

Erbium-doped *a*-plane GaN epilayers synthesized by metal-organic chemical vapor deposition

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Abstract: A-plane GaN epilayers doped with erbium (GaN:Er) have been grown on r-plane sapphire substrates by metal organic chemical vapor deposition. The 1.54 μm emission properties were probed by photoluminescence (PL) emission spectroscopy and compared with those of c-plane GaN:Er. It was found that the emission intensity from a-plane GaN:Er is 4 times higher than that of c-plane GaN:Er. The intensity of the 1.54 μm emission was found to increase with increasing Er molar flux. A-plane Er-doped GaN epilayers exhibit a small thermal quenching effect, with only a 12% decrease in the integrated intensity of the 1.54 μm PL emission, occurred between 10 and 300 K.

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OCIS codes: (160.5690) Rare-earth-doped materials; (250.5230) Photoluminescence.

References and links

1. R. G. Wilson, R. N. Schwartz, C. R. Abernathy, S. J. Pearton, N. Newman, M. Rubin, T. Fu, and J. M. Zavada, "1.54- μm photoluminescence from Er-implanted GaN and AlN," *Appl. Phys. Lett.* **65**(8), 992–994 (1994).
2. M. Thaik, U. Hommerich, R. N. Schwartz, R. G. Wilson, and J. M. Zavada, "Photoluminescence spectroscopy of erbium implanted gallium nitride," *Appl. Phys. Lett.* **71**(18), 2641–2643 (1997).
3. M. R. Brown, A. F. J. Cox, W. A. Shand, and J. M. Williams, "The spectroscopy of rare earth doped chalcogenides," *Advances in Quantum Electronics* **2**, 69–155 (1974).
4. J. T. Torvik, R. J. Feuerstein, J. I. Pankove, C. H. Qiu, and F. Namavar, "Electroluminescence from erbium and oxygen coimplanted GaN," *Appl. Phys. Lett.* **69**(14), 2098–2100 (1996).
5. A. J. Steckl, J. Heikenfeld, M. Garter, R. Birkhahn, and D. S. Lee, "Rare earth doped gallium nitride — light emission from ultraviolet to infrared," *Compound Semiconductor* **6**(1), 48–52 (2000).
6. J. M. Zavada, S. X. Jin, N. Nepal, H. X. Jiang, J. Y. Lin, P. Chow, and B. Hertog, "Electroluminescent properties of erbium-doped III–N light-emitting diodes," *Appl. Phys. Lett.* **84**(7), 1061–1063 (2004).
7. H. Ennen, J. Schneider, G. Pomrenke, and A. Axmann, "1.54- μm luminescence of erbium-implanted III–V semiconductors and silicon," *Appl. Phys. Lett.* **43**(10), 943–945 (1983).
8. P. N. Favennec, H. L'Halidon, M. Salvi, D. Moutonnet, and Y. Le Guillou, "Luminescence of erbium implanted in various semiconductors: IV, III–V and II–VI materials," *Electron. Lett.* **25**(11), 718–719 (1989).
9. A. J. Neuhalfen and B. W. Wessels, "Thermal quenching of Er³⁺-related luminescence in In_{1-x}Ga_xP," *Appl. Phys. Lett.* **60**(21), 2657–2659 (1992).
10. J. M. Zavada and D. Zhang, "Luminescence properties of erbium in III–V compound semiconductors," *Solid-State Electron.* **38**(7), 1285–1293 (1995).
11. U. Hömmerich, J. T. Seo, M. Thaik, C. R. Abernathy, J. D. MacKenzie, and J. M. Zavada, "Near infrared (1.54 μm) luminescence properties of erbium doped gallium nitride," *J. Alloy. Comp.* **303–304**, 331–335 (2000).
12. J. D. MacKenzie, C. R. Abernathy, S. J. Pearton, U. Hommerich, J. T. Seo, R. G. Wilson, and J. M. Zavada, "Er doping of GaN during growth by metalorganic molecular beam epitaxy," *Appl. Phys. Lett.* **72**(21), 2710–2712 (1998).
13. S. Kim, S. J. Rhee, D. A. Turnbull, X. Li, J. J. Coleman, S. G. Bishop, and P. B. Klein, "Trap-mediated excitation of Er³⁺ photoluminescence in Er-implanted GaN," *Appl. Phys. Lett.* **71**(18), 2662–2664 (1997).
14. J. M. Zavada, M. Thaik, U. Hommerich, J. D. MacKenzie, C. R. Abernathy, S. J. Pearton, and R. G. Wilson, "Luminescence characteristics of Er-doped GaN semiconductor thin films," *J. Alloy. Comp.* **300–301**, 207–213 (2000).

15. M. Garter, J. Scofield, R. Birkhahn, and A. J. Steckl, "Visible and infrared rare-earth-activated electroluminescence from indium tin oxide Schottky diodes to GaN:Er on Si," *Appl. Phys. Lett.* **74**(2), 182–184 (1999).
16. S. Kim, S. J. Rhee, X. Li, J. J. Coleman, S. G. Bishop, and P. B. Klein, "Selective enhancement of 1540 nm Er³⁺ emission centers in Er-implanted GaN by Mg codoping," *Appl. Phys. Lett.* **76**(17), 2403–2405 (2000).
17. D. S. Lee, J. Heikenfeld, A. J. Steckl, U. Hommerich, J. T. Seo, A. Braud, and J. M. Zavada, "Optimum Er concentration for *in situ* doped GaN visible and infrared luminescence," *Appl. Phys. Lett.* **79**(6), 719–721 (2001).
18. C. Ugolini, N. Nepal, J. Y. Lin, H. X. Jiang, and J. M. Zavada, "Erbium-doped GaN epilayers synthesized by metal-organic chemical vapor deposition," *Appl. Phys. Lett.* **89**(15), 151903 (2006).
19. C. Ugolini, N. Nepal, J. Y. Lin, H. X. Jiang, and J. M. Zavada, "Excitation dynamics of the 1.54 μm emission in Er doped GaN synthesized by metal organic chemical vapor deposition," *Appl. Phys. Lett.* **90**(5), 051110 (2007).
20. S. C. Cruz, S. Keller, T. E. Mates, U. K. Mishra, and S. P. DenBaars, "Crystallographic orientation dependence of dopant and impurity incorporation in GaN films grown by metalorganic chemical vapor deposition," *J. Cryst. Growth* **311**(15), 3817–3823 (2009).
21. I. W. Feng, J. Li, A. Sedhain, J. Y. Lin, H. X. Jiang, and J. Zavada, "Enhancing erbium emission by strain engineering in GaN heteroepitaxial layers," *Appl. Phys. Lett.* **96**(3), 031908 (2010).

1. Introduction

Doped in a solid host, Er³⁺ ion has allowable intra-4f shell transition from its first excited state ⁴I_{13/2} to the ground state ⁴I_{15/2} and the transition corresponds to a wavelength of minimum optical loss in silica based optical fibers (1.54 μm) [1–7]. Thus, Er-doped materials are ideal candidates to make amplifiers for optical communications. There has been considerable effort devoted to the development of Er doped semiconductors aimed at achieving emitters, optical amplifiers, and photonic integrated circuits with the combined functionalities of Er optical fibers and narrow bandgap semiconductors such as InGaAsP. While the Er³⁺ related emission energy is not affected by the host, its emission intensity decreases as the temperature is raised. This thermal quenching effect limits the light emission efficiency at room temperature and, as a consequence, prevents the employment of the existing Er-based semiconductor devices in broad applications. The fact that the thermal stability of Er emissions increases with an increase in the energy gap of the semiconductor host material [8,9] has motivated research efforts to use wide bandgap semiconductors as host materials for Er. GaN has also been studied [10–13] for its optoelectronic application of infrared emission at 1.54 μm using Er doping. GaN has advantages over other semiconductors such as a direct bandgap transition which is very important in optical applications, a large energy bandgap (3.4 eV) which results in very low thermal quenching [14], and thermal and chemical robustness. GaN has proven to be an excellent host for Er ions [15–17]. C-plane Er doped GaN grown by metal organic vapor deposition (MOCVD) showed a low degree of thermal quenching (20%) between 10 and 300 K [18,19]. However, further improvements in quantum efficiency at 1.54 μm through the reduction of thermal quenching and increasing of Er incorporation in the optically active sites are still needed.

So far, all reported works on GaN:Er are for epilayers grown on c-plane and there have been no reports on the impact of the crystal orientation on Er incorporation and Er related emission in non-c-plane GaN films. Strong effects of the crystal orientation on dopant and impurity incorporation have been observed in GaN, GaAs, and InP [20]. In this work, we investigate MOCVD growth of non-polar *a*-plane GaN doped with Er and the emission properties of the Er³⁺ related emission at 1.54 μm . The results showed that the 1.54 μm emission intensity from *a*-plane GaN:Er is about 4 times higher than that from c-plane GaN:Er and the thermal quenching in the temperature range of 10 - 300 K for *a*-plane GaN:Er is about 12% compared to about 20% for c-plane GaN:Er.

2. Experimental details

GaN:Er epilayers were deposited on r- and c-plane sapphire substrates using MOCVD. For comparison, both r-plane and c-plane sapphire substrates were put side by side in the reactor. Hence, both *a*-plane and c-plane GaN:Er epilayers were grown under identical conditions.

Both samples have approximately the same thickness, indicating equal growth rates. The gallium source was trimethylgallium (TMG) and the nitrogen source was ammonia (NH₃). Tris(isopropyl)cyclopentadienyl erbium (TRIPER) was used for the *in situ* Er doping. Hydrogen was the carrier gas and was kept constant at 2 standard liter per minute (SLM). The growth started with a thin (50 nm) GaN buffer layer grown at 550 °C and 300 Torr, followed by the growth of ~1.4 μm GaN template grown at 1060 °C and 300 Torr and a 0.5 μm Er-doped GaN epilayers grown at 1040 °C and 30 Torr. The inset of Fig. 1 shows the layer structure of Er doped *a*-plane GaN epilayer grown on *r*-plane sapphire substrate. X-ray diffraction (XRD) was used to determine crystalline quality. Photoluminescence (PL) spectroscopy was employed to study the optical properties of the Er-doped GaN epilayers. The PL system consists of a frequency doubled, tripled, and quadruple 100 femtosecond Ti: Sapphire laser with an average power of 150 mW at 263 and 395 nm, 1–3 mW at 196 nm, and a repetition rate of 76 MHz. Infrared detection was accomplished by an InGaAs detector.

3. Results and discussion

The growth surface of GaN:Er epilayers grown on *r*-plane sapphire was determined to be *a*-plane using x-ray diffraction (XRD) θ - 2θ scan, which detected sapphire (10-12), (20-24), and *a*-plane GaN (11-20) reflections, as illustrated in Fig. 1. Only the *a*-plane reflection peak at $2\theta = 57.7^\circ$ was observed. Since the GaN (0002) reflection at $2\theta = 34.56^\circ$ was not detected, demonstrating that the films were uniformly *a*-plane oriented. The measured full-width at half-maximum (FWHM) of the (11-20) peak in the ω -scan (x-ray rocking curve) is about 600 arcsec. For comparison, the XRD rocking curve FWHM of the (0002) peak of *c*-plane GaN:Er is about 450 arcsec.

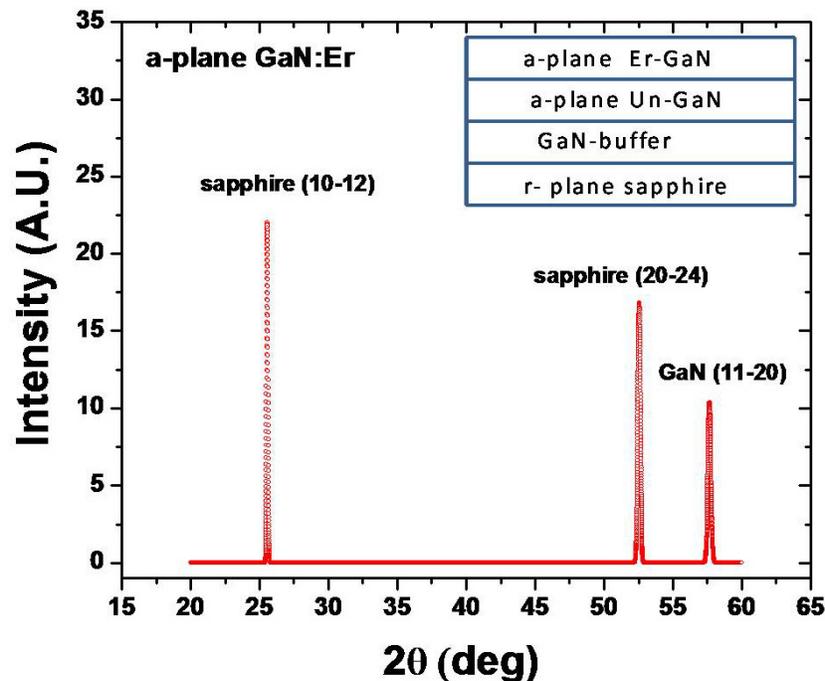


Fig. 1. X-ray diffraction θ - 2θ scan of *a*-plane Er doped GaN epilayer grown on *r*-plane sapphire substrate. Inset: Schematic layer structure of *a*-plane Er doped GaN epilayer (GaN:Er).

Figure 2 compares the room temperature (300 K) PL spectra of *a*-plane and *c*-plane GaN:Er epilayers covering a spectral range from 1.40 to 1.65 μm, using excitation wavelength

$\lambda_{\text{exc}} = 375$ nm. It is observed that the $1.54 \mu\text{m}$ emission intensity from *a*-plane GaN:Er is about 4 times higher than that of *c*-plane GaN:Er. SIMS results provide a concentration of about $2 \times 10^{20} \text{ cm}^{-3}$ in both samples, which suggest that the Er doping concentration is predominantly controlled by the Er flow rate and GaN growth rate. However, a small impact of the surface orientation on the Mg, Si, and Fe dopant incorporation was observed by Cruz et al [20]. The Er dopant's environment and strain are not identical in *a*-plane and *c*-plane GaN:Er epilayers, which could contribute to the observed difference in the $1.54 \mu\text{m}$ emission intensities [21].

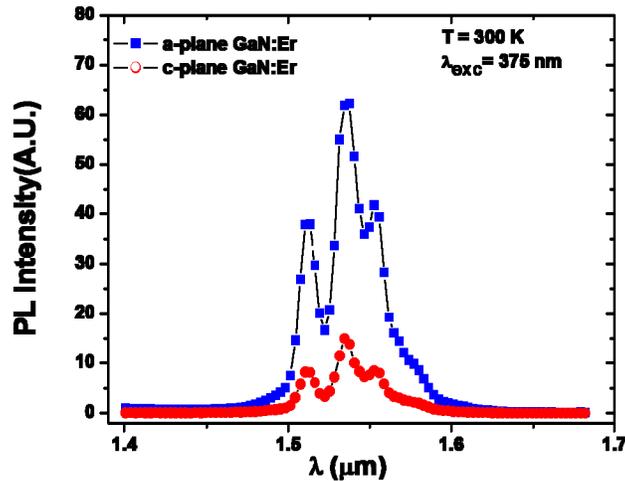


Fig. 2. Comparison of room temperature (300 K) Er-related PL spectra near $1.54 \mu\text{m}$ between *a*-plane and *c*-plane Er doped GaN epilayers. The excitation wavelength used was 375 nm.

It is well known that the intensity of the Er-related PL spectra is a strong function of the Er concentration in GaN epilayer. Hence the Er concentration was varied to determine the optimal Er concentration. The Er concentration was controlled by changing the Er molar flux from 20.2 - 31.5 $\mu\text{mol}/\text{min}$. Figure 3 shows the dependence of the $1.54 \mu\text{m}$ PL spectra measured at 300 K for $\lambda_{\text{exc}} = 375$ nm on Er molar flux. It can be seen that the $1.54 \mu\text{m}$ emission line shape is independent of the Er flux. The inset of Fig. 3 shows the PL intensity at $1.54 \mu\text{m}$ as a function of Er molar flux and demonstrates a direct correlation between PL intensity at $1.54 \mu\text{m}$ and Er molar flux in the studied range. The results indicate that the number of optical active centers increases almost linearly with an increase of the Er molar flux.

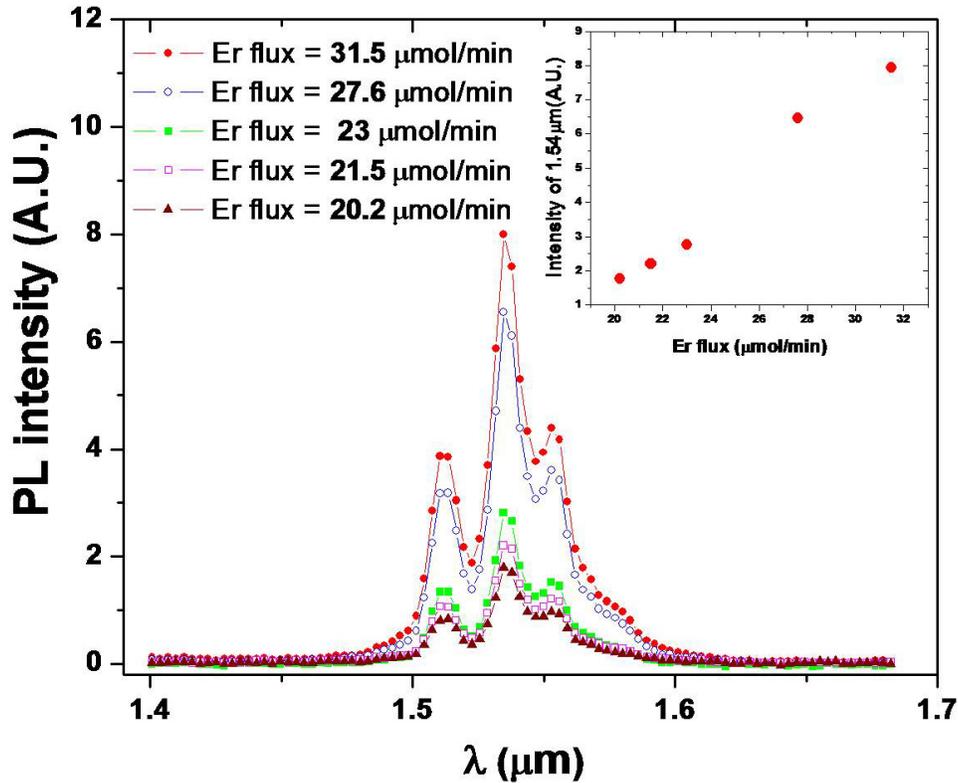


Fig. 3. Room temperature PL spectra of *a*-plane Er doped GaN epilayers showing the variation in the emission intensity of the Er related emission at 1.54 μm with the Er molar flux employed during the growth. The excitation wavelength used was 375 nm. Inset: The PL intensity at 1.54 μm as a function of Er molar flux.

Figure 4 shows the temperature evolution of the PL spectra near 1.54 μm of *a*-plane GaN:Er measured in the spectral range of 10 - 400 K. The 1.54 μm emission peak due to the radiative intra- $4f$ Er^{3+} transition from ${}^4I_{13/2}$ to ${}^4I_{15/2}$ is dominant at low temperatures with its emission intensity decreases with an increase in temperature. The transition located at 1.51 μm is not visible at 10 K and its emission intensity increases with temperature following a different trend compared to the 1.54 μm peak.

Knowledge on the thermal-quenching properties of the Er-related luminescence at 1.54 μm can be obtained from temperature evolution of the PL spectra. Figure 5 shows the integrated PL emission intensity (I_{int}) of the 1.54 μm emission line (integrated over a range of 1.45 - 1.62 μm) for *a*-plane GaN:Er as a function of temperature between 10 and 400 K. The temperature dependence of the integrated PL shows a 12% decrease in I_{int} between 10 and 300 K. This small degree of thermal quenching is even lower than the reported thermal quenching (20%) for Er doped *c*-plane GaN [18,19].

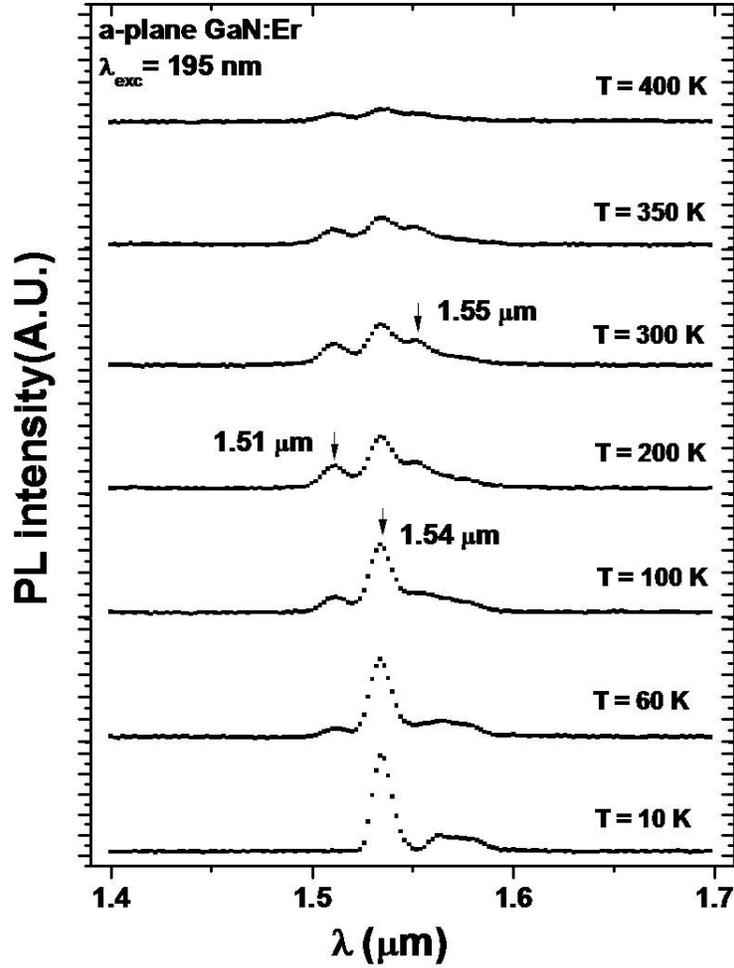


Fig. 4. PL spectra of the 1.54 μm emission of *a*-plane Er doped GaN measured at different temperatures from 10 to 400 K. The excitation wavelength used was 195 nm.

The inset of Fig. 5 shows the Arrhenius plot of the integrated PL intensity near 1.54 μm between 10 and 400 K for *a*-plane GaN:Er. The solid line is the least squares fit of the measured data to the equation

$$I_{\text{int}}(T) = I_0 \left[1 + C e^{\left(\frac{-E_0}{kT} \right)} \right]^{-1} \quad (1)$$

where $I_{\text{int}}(T)$ and I_0 are, respectively, the integrated PL intensities at finite temperature T and 0 K, while E_0 is the activation energy of the thermal quenching, and k is Boltzmann's constant. The PL intensity is thermally activated with an activation energy of about 258 ± 15 meV, which is larger than a value of 191 meV observed in *c*-plane GaN:Er [19] and explains the reduced thermal quenching of the 1.54 μm emission in *a*-plane GaN:Er. This again suggests that the Er dopant's environment is not identical in *a*-plane and *c*-plane GaN:Er epilayers.

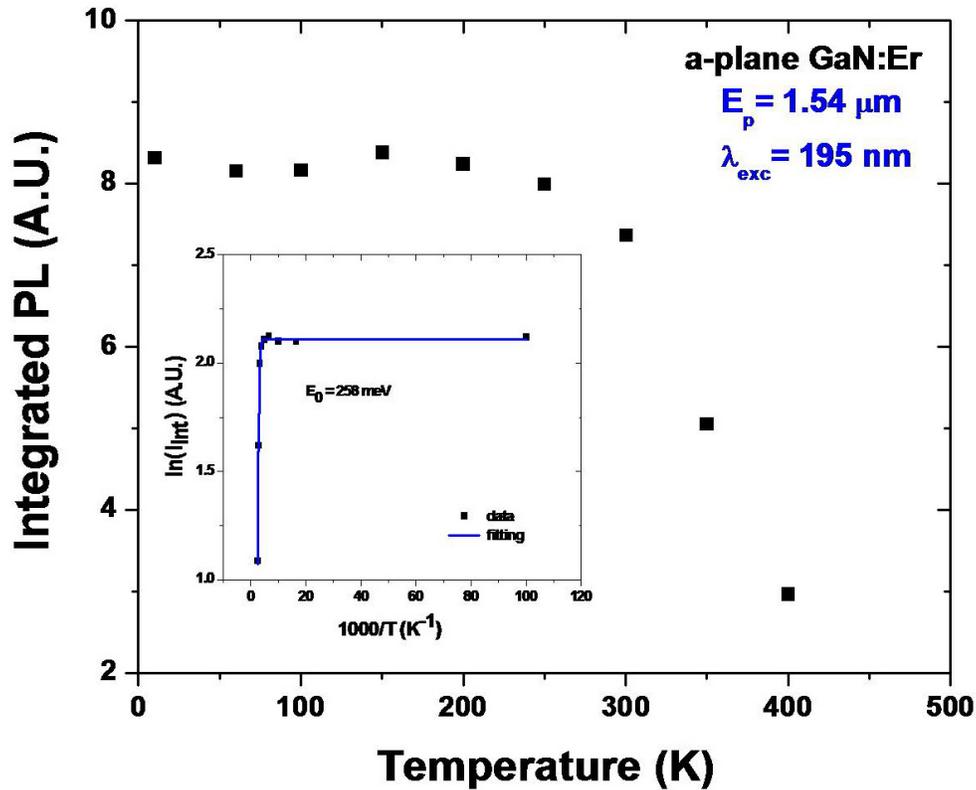


Fig. 5. Integrated PL emission intensity of the 1.54 μm emission of *a*-plane Er doped GaN vs sample temperature. Inset: Arrhenius plot of the integrated PL intensity of the 1.54 μm emission. The solid line in the plot is the least squares fit of the measured data to Eq. (1). The excitation wavelength used was 195 nm.

4. Conclusion

In summary, *a*-plane Er doped GaN epilayers were grown by MOCVD. It was found that the room temperature intensity of Er-related PL emission at 1.54 μm for *a*-plane Er-doped GaN is four times stronger than that of *c*-plane Er-doped GaN. The emission intensity at 1.54 μm was found to increase almost linearly with an increase of Er molar flux employed during the growth. Thermal quenching of the emission at 1.54 μm in *a*-plane GaN:Er from 10 to 300 K is 1.6 times smaller than that measured in *c*-plane GaN:Er. The results imply that the use of *a*-plane GaN as a host crystal for Er doping represents a promising route for realizing 1.54 μm optoelectronic and photonic devices with improved emission efficiency.

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