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Nitride-based laser diodes by plasma-assisted MBE—From violet to green emission

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ABSTRACT

We present recent progress in growth of nitride-based laser diodes (LDs) and efficient light-emitting diodes (LEDs) made by plasma-assisted MBE (PAMBE). This technology is ammonia free, and nitrogen for growth is activated by RF plasma source from nitrogen molecules. The recent demonstration of CW blue InGaN LDs has opened a new perspective for PAMBE in optoelectronics. The LDs were fabricated on low threading dislocation density (TDD) bulk GaN substrates at low growth temperatures 600–700 °C. In this work, we describe the nitride growth fundamentals, the influence of the TDD on the layer morphology, the peculiarities of InGaN growth as well as properties of LEDs and LDs made by PAMBE.

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1. Introduction

The potential associated with nitride-based light-emitting diodes (LEDs) and laser diodes (LDs) for solid-state lighting has generated a continuing interest in group III-N materials and their alloys. Until very recently, the key achievements and developments in the field of InGaN laser diodes have been made by the metal-organic vapour-phase epitaxy (MOVPE) technique [1,2]. In spite of many potential advantages for MBE growth, such as *in situ* monitoring techniques [3–5], the low quantum efficiency of MBE-grown optoelectronic structures compared with MOVPE led many researchers to conclude that MBE cannot compete with MOVPE. It was also commonly believed that the only solution for MBE would be to bring the growth conditions as close as possible to the MOVPE, i.e. to apply high growth temperatures (~1050 °C for GaN) and high nitrogen precursor overpressure. Indeed, ammonia MBE, which uses atomic beams for group III elements coupled with a large excess supply of NH₃ as the nitrogen precursor [3,4], has been successful in improving the optical quality of nitride films and recently produced the first room-temperature continuous wave (CW) operation of a 405 nm LDs [6]. However, the corrosive nature of ammonia compounded by its large flows creates additional hazards and technological challenges, and also leads to undesirable high hydrogen background during the epitaxial

process. In the more widely employed plasma-assisted MBE (PAMBE) [5], purified nitrogen gas is activated using an RF-plasma and supplied to the growth surface at typical flow rates of 1–2 sccm. Early experimental results [7] showed that unlike ammonia MBE, PAMBE requires group III-rich conditions to achieve good material quality. Much progress has been made in both theoretical [8,9] and experimental [10–13] understanding of the growth kinetics for such metal-rich conditions. In spite of the relatively low growth temperatures, 650–750 °C, state-of-the-art GaN/AlGaN heterostructures with record high mobilities of two-dimensional electron gas [14–16] have been grown with PAMBE, making it the technique of choice for the production of electronic devices. Although, sustained refinement of the PAMBE growth conditions on GaN/sapphire MOVPE templates allowed the demonstration of promising LED devices [17], it was the introduction of high-quality GaN substrates which led to dramatic improvements in the optical quality of PAMBE-grown structures, resulting in the room-temperature high-power pulsed and CW blue-violet lasers [18,19]. This paper reports on the progress made by PAMBE in the area of optoelectronic devices, which demonstrates the potential of this technology.

2. Experimental procedure

The growth of all nitride structures presented in this paper was performed in a customized VG90 Oxford MBE reactor equipped with a Veeco RF plasma source (operating at 240 W for 0.8 sccm

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N_2 flow). The pressure during growth was 1.5×10^{-5} Torr. The substrates used were either high-pressure-grown bulk GaN [20] or GaN/ Al_2O_3 templates made by MOVPE. The epi-ready bulk substrates were prepared either in a three-step process of mechanical polishing, dry etching, and deposition of a $2 \mu m$ GaN:Si buffer layer in the MOVPE reactor or by the two-step procedure: mechanical polishing and mechano-chemical polishing. The back surfaces of the substrates were coated with a $0.7 \mu m$ molybdenum layer to improve the thermal coupling for radiative heating. Special holders capable of accommodating small, irregularly shaped substrates, and designed to minimize edge effects ensured high temperature uniformity across the entire substrate area. The typical size of GaN bulk high-pressure substrates was $4 mm \times 5 mm$, while MOVPE GaN/ Al_2O_3 templates were $10 mm \times 10 mm$. All layers discussed in this paper were grown on the (0001) Ga-polarity surface.

3. Growth conditions for PAMBE

The analysis of experimental data for GaAs or Si indicates that the optimum growth temperature for 2D step-flow mode equals about half of the melting temperature, T_M [21,22]. This rule is also valid for GaN growth, where the GaN melting point lies somewhere between 2540 K (experimentally determined for 6 GPa [23]) and 2800 K (theoretically calculated [24]), and the optimum growth temperature used in MOVPE is in the range 1050–1100 °C (1320–1370 K). On a microscopic scale, at $T=0.5 \cdot T_M$ atomic kinks start to become active, i.e. atoms can detach from an atomic kink [21,22] at a rate comparable to typical attachment rates during deposition. This rule is strict for Kossel crystal and is also fulfilled for real crystal structures.

The difficulty for GaN growth in MBE is that for temperature $0.5 \cdot T_M=1320 K$, a high nitrogen overpressure (about 60 bars for N_2) is required to prevent GaN decomposition (see Fig. 1) [25,26]. For more chemically active N precursors, like ammonia, the situation is better, but still an overpressure in the range 0.1–2 bar must be used. Such overpressures are not compatible with MBE technology, which relies on high vacuum conditions for the delivery of atoms from the effusion cells to the growing layer. Gallium nitride is a strongly bonded compound (with bonding energy of 9.12 eV/atom pair [27]) compared with typical III–V semiconductors like GaAs (bonding energy of 6.5 eV/atom pair [27]). Consequently, the free energy of the crystal is very low in relation to the reference state of free N and Ga atoms. On the other hand, the N_2 molecule is also strongly bonded (4.9 eV/atom).

Fig. 1. p – T phase diagram of GaN(s)–Ga(l)– N_2 (g) system, determined by Karpinski et al. [25,26].

Therefore, the free energy of GaN constituents at their normal states, Ga and N_2 , is quite close to that of the crystal. This is illustrated in Fig. 2, in which the free energy of GaN (1 mole) and the free energy of the system of its constituents ($Ga+1/2N_2$) is shown as a function of temperature and N_2 pressure. With increasing temperature, the composite Gibbs free energy of the constituents $G_{Ga+1/2N_2}(T)$ decreases faster than $G_{GaN}(T)$ of the crystal, and at higher temperatures, the nitride becomes thermodynamically unstable. The crossing of $G(T)$ curves determines the equilibrium temperature at which GaN coexists with its constituents at given N_2 pressure. The application of pressure increases the free energy of the constituents to a much higher degree than $G(T)$ of the crystal. As a consequence, the equilibrium point shifts to higher temperatures and GaN stability range is extended.

In fact, for typical MBE growth conditions GaN begins to decompose rapidly at temperatures above 800 °C, restricting the epitaxy to temperatures much below the optimum point. Indeed, due to the arguments given above and the low diffusivity of N adatoms, the early efforts to grow GaN in MBE at temperatures below 800 °C, using group V-rich conditions typical for III–V epitaxy, gave unsatisfactory results. These early difficulties, seemingly well grounded in the simple thermodynamics of the processes involved, led many to believe that the only path for successful growth of nitrides in MBE reactors is to push the growth conditions as close as possible to those present in MOVPE reactors.

The breakthrough in the study of growth kinetics in PAMBE came with the finding that, in Ga-rich conditions, it is possible to grow relatively smooth layers at low growth temperatures [7,10–14,16]. However, such growth conditions were prone to the formation of Ga droplets on the GaN surface and high-quality material was mainly formed in the regions between the droplets. Further study of the Ga auto-surfactant effect revealed that this problem could be avoided provided the Ga to N flux ratio was maintained in a very narrow range of values: low enough to be just below the formation of the droplets, but high enough to ensure the formation of a metallic Ga bilayer on the Ga polarity surface [10–14,16]. As an example we demonstrate in Fig. 3, the PAMBE-grown GaN surface morphology for N-rich (rough surface—3D growth) and Ga-rich (smooth surface—2D growth) conditions and growth temperature 710 °C.

This apparent conundrum of finding 2D step-flow growth at temperatures significantly lower than $0.5 \cdot T_M$ can be resolved by considering that the decomposition of GaN (and therefore the

Fig. 2. Gibbs free energy of GaN and its constituents as a function of temperature: Ga+N₂ GaN—thick solid line, the constituents: Ga+N₂— $p=1$ bar dash—dotted line, $p=1$ kbar—dotted line, $p=10$ kbar—dashed line, and $p=20$ kbar—solid line.

Fig. 3. The scanning electron microscope image of GaN layers grown by PAMBE. (a) Nitrogen-rich conditions—3D growth mode and (b) Ga-rich conditions—2D growth mode.

Fig. 4. (a) A schematic diagram showing reaction of decomposing GaN with and without catalyst (after Newman [30]) and (b) with liquid Ga, the surface decomposition rate is much faster.

activity of atomic kinks) is controlled by a kinetic barrier (see Fig. 4). Schoonmaker et al. [28] demonstrated over 40 years ago that the height of this barrier is strongly reduced when GaN is covered by liquid Ga or In (Fig. 4). It was also experimentally evidenced and discussed by others [29,30] that, for a given temperature, the presence of a metal layer on a GaN surface accelerates etching of GaN in comparison to the exposed surface. Therefore, in the metal-rich regime for PAMBE (we can call it liquid phase epitaxy-like conditions) the formation energies of the kink sites are reduced, substantially lowering the minimum growth temperature necessary to sustain the desirable 2D step-flow growth mode.

Another important process necessary for high-quality growth at low temperature, which has also been a subject of intensive theoretical studies, is the adatom diffusivity. Perhaps the most insightful were recent results of modeling based on the density-functional theory, which pointed towards the existence of a very efficient lateral diffusion channel for nitrogen adatoms on semiconductor surface just below the thin metallic film [8]. Surprisingly, small activation energies for this so-called adlayer enhanced lateral diffusion (AELD) facilitate high-quality step-flow epitaxy at low temperatures. This enhanced surface

mobility of nitrogen adatoms, coupled with the reduction in the kink-formation energy discussed earlier, promotes 2D growth nucleation and step-flow growth at temperatures far below those needed for “classical” group V-rich conditions.

The importance of the two mechanisms described above, i.e. kink activity and adatoms mobility, is well illustrated by recent results demonstrating that high-quality GaN can be grown with PAMBE under “classical” N-rich conditions, *in the absence of metallic layer on the surface*, provided that growth temperature is high enough to cause concurrent decomposition of GaN [31]. It should be emphasized nevertheless that such growth conditions are not suitable for the epitaxy of InGaN because of the prohibitively high evaporation rate of In, and are likely to have only limited use for the epitaxy of AlGaIn, because of the pinning of the atomic step kinks on Al atoms (due to much higher AlN bond energy) and too low Al atoms surface mobility. Still, such growth conditions provide very useful alternative for PAMBE growth of high-quality GaN without the danger of Ga droplet formation.

4. The role of threading dislocations (TDs) and miscut angle on surface morphology in low-temperature PAMBE

It was found that for PAMBE, in the group III metal-rich regime, spiral growth is always present on substrates containing a high density of threading dislocations [12]. Such morphological features are located at the intersection of TDs with the crystal surface, and are formed in all crystal growth processes proceeding through 2D nucleation [32]. It is important to realize that such spiral growth leads to the formation of hillocks whose steepness increases monotonically with increasing tightness of the spiral. In Fig. 5, we present a comparison of the surface morphology of InGaIn layers grown by PAMBE on GaN/sapphire templates and on GaN bulk crystals grown by the high-nitrogen-pressure solution growth technique (TDs density $< 100 \text{ cm}^{-2}$). Nearly straight and parallel atomic steps are present on InGaIn layers grown on bulk substrate, while a dislocation-mediated step-flow growth mode is observed on the GaN/sapphire substrate. From our experience, at temperatures higher than 710°C PAMBE growth on high-dislocation density GaN/sapphire substrates results in nearly flat surfaces (hillocks height below 2 nm). However, as the growth temperature decreases, the tightness of the spirals and thus the height of the hillocks increases rapidly. Indeed, at temperatures around 600°C (where InGaIn layers are typically grown) hillock heights can exceed 7 nm. As a consequence of the non-uniform dislocation distribution in the substrate, the hillock diameters and sidewall inclinations vary substantially across the surface. Such local disorder on the surface acts as a seed for “catastrophic”

Fig. 5. (a) Atomic force microscope images of $\text{In}_{0.02}\text{Ga}_{0.98}\text{N}$ layers on a bulk crystal and (b) on a GaN/sapphire template.

Fig. 6. Atomic force microscope images of $\text{In}_{0.02}\text{Ga}_{0.98}\text{N}$ layers grown by PAMBE on bulk GaN substrates with miscut orientation towards (a) $[1\bar{1}00]$ and (b) $[11\bar{2}0]$, respectively.

degradation of epitaxial layers, particularly when the layers exceed the critical thickness. Thus, low dislocation substrates should enable the growth of metastable layers, allowing the system to “supercool” considerably before various undesirable crystallographic and morphological transitions are triggered (e.g., inversion domains, InN precipitates, Mg precipitates, dislocation clusters, etc.). Importantly, even though metastable, such layers can be extremely robust once grown, due to the very high melting temperatures for GaN and related compounds.

As a consequence of the GaN wurzite symmetry, growth processes are highly anisotropic, impacting atomic step morphology. In Fig. 6, we present a comparison of the morphology of InGa N layers grown under identical conditions on two bulk GaN substrates having different miscut orientation: towards $[1100]$ (Fig. 6a) and $[1120]$ (Fig. 6b) directions. For miscut towards $[1100]$ we observe nearly straight parallel atomic steps, while for $[1120]$ the atomic steps are considerably more disordered, showing step meandering, e.g. atomic steps with characteristic zigzag shape [33].

5. Growth of InGa N by PAMBE

The growth of efficient, high In content InGa N quantum wells (QWs), required for green LEDs and LDs, continues to be a subject of intensive study [34–36]. Considerable technological hurdles encountered during the growth of high In content layers result in the rapid decrease of performance of light emitters with

increasing emission wavelength. Until now only 488 nm LDs have been demonstrated in nitride-based structures by MOVPE [35]. The melting point for InN crystal is much lower than for GaN—recent experimental data indicate that it can be around 1800 °C [24]. Therefore for the InGa N growth, low temperatures are essential to avoid InGa N decomposition. As discussed earlier, PAMBE relies on a very different growth mechanism than MOVPE, one enabling deposition of device-quality nitride structures at considerably lower temperatures (by 200–300 °C compared to MOVPE). This makes PAMBE technology particularly suitable for the growth of strained, high In content structures, required for green emitters. It has already been demonstrated that even with generous oversupply of In, its incorporation into $\text{In}_x\text{Ga}_{1-x}\text{N}$ layers during the PAMBE process can be limited by (i) the growth temperature T_G , (ii) the Ga flux Φ_{Ga} , and (iii) the active nitrogen flux Φ_{N} [37–42]. Lowering of T_G has been used extensively to increase the In content of InGa N layers. Indeed, at sufficiently low growth temperature and fixed Φ_{N} , the In content x in $\text{In}_x\text{Ga}_{1-x}\text{N}$ layer is limited only by Φ_{Ga}

$$X = 1 - \Phi_{\text{Ga}}/\Phi_{\text{N}}, \quad \text{for } \Phi_{\text{Ga}} < \Phi_{\text{N}} \quad (5.1a)$$

and

$$X = 0, \quad \text{for } \Phi_{\text{Ga}} > \Phi_{\text{N}} \quad (5.1b)$$

This is because Ga efficiently displaces In from InN bond as a result of considerable gain in the bond energy. However, for high In content, the temperature necessary to uphold this simple equation is too low to ensure high-quality epitaxy. With

increasing growth temperature another effect becomes very important, namely thermal cracking of relatively weak In–N bonds. The rate of this process is influenced by number of factors such as growth temperature, presence of metallic layer on the surface, surface orientation, and whether the grown InGaN layer is strained or relaxed. In general, though we expect a thermally activated rate of In loss from the layer for fixed growth temperature this decomposition rate will be increasing with increasing effective layer composition [38]. This will result in less In being incorporated during the steady-state growth conditions, decreasing the effective growth rate and leading to In content lower than expected from Eq. (5.1a).

The impact of this process on final structures can be very significant. To illustrate that, we discuss below the growth of InGaN MQWs at $T_G=600^\circ\text{C}$. For that particular structure we used $\Phi_{\text{Ga}}=0.62\text{ nm/min}$ ($\Phi_{\text{Ga}}/\Phi_{\text{N}}=0.15$) for the wells and $\Phi_{\text{Ga}}=3.98\text{ nm/min}$ ($\Phi_{\text{Ga}}/\Phi_{\text{N}}=0.97$) for the barriers. The same Φ_{N} of 4.1 nm/min was used for both wells and barriers, while Φ_{In} was kept at level ensuring continuous metallic In surface coverage during growth. In the “classical” group III-rich conditions all active nitrogen atoms would have been incorporated. The deposition times were chosen in such a way that the targeted barrier and well width would have been 7 and 10 nm, respectively, for such growth conditions (i.e. the same growth rate of 4.1 nm/min for both wells and barriers). However, for low Ga flux and $T_G=600^\circ\text{C}$, we expect

that above-discussed decomposition of InGaN will be significant, leading to substantial reduction of the growth rate. Indeed, the TEM picture presented in Fig. 7 indicates that thickness of the barriers is 7 nm—as targeted, but for the wells we observe drastic reduction of their thickness—down to 2.5 nm, i.e. 4 times smaller than the target value. It means that the growth rate for wells was reduced to about 0.25 of the growth rate for barriers, or that 75% of active nitrogen atoms were not used for the well growth despite the abundance of In atoms on the surface. The data obtained by TEM is confirmed by XRD scans, where measured period is indeed equal to the sum of barrier and well widths. Having the barrier and well thickness from TEM we were able to determine In content (equal to 15% for wells and 2% for barriers) using XRD results with the assumption that InGaN QWs are fully strained. Notably, for the case of well growth the actual In content is only 15%, while 86% would be expected for the growth at very low temperature. It is a very intriguing example where the MBE growth is performed in excess of both group III (In) and group V (N) group elements. In this respect, such growth conditions indeed resemble liquid phase epitaxy.

The high structural quality of the epitaxial growth of InGaN on GaN bulk crystals using such LPE-like growth conditions, evident from the TEM image shown in Fig. 7, is further corroborated by AFM data shown in Fig. 6a, obtained for the same sample. Also, it is very important that at such growth conditions we consistently obtain structures with low dislocation density. The number of the threading dislocations in the final layer, as revealed by defect selective etching (DSE), is below 10^4 cm^{-2} [22]. The dislocations are concentrated in certain regions, suggesting that their origin is related to substrate preparation rather than to the growth itself.

To move the emission of QWs from blue–violet to green region, we increased the In content in wells by reducing the growth temperature. In Fig. 8a, we present photoluminescence (PL) spectra of 10 QWs structures (shown schematically in Fig. 8b). By optimization of the growth conditions we were able to increase internal quantum efficiency to 12% for 500 nm [41]. Alternatively, the In content can be increased by increasing the active nitrogen flux. These two parameters (T and Φ_{N}) permit effective optimization of growth condition for fairly wide range of In compositions.

6. Laser diodes by PAMBE

Laser diode structures were grown on (0001) Ga-polarity, conductive, low dislocation density, high-pressure-grown GaN

Fig. 7. Transmission electron microscope image of the MQWs grown with Ga/N=0.15 for wells and Ga/N=0.97 for barriers.

Fig. 8. Photoluminescence spectra taken at $T=300\text{ K}$ of $\text{In}_x\text{Ga}_{1-x}\text{N}/\text{In}_{0.02}\text{Ga}_{0.98}\text{N}$ multiquantum wells grown on bulk GaN crystals (a) and the structures of grown samples (b). The indium content x is equal to 0.04, 0.12, 0.24, and 0.27, respectively.

Fig. 9. (a) L – I – V characteristics of PAMBE-grown CW laser diode and (b) laser diode structure.

Fig. 10. Transmission electron microscope image of the active region of a PAMBE laser diode with 3 QWs.

bulk substrates. The 40 nm GaN:Si buffer layer and 450 nm Al_{0.08}Ga_{0.92}N:Si cladding were grown under Ga-rich conditions at 720 °C. The bottom waveguide, MQWs, electron blocking layer (EBL), top waveguide, top cladding, and contact layer were grown under In-rich conditions at 600 °C. The active region consisted of five 3-nm In_{0.1}Ga_{0.9}N wells with 7-nm In_{0.02}Ga_{0.98}N barriers (see Fig. 9). The devices were processed as ridge-waveguide, oxide-isolated lasers. The mesa structure was etched to a depth of 0.3 μm. The 20-μm-wide and 500-μm-long stripes were used as laser resonators. The oxidized Ni/Au ohmic contacts were deposited on the top surface of the device, and Ti/Au contacts were deposited on the backside of the highly conducting n-GaN substrate crystal. The cleaved laser mirror facets were coated with symmetrically reflecting mirrors. Fig. 9a shows the light-current-voltage (L – I – V) characteristics of the CW LDs with lasing threshold current density and voltage of 5.5 kA/cm² and 5.7 V, respectively. Lasing was observed up to 60 mW of optical output power (30 mW per facet) at a wavelength of 411 nm [19]. This confirms that growth of high-quality layers by PAMBE has been achieved; low dislocation density substrates are required to avoid defect formation that would otherwise occur even under optimized 2D growth conditions. When low TDs density GaN bulk substrates were used, the two-dimensional step-flow growth

mode with parallel atomic steps was observed in PAMBE layers even at low growth temperatures. Therefore, the smooth interfaces required for LDs can be obtained as shown in Fig. 10.

7. Towards green emitters

Optimizing quantum efficiency of LEDs is an effective strategy when preparing for more demanding laser diodes. The LEDs were grown on 0.5° miscut GaN substrates. A 100-nm InGaN:Si buffer layer was followed by 3-nm wells and 10-nm barriers, the 14-nm In_{0.02}Al_{0.16}Ga_{0.82}N:Mg electron blocking layer, and 200-nm In_{0.02}Ga_{0.98}N:Mg layer. The barriers were Si-doped at 2×10^{18} cm^{−3}, while the EBL and upper In_{0.02}Ga_{0.98}N:Mg layers were Mg-doped at 10^{20} cm^{−3} [43]. The LED structures were grown at the growth conditions established for MQWs. The light-current (L – I) characteristic of one of our LEDs (obtained on a nonprocessed wafer structure) is presented in Fig. 11. The insets to this figure show details of the sample-detector configuration for electroluminescence (EL) experiments and the EL spectrum. The light was collected from the bottom side of the substrate (passing through GaN substrate) and measured by a silicon detector attached to the bottom side of the wafer. The output power for this LED exceeded 3 mW at 100 mA and 450 nm. The nonlinear behaviour of L – I characteristic for high currents is in our case a combination of two effects: (i) so-called droop effect [44] manifesting as drop in quantum efficiency for large current injection and (ii) heating of the indium p-type top contact. The optical power curves for LEDs grown by PAMBE measured at 20 mA for several wavelengths spanning from blue-violet to green are shown in Fig. 12. We have obtained high optical output power of 1.5 mW at 20 mA for blue-violet LED. This power is reduced to 0.1–0.2 mW for the green LEDs. The reduction of the EL emission for the LEDs grown by PAMBE is similar to that observed on structures grown by the MOVPE technique [36]. Deterioration of the structural quality of the quantum wells with increased In composition, reported for MOVPE [34], is not likely to be the cause here since TEM, XRD, and AFM studies done on our LED layers give no indication of such problems. One likely explanation is presence of large built-in electric fields for the layers grown along the polar c -axis. The strength of these fields increases drastically for high In content QWs, leading to the reduction of oscillator strength and laser gain [45]. These issues can be effectively circumvented by

Fig. 11. The output power of LED grown by PAMBE as a function of current. The inset shows the electroluminescence spectrum.

Fig. 12. The optical power output for light-emitting diodes at current 20 mA as a function of the wavelength λ .

switching to non-polar or semi-polar substrates [46,47]. Indeed, it is widely believed that low-dislocation density non-polar or semi-polar GaN substrates can provide the key to successful fabrication of green LEDs [48,49]. Another candidate for the cause of the observed quantum efficiency drop with increasing emission wavelength could be non-radiative recombination centers related to N vacancies (V_N). Such correlations have been observed for MOVPE growth of laser diodes [50]. Indeed, high In content InGaN layers require high In precursor overpressure during the growth. For insufficient ammonia supply, the concentration of V_N related recombination centers may increase drastically in the grown film. More studies, e.g. positron annihilation experiments, should elucidate whether V_N are present in PAMBE high In content layers. Finally, one should consider the acceleration of Auger recombination processes with narrowing the bandgap of the well material, which would lead to the decrease in the quantum efficiency for radiative recombination, particularly at high injection currents [51]. However, importance of Auger recombination for InGaAs-based devices is still very controversial, with the most recent fully microscopic many body calculations showing that Auger processes are unlikely to play a role for this material system [52].

As can be seen from the above considerations, regardless of the actual reason for the presently encountered difficulties when moving towards the green emission, the long-term prospects for PAMBE in this area appear to be as good as, or better than prospects for MOVPE. For nitride-based optoelectronic technology this is a very new situation.

8. Summary

The emergence of low-temperature PAMBE technology as a viable alternative for the fabrication of blue-violet laser diodes discussed in this paper, as well as comparable performance of green LEDs, may signal the end of the exclusive domination of MOVPE in this field of technology. This is highly desirable, since the state-of-the-art in this sector still lags considerably behind GaAs- or InP-based laser devices. The resulting competition is bound to accelerate the refinement of the technology, much as it did for the previously established material systems. Even more importantly, the combination of PAMBE with high-quality GaN substrates may open a pathway to the growth of high-performance InGaN-based green lasers due to its unique capability for 2D growth at low temperatures. This target currently appears to be incompatible with MOVPE growth.

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