffraction and Hall measurement are employed for the study of anisotropic electrical characteristics. As shown in Fig. 2, the FWHM measured toward the *c* axis is lower than that toward the *m* axis. The XRD scan results show much greater differences along different directions, which do not appear in the *c*-plane GaN. From Fig. 3, we conclude that the FWHM reaches minimum and maximum values along the [0001] and the $[1\bar{1}00]$ directions.

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Table 1. Hall voltage results comparison for *c*-plane and *a*-plane GaN.

Sample	V_1 (V)	V_2 (V)	V_3 (V)	V_4 (V)	V_5 (V)	V_6 (V)	V_7 (V)	V_8 (V)
c-plane	0.017	0.017	0.016	0.016	0.018	0.018	0.016	0.016
a−plane	0.013	0.013	0.032	0.032	0.012	0.012	0.032	0.032

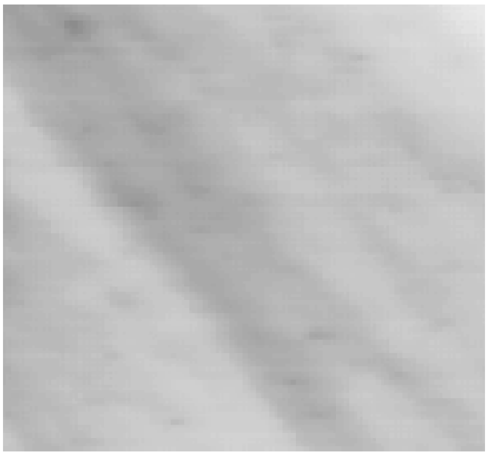

Fig. 1. AFM image of GaN on *r*-plane sapphire.

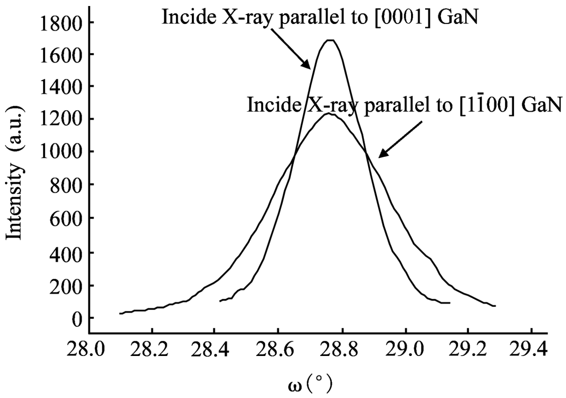

Fig. 2. XRD rocking curves of *a*-plane GaN along the *c* and *m* directions.

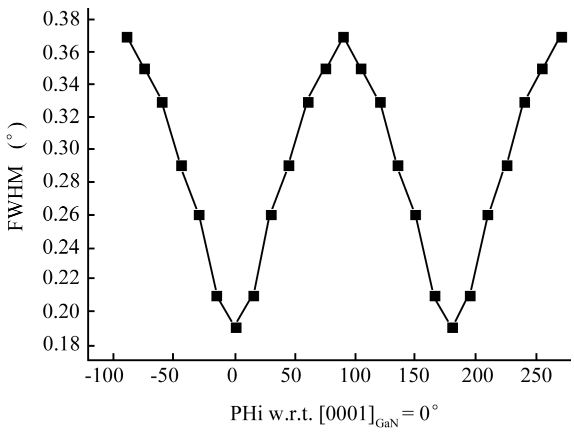

Fig. 3. XRD FWHMs anisotropy with respect to the in-plane beam orientation.

At the same time, the Hall measurements also show great differences. As shown in Table 1, the Hall voltages of the *c*-plane are 0.17, 0.16, 0.18 and 0.16 V, while those of the *a*-plane are 0.13, 0.32, 0.12 and 0.32 V, respectively. It is interesting to note that the mobility changes with the degree of the van der Pauw square diagonal to the *c* direction. The

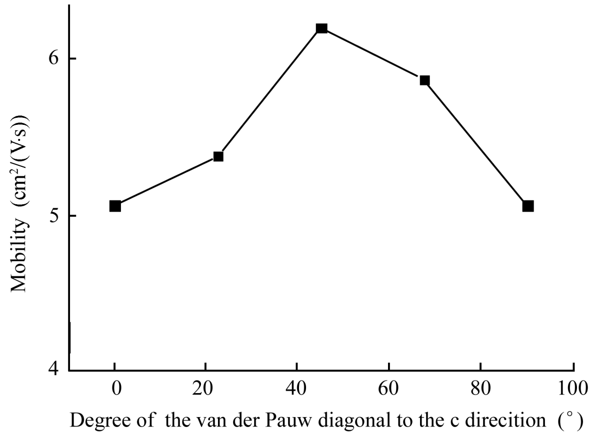

Fig. 4. Dependence of electron mobility on the degree of the van der Pauw square diagonal to the *c* direction.

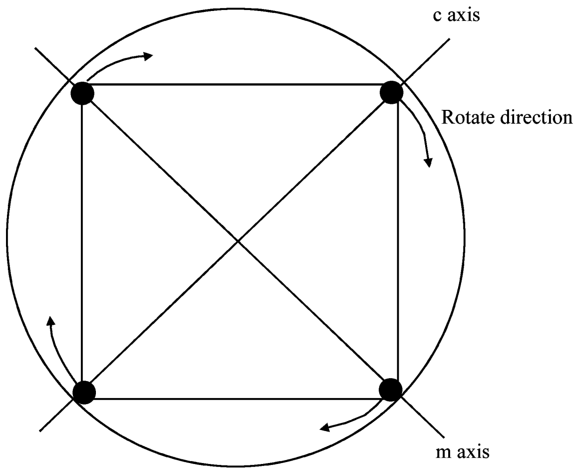

Fig. 5. Van der Pauw Hall anisotropic measurement for *a*-plane GaN.

relation of the mobility to the degree is shown in Fig. 4. Statistical regulation indicates that the electrical anisotropy of *a*-plane GaN appears, but the *c*-plane GaN does not have this property. It is worthwhile noting that the *c*-plane GaN is also free from anisotropic structural characteristics. The unequal growth rates will lead to anisotropic structural characteristics, and structural anisotropy will result in electrical anisotropy. It is well known that the FWHMs of the ω-scan rocking curve for the symmetric and skew symmetric planes indirectly represent the density of screw and edge dislocations and are correlated with the mean tilt and twist angles of the mosaic crystals, respectively[10−13]. So the dislocations along the [0001] direction are much lower than those along the [1$\bar{1}$00] direction. The Hall measurement result is affected by the duplex directions, in which the carriers suffer the dislocation scattering centers in different directions. Considering the influence of these defects of dislocation scatter, the electrical anisotropy of *a*-plane GaN can be explained. The van der Pauw Hall anisotropic measurement for *a*-plane GaN is shown in Fig. 5.

Table 2. Characterization results of the *a*-plane GaN samples.

Sample	FWHM of (110) DCXRD (arcsec)	FWHM of (101) DCXRD (arcsec)	Mobility ($cm^2/(V\cdot s)$)	Carrier concentration (cm^{-3})
A	742	2363	5.83	5.39×10^{17}
B	726	1909	7.11	9.93×10^{17}
C	660	1318	12.21	2.17×10^{18}

Likewise, we investigated the effect of the structural quality on the electron mobility of *a*-plane GaN by using HRXRD and room temperature Hall measurements on different series of samples. It is well known that the density of screw dislocations and the density of edge dislocations can be indirectly represented by the FWHM of XRD for the symmetric geometry and skew symmetric plane. To determine the lattice structure parameters of the *a*-plane GaN layers, various GaN reflections were measured under symmetric (11$\bar{2}$0) and skew symmetric (10$\bar{1}$1) diffraction geometries. These samples were prepared under different growth conditions (V/III ratio, pressure, growth temperature, and so on) and different FWHMs of rocking curves were achieved. It is found that the FWHM values for the (11$\bar{2}$0) plane are not remarkable with increasing electron mobility. However, as shown in Table 2, there is a very well-regulated relationship between the electron mobility and the FWHM of the (10$\bar{1}$1) plane. This implies that the electrical properties of the nonpolar *a*-plane GaN films are significantly influenced by the effect of edge dislocations. It confirms that the electron mobility decreases with the increase of edge dislocation density in the *a*-plane samples. There are many dangling bonds along the edge dislocation lines. They may provide acceptor traps and form negatively charged Coulombic scattering centers in GaN. These edge dislocation lines will cause a reduction of electron mobility. At the same time, the acceptors introduced by edge dislocations in n-type GaN distributed along the edge dislocation lines can capture electrons from the conduction band in n-type semiconductors. So the net carrier concentration increases with the decrease of FWHM at the (10$\bar{1}$1) plane[14].

The dependence of net carrier concentration for the *a*-plane GaN samples is shown in Table 2. Sample C has the lowest FWHM of the (10$\bar{1}$1) plane, and at the same time it also has the highest mobility and carrier concentration, which is consistent with the discussion above.

4. Conclusions

In conclusion, high-quality triangle pit-free *a*-plane GaN (11$\bar{2}$0) films have been grown on *r*-plane sapphire (1$\bar{1}$02) substrate. We have demonstrated the anisotropic structural and electrical characteristics. The unequal growth rates will cause anisotropic structural characteristics, and the structural characteristic anisotropy will result in electrical anisotropy. For the *a*-plane GaN samples it is found that the electron mobility decreases when the edge dislocation density increases. Further research indicates that the acceptors introduced by edge dislocations in *a*-plane GaN lead to a reduction of the electron mobility.

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