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High-pressure crystallization of GaN for electronic applications

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Abstract
The results obtained with the use of pressure-grown GaN single-crystalline substrates allow us to draw the following conclusions important for the construction of In-free UV light emitting diodes and lasers and InGaN-based high-power blue lasers.

(1) The application of pressure-grown GaN single-crystalline substrates allows us to grow near-dislocation-free layer structures by both metal–organic chemical vapour deposition and molecular beam epitaxy.
(2) The elimination of dislocations leads to highly efficient UV emission from GaN and GaN/AlGaN quantum wells, which is impossible for strongly dislocated structures grown on sapphire.
(3) At high excitations (e.g. in lasers), dislocations are also effective non-radiative recombination centres in InGaN-containing structures, so the elimination of these defects is crucial for better performance of blue lasers.

In this paper, the optical and structural properties of the near-dislocation-free GaN-based structures leading to the above conclusions are discussed.

(Some figures in this article are in colour only in the electronic version)

1. Introduction

Because of its large direct bandgap, high thermal stability, high electron velocities and other physical properties, GaN and its alloys with Al and In recently became the basic materials for short-wavelength optoelectronics and also for high-power, high-temperature electronic devices. The energy gaps in the considered compounds (6.2, 3.4 and 1.9 eV for AlN, GaN and InN respectively) cover the whole visible spectrum and a large part of the UV range. At present, high-brightness blue and green light emitting diodes (LEDs) and low-power blue laser diodes (LDs) are commercially available [1]. On the other hand, however, the development of GaN-based technology was, and still is, strongly limited by difficulties in obtaining large, high-quality crystals which could be used as substrates for epitaxial deposition of multilayer

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quantum structures necessary for devices. This is a direct consequence of thermodynamical properties of GaN [2] (and also AlN [3]), in particular its melting conditions, which are so extreme that the application of the common growth methods from stoichiometric liquids is technically impossible.

The spectacular development of GaN-based optoelectronic devices was possible due to elaboration of the two-step metal–organic chemical vapour deposition (MOCVD) processes for the growth of both n- and p-type (GaAlIn)N epitaxial structures on highly mismatched sapphire substrates [1, 4]. In these structures dislocation densities are as high as 10^8–10^{10} cm^{-2} but, nevertheless, very efficient luminescence is possible, especially if the structures contain InGaN [5]. It was therefore suggested that GaN and its alloys are much more 'tolerant' to the presence of structural defects than the conventional III–V systems due to the much lower minority carrier diffusion length and the relative immobility of structural defects in these strongly bonded crystals [6]. It is however well established [7] that the structural defects present in heteroepitaxial nitrides influence the optical and electrical properties of GaN-based epitaxial layers and quantum structures, which lowers the efficiency and lifetime of some important classes of devices.

The dominant extended defects in the typical heteroepitaxial nitrides are threading dislocations (TDs) and, associated with them, so called V defects, being the perturbation of the planar growth of InGaN and AlGaN due to the formation of low-energy [10, 11] planes around dislocations. The role of the TDs in the non-radiative recombination processes in GaN and InGaN has been revealed by e.g. the cathodoluminescence (CL) mapping correlated with the TEM imaging [8]. These findings were supported by theoretical calculations [9] showing that some types of TD in nitrides are electrically and optically active.

The extremely high efficiency of the luminescence observed for InGaN, despite a very large number of dislocations, is often explained by the strong localization of carriers in deep potential wells caused by compositional fluctuations in InGaN alloys, which eliminates interaction of the carriers with dislocations [5]. The size and the nature of these fluctuations are still not precisely defined. This is difficult since the incorporation of indium into the GaN lattice is a very unstable process (due to the instability of InN at high temperatures and the miscibility gap in the GaN–InN system), very sensitive to the growth mechanisms and conditions. Thus InGaN samples of very different microstructures are grown in different laboratories.

At high excitations however, even for In-containing structures, dislocations acting as the non-radiative recombination centres [5] quench the luminescence and the efficiency of light emitting devices is significantly limited. Therefore, in contrast to the InGaN-based LEDs, for the LDs dislocations limit both the efficiency and the lifetime of devices, and this is the most important obstacle in the construction of high-power blue lasers.

The epitaxial growth techniques for III–V nitrides are already very well developed, and it seems that the real technological challenge at present is to obtain large, high-quality GaN and/or AlN crystals. As shown below, crystallization of GaN from solutions, at high N_2 pressure, leads to perfect, dislocation-free material. The only drawback of the high-pressure growth technique is the relatively small size of the crystals. Therefore, methods to overcome this problem are proposed.

2. Thermodynamical properties of GaN

Table 1 compares melting temperatures and pressures of most typical semiconductor materials. The melting temperature T^M, and the corresponding equilibrium pressure at melting, of GaN as for other III–N compounds has not been measured. The melting temperature has been calculated by the use of Van Vechten's quantum dielectric theory of chemical bonding [10].

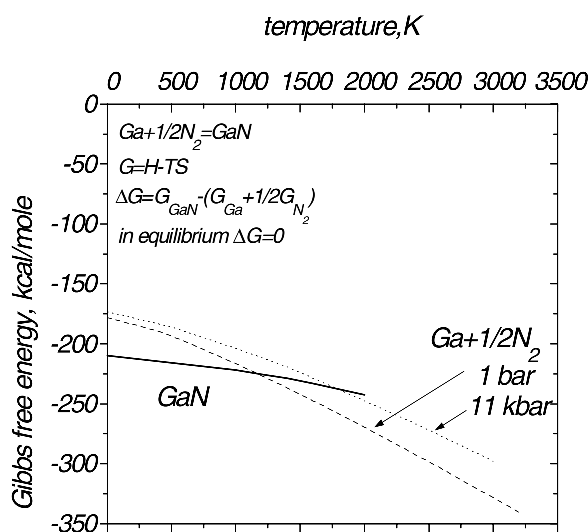


Figure 1. Gibbs free energy of GaN and its constituents.

Table 1. Melting conditions of semiconductors.

Crystal	T^M (°C)	p^M (atm)
Si	1400	<1
GaAs	1250	15
GaP	1465	30
GaN	2500	45 000
Diamond (synthesis)	1600	60 000

The corresponding pressure follows from the extrapolation of the experimental equilibrium data [2, 11]. The table shows that both temperature and pressure at melting of GaN are much higher than for typical semiconductors. They are rather similar to the conditions used for high-pressure synthesis of diamond.

Gallium nitride is a strongly bonded compound (with bonding energy of 9.12 eV/atom pair [12]) in comparison with typical III–V semiconductors such as GaAs (bonding energy of 6.5 eV/atom pair). 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). Therefore, the free energy of GaN constituents at their normal states, Ga and N_2, becomes quite close to that of the crystal. This is illustrated in figure 1, where 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, $G(T)$ of the constituents decreases faster than $G(T)$ of the crystal, and at higher temperatures the nitride becomes thermodynamically unstable. The crossing of $G(T)$ curves determines the equilibrium temperature where 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 the GaN stability range extends. The equilibrium p_{N_2}–T conditions for GaN have been studied by several groups [2, 11, 13–15]. The most complete and consistent

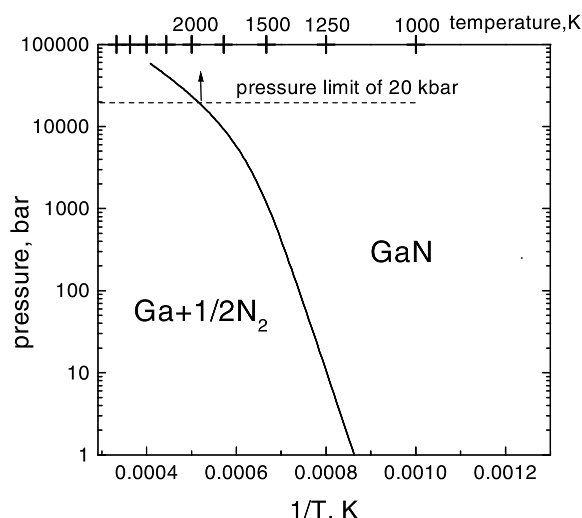


Figure 2. Equilibrium curve for GaN [2, 11].

results have been obtained by Karpinski *et al* [11, 12] by direct synthesis and decomposition experiments performed by both the gas pressure technique (for pressures up to 20 kbar) and the high-pressure anvil technique (up to 70 kbar). The curve following from these data is shown in figure 2. Crystallization processes discussed in this paper have been carried out at N_2 pressure up to 20 kbar, which corresponds to a GaN stability limit of 1960 K.

The extension of the GaN stability range by the application of pressure allows us to grow GaN crystals from solution in liquid Ga. Even the highest available temperature of 1960 K is quite far from the melting temperature of GaN (table 1). Therefore the N concentrations are not high (below 1 at.%) and the growth experiments have to be long to obtain high-quality crystals with dimensions appropriate for research and applications. Therefore the long time ($>$100 h) of stable work is an additional requirement for the growth system.

The analysis of thermodynamical properties of the system of GaN and its constituents explains the role of high pressure as a factor increasing the thermodynamical potential of the constituents, which makes the crystal stable at temperatures necessary for crystallization. The pressure however is also important for kinetics of GaN synthesis. The synthesis of GaN from its constituents is possible due to dissociative chemisorption of the N_2 molecule on the Ga surface as shown by Krukowski *et al* [16] by the density functional theory calculations. The nitrogen molecule dissociates on the Ga surface only if it overcomes the potential barrier of about 3.5 eV. Therefore, as shown in [16], the density of the interacting gas is of crucial importance for the Ga–N_2 system. The molecular nitrogen, being in contact with gallium, dissociates at the surface and dissolves in the metal in the atomic form.

Both the increase of the thermodynamical potential of gaseous nitrogen and the enhancement of the creation of the atomic nitrogen at the Ga surface can be achieved by other (not compression) methods, such as the excitement of N_2 plasma or the use of species containing nitrogen atoms bonded more weakly than in the N_2 molecule. It was demonstrated [18, 19] that in a microwave nitrogen plasma of 60 Torr and at temperature as low as 1100 °C, a very efficient GaN (and even InN!) synthesis is possible. The excited N_2 gas was a very efficient source to saturate a Ga droplet with atomic nitrogen and to form a GaN crust on the droplet surface. However, since the pressure in the system was much lower than the equilibrium one

Figure 3. The result of GaN synthesis from the liquid Ga and N_2 plasma. The liquid has been pushed out from the initial droplet covered by GaN crust by the growing N_2 bubbles.

(about 100 bar for 1100 °C), the N_2 gas phase started to nucleate and grow in the liquid. This resulted in the formation of the structure shown in figure 3, which was the empty 'Ga droplet' covered with irregular polycrystalline GaN.

3. Experimental details

At present, GaN is crystallized in gas pressure chambers with volume up to 1500 cm^3 allowing crucibles with the working volume of 50–100 cm^3. The high-pressure–high-temperature reactor consisting of the pressure chamber and the multizone furnace is equipped with additional systems necessary for: *in situ* annealing in vacuum, electronic stabilization and programming of pressure and temperature and cooling of the pressure chamber. Pressure in the chamber is stabilized with precision better than 10 bar. The temperature is measured by a set of thermocouples arranged along the furnace and coupled with the standard input power control electronic systems based on Eurotherm units. This allows stabilization of temperature $\pm 0.2°$ and programmable changes of temperature distribution in the crucible.

GaN crystals presented in this paper were grown from the solutions in pure liquid gallium and in Ga alloyed with 0.2–0.5 at.% of Mg or Be at pressures in the range of 10–20 kbar and temperatures of 1400–1600 °C. Magnesium and beryllium, as the most efficient acceptors in GaN, were added to the growth solutions in order to reduce the concentration of free electrons in the crystals by compensation of residual donors.

The supersaturation in the growth solution has been created by the application of a temperature gradient of 2–20 °C cm^{-1} along the axis of the crucible. This method was chosen since the axial temperature gradients in multizone furnaces, working at high gas pressure, can be controlled with high precision and the method assures a continuous flow of nitrogen from the hotter part of the solution to the cooler one. If the crystallization experiments were performed without an intentional seeding, the crystals nucleated spontaneously on the internal surface of the polycrystalline GaN crust covering liquid Ga at the cooler zone of the solution. The typical duration of the processes was 120–150 h.

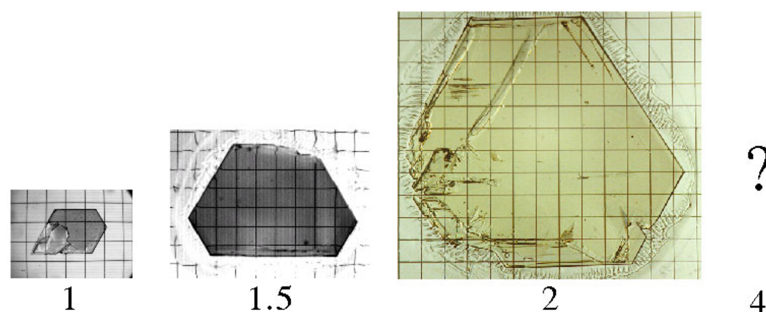


Figure 4. GaN crystals grown in high-pressure chambers of different sizes. The numbers are proportional to the diameters of the chambers. The distances between grid lines correspond to 1 mm.

4. Crystals grown without intentional seeding

The GaN crystals grown by the high-nitrogen-pressure solution method are of wurzite structure, mainly in the form of hexagonal platelets. The large hexagonal surfaces correspond to $\{0001\}$ polar crystallographic planes. The side faces of the crystals are mainly the polar $\{10\bar{1}1\}$ and also non-polar $\{10\bar{1}0\}$ planes.

The crystals in the form of hexagonal platelets grown slowly, with a rate <0.1 mm h^{-1} in $\{10\bar{1}0\}$ directions (perpendicular to the c-axis), are usually single crystals of perfect morphology suggesting stable layer-by layer growth. They are transparent, with flat mirror-like faces. The habit of the crystals does not change for solutions containing Mg or Be. The average size of crystals grown without an intentional seeding scales with the diameter of the high-pressure reactor as shown in figure 4. As one can deduce from the form of the crystals, the growth is strongly anisotropic, being much faster (about 100 times) in directions perpendicular to the c-axis. This relation is valid at supersaturations corresponding to the average growth rate in $\{10\bar{1}0\}$ directions of 0.05–0.1 mm h^{-1}.

5. GaN crystals—intentional seeding

The thin GaN platelets described in the previous section can be used for further crystallization as seed crystals. Seeded growth in directions parallel to the c-axis seems to be much more challenging, since the growth rates observed in spontaneous crystallization are small (1–2 μm h^{-1}) and the growth shows strong tendencies to be unstable. Directional crystallization in the vertical configuration, in stabilizing the temperature gradient perpendicular to the growth front, gives promising results in terms of the growth rates (over 10 μm h^{-1}), growth stability and the quality of the material (it is nearly dislocation free [19]). An example of GaN grown on the (0001) surface of a GaN substrate is shown in figure 5. The 200 μm thick GaN platelet obtained by slicing of the new grown material is also shown in the figure.

6. Physical properties of pressure-grown GaN crystals

The synthesis of GaN from its constituents is possible due to the dissociative chemisorption of the N_2 molecule on the Ga surface [20]. The nitrogen molecule dissociates on the Ga surface only if it overcomes the potential barrier of about 3.5 eV. For oxygen interacting with Ga there is no potential barrier for dissociation [21], and therefore even traces of this impurity in the growth

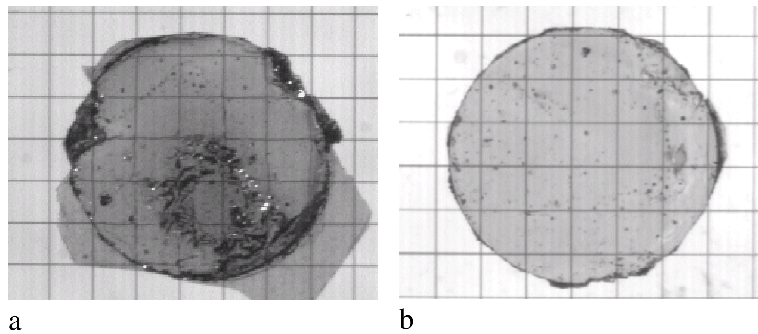


Figure 5. Directional crystallization of GaN: (a) bulk GaN grown on the (0001) Ga-polar surface of a GaN pressure-grown substrate crystal; (b) a slice of GaN grown on the substrate.

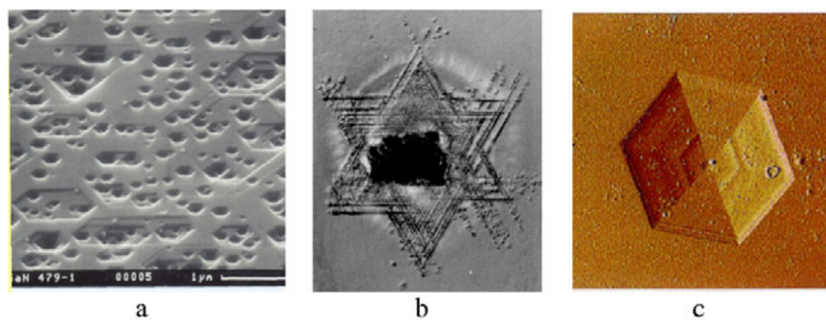
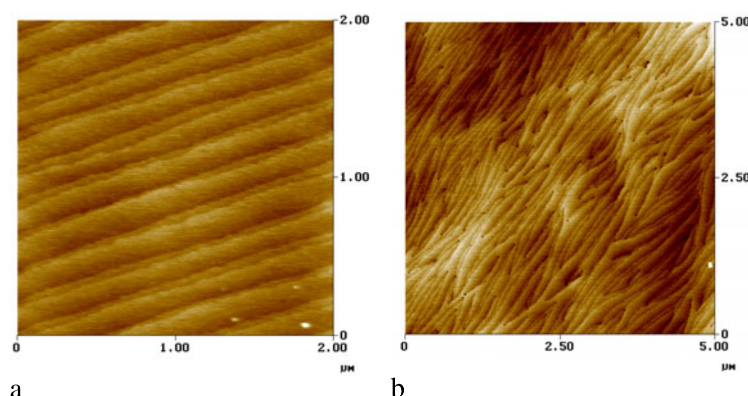


Figure 6. Defect-selective etching of GaN: (a) GaN/sapphire heteroepitaxial layer after (b) GaN pressure-grown single crystal after indentation with diamond and etching—the average size of the starlike pattern is 100 μm; (c) etch pit (EP) on the GaN crystal surface—observed density of EPs 10–10^2 cm^2.

system are a source of unintentional oxygen doping of GaN. Therefore the crystals are strongly n-type with a free electron concentration of about 5×10^{19} cm^{-3} (metallic conductivity) and a mobility of about 60 cm^2 V^{-1} s^{-1} [22]. These free carriers can be fully eliminated by Mg or Be acceptors added to the growth solution. Then the resistivity of the crystals becomes as high as 10^4–10^6 Ω cm at 300 K. Usually the GaN:Mg crystals become p-type with an activation energy of 150 meV, at temperatures slightly exceeding 300 K [23]. The GaN:Be crystals are also highly resistive, with the constant activation energy of 1.46 eV at temperatures up to at least 1000 K [24].

As a rule the extended defects are not observed by TEM in the crystals used as substrates for both N and Ga polarity epitaxial growth. Therefore, if the epitaxy is properly performed both the substrate and the layer quantum structure do not contain dislocations. Unfortunately, the TEM techniques allow us to analyse only a very small-area sample. Therefore, in order to measure dislocation densities in GaN, defect-selective etching methods have been developed [25]. It was shown that etching in molten KOH–NaOH eutectics reveals dislocations in both GaN heteroepitaxial layers and GaN pressure-grown single crystals. The examples are presented in figure 6.

Figure 6(a) shows the result of defect-selective etching of a typical GaN heteroepitaxial layer grown by MOCVD on a sapphire substrate. The high density of EPs is clearly visible. The same method applied for bulk crystals gives a very small (10–100 cm^{-2}) number of EPs. The



a b

Figure 7. Surface morphology of GaN epitaxial layers deposited by MOCVD [27] on different substrates: (a) GaN pressure-grown crystal, (b) sapphire.

typical pit observed on both heteroepitaxial layers and crystals is shown in figure 6(c). The pattern in figure 6(b) is the result of DSE of a GaN single crystal with dislocations generated intentionally by indentation with diamond [26]. The EPs surround the imprint of the diamond, showing the area where dislocations are present, and that the remaining material is dislocation free.

7. Homoepitaxy of GaN on pressure-grown GaN substrates

If the surface preparation and the conditions of the epitaxial growth are right, the structure of GaN homoepitaxial layers, in terms of dislocation density, follows the structure of the GaN substrates. Figure 7 compares the surface morphology of GaN epitaxial layers deposited by MOCVD on a GaN crystal and on a GaN/sapphire substrate, in the same run [27]. The atomic step flow on the surface of the homoepitaxial layer is not perturbed as observed for a GaN layer containing TDs. If a GaN homoepitaxial layer is nearly dislocation free, its optical properties depend mostly on the concentration and distribution of point defects incorporated during the epitaxial growth. This is a function of the purity of the growth system, the growth conditions and the orientation of the substrate.

Due to the lack of strains and the high degree of homogeneity of the homoepitaxial GaN, the exciton-related peaks in low-temperature PL spectra are usually very narrow. The FWHM of the bound exciton lines less than 0.5 meV is observed for layers grown by both MOCVD (see e.g. [28, 29]) and MBE (see e.g. [30, 31]) on the Ga-polar surfaces of the GaN substrates.

For GaN layers grown by MBE with an NH_3 nitrogen source, it was shown [31] that the band edge PL integrated intensity at RT for the homoepitaxial GaN was about 60 times larger than for GaN grown on sapphire at the same growth conditions. Also, the decay of the PL in these homoepitaxial layers was much slower than for corresponding (grown in similar conditions) heteroepitaxial GaN [31]. This indicated a substantial suppression of the non-radiative recombination by elimination of TDs from the layers.

The PL decay has been also studied at high excitations [32] for GaN layers grown by MOCVD on sapphire and pressure-grown GaN substrates. The decay time (at room temperature (RT)) of 450 ps measured for homoepitaxial material was five times longer than for a corresponding heteroepitaxial layer grown under the same conditions. Figure 8 shows the decay of spontaneous luminescence measured at high excitation, close to the stimulated emission threshold for homo- and heteroepitaxial GaN.

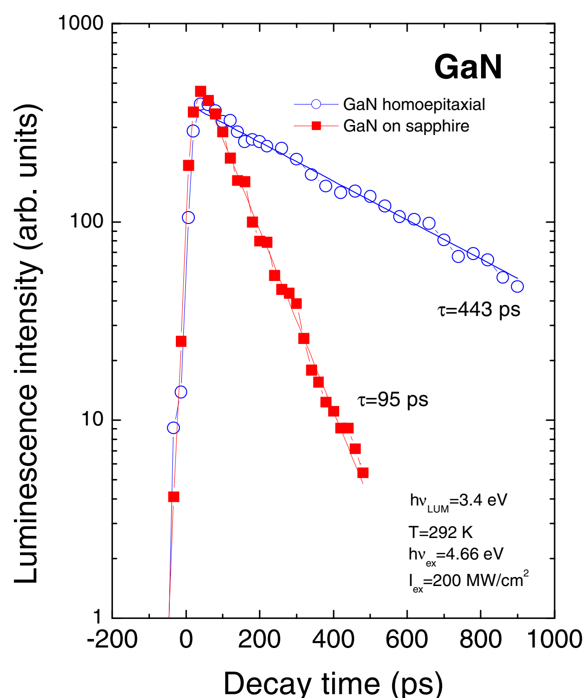


Figure 8. Spontaneous luminescence transients for GaN films grown on sapphire (■) and on GaN substrates (○) [32].

8. Quantum structures

The growth of GaN/AlGaN structures on GaN pressure-grown crystals has been studied mainly by MBE [33–36]. Near-dislocation-free GaN/AlGaN multi-quantum wells and structures have been grown on the N-polar surfaces of GaN crystals by RF plasma-assisted MBE (PA MBE) [33, 34] and on the Ga-polar surfaces by MBE with an NH_3 nitrogen source (R MBE) [31, 35]. As a rule, the optical properties of the structures without dislocations were much better than for similar structures grown under the same conditions on sapphire.

For example, in figure 9, the integrated intensities of the PL from similar structures (8 ML GaN SQW with $Al_{0.1}Ga_{0.9}N$ 50 nm barriers) grown by R MBE [31] on sapphire and bulk pressure-grown GaN substrates are compared.

It is shown that for the structure deposited on GaN the PL is much stronger, especially at RT. For the structure grown on GaN substrate the PL intensity starts to decrease with temperature only at about 100 K, mainly due to the thermal escape of carriers from the quantum well towards the AlGaN barriers, as observed for classical III–V QW heterostructures with dislocation densities lower than 10^3 cm^{-2} [36]. For a heteroepitaxial structure, the presence of dislocations strongly influences the non-radiative recombination processes. The PL intensity starts to decrease at much lower temperatures due to delocalization of excitons and their further interaction with dislocations.

For devices on dislocation-free substrates, it is especially important that further increase of the emission efficiency can be achieved by increasing the Al content in the barriers. This opens the possibility for construction of highly efficient UV LEDs and lasers.

The semi-insulating GaN:Mg substrates have been used for growth of GaN/AlGaN heterostructures with a two-dimensional electron gas (2DEG) by R MBE [35]. A Hall mobility

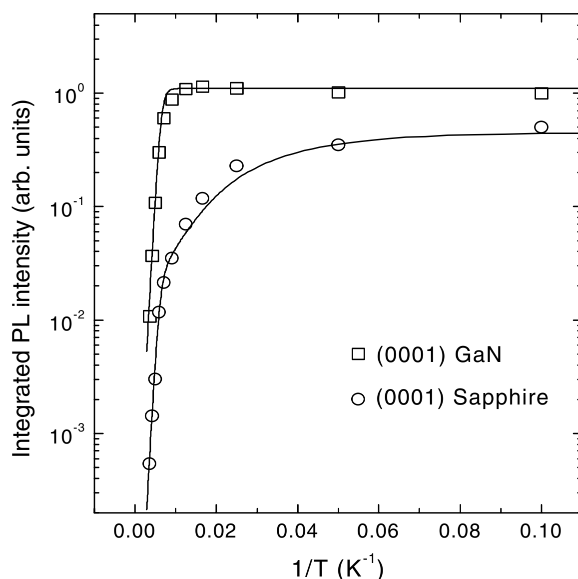


Figure 9. Temperature dependence of the integrated PL intensity of homoepitaxial (□) and heteroepitaxial (○) $GaN/Al_{0.1}Ga_{0.9}N$ QWs [31].

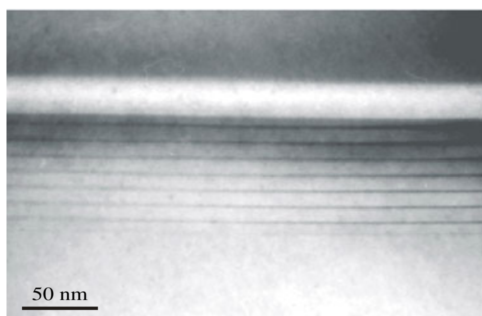


Figure 10. InGaN MQW grown on the Ga-polar (0001) surface of a GaN substrate by MOCVD [37], TEM courtesy of M Albrecht.

for the 2DEG as high as 60 100 $cm^2\ V^{-1}\ s^{-1}$ at 1.5 K (one of the highest reported for GaN) has been measured. The Shubnikov–de Haas oscillations starting at 1.8 T and well defined QHE indicated high quality of the MBE-grown heterostructure. From the analysis of the scattering processes it was suggested that in such a near-dislocation-free structure the ionized impurity scattering was the dominant low-temperature mobility limiting mechanism.

The InGaN near-dislocation-free epitaxial layers and quantum wells were grown on the pressure-grown crystals mainly by MOCVD by Leszczynski *et al* [37], in the High Pressure Research Centre Unipress, Warsaw, Poland. A standard 470 nm LED structure with a single InGaN quantum well has also been grown by Nakamura [39] on a pressure-grown crystal. In figure 10 the TEM images of one of the InGaN MQWs grown in Warsaw is shown to demonstrate that the structures are dislocation free.

The compositional homogeneity of these materials was extensively studied by Albrecht *et al* [38] using TEM and CL methods. It was shown that the InGaN microstructure is very

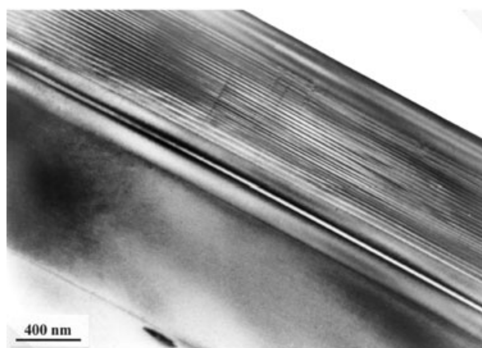


Figure 11. TEM image of the multilayer structure deposited on a GaN substrate by MOCVD; sequence of layers from the lower left corner: n-GaN, n-$Al_{0.11}Ga_{0.89}N$/n-GaN superlattice, n-GaN, $In_{0.09}Ga_{0.91}N$, p-GaN, p-$Al_{0.14}Ga_{0.86}N$/p-GaN superlattice, p-GaN.

sensitive to the growth conditions (temperature, growth rate and microscopic growth mode) and different InGaN microstructures can be observed even for samples of the same average In content. For example, for InGaN QWs with 15% In, the HRTEM tetragonal distortion analysis has shown either perfectly homogeneous distribution of In or 8–20% fluctuations of the In content with dimensions of a few nanometres. Nevertheless the optical properties of these QWs, with very different microstructures, were strikingly similar.

These studies allowed us to suggest the following.

- The strong localization of carriers in InGaN is not due to detectable compositional fluctuations, because the emission from the homogeneous InGaN is the same as that from InGaN with large fluctuations detected, with similar average In content.
- The presence of centres localizing the carriers is an intrinsic property of InGaN on the atomic level.

Additional support for this statement is that the emission peaks from InGaN of the same average composition are relatively broad (a few tenths of a millielectron-volt even for homogeneous, dislocation-free material) and of the same FWHM independent of the presence of detectable In content fluctuations.

It is obvious that for dislocation-free InGaN structures, especially those to be used for lasers, the optimum is the homogeneous distribution of In in the quantum wells. Any fluctuations, whatever their nature is (e.g. caused by composition or thickness inhomogeneities), are no longer necessary. To achieve a high degree of homogeneity in the InGaN layers the best solution seems to be the growth on slightly misoriented substrates. Then the continuous presence of the well determined steps on the surface allows us to limit the growth mechanism to the step flow mode, eliminating the necessity of two-dimensional nucleation, which could lead to both the segregation of In and less perfect interfaces between quantum wells and barriers.

Dislocations in the structures grown on the almost defect-free substrates can obviously appear as a result of the lattice mismatch between GaN and its ternaries InGaN and AlGaN. This was analysed by Leszczynski *et al* [36] by the x-ray measurements of lattice parameters of various GaN-based epitaxial layers deposited on GaN substrates as a function of InGaN (AlGaN) composition and thickness. It was shown that the boundary between fully strained and relaxed InGaN (AlGaN) layers is at a safe distance from the mismatch-thickness area for layers necessary for a typical blue laser structure. Figure 11 shows an example of the multilayer structure, similar to the full structure of a blue laser.

The example confirms that no mismatch dislocations are generated if the structure is grown on dislocation-free GaN substrate.

The pressure-grown GaN crystal has been tested by Nakamura [39] as a substrate for the InGaN 405 nm laser. Cw 30 mW devices with lifetime exceeding 3000 h were constructed. 30 mW power has been achieved at a current of 62 mA. For typical InGaN MQW LDs, at a current of 62 mA the output power was about 15 mW. This increase in the efficiency confirmed that dislocations are the main limiting factors for high-power GaN-based lasers.

If the substrate is dislocation free, the strain relaxation processes in low-misfit heteroepitaxy are unaffected by the presence of TDs. Both elastic and plastic relaxation have been studied [33] for GaN/AlGaN multilayer structures grown by plasma-assisted MBE on GaN pressure-grown substrates. It was shown that the crystal growth and relaxation in III nitrides is not different from that in conventional III–V systems. Elastic strain relaxation by sinusoidal undulation of the AlGaN layers has been observed above a critical thickness depending on the Al content. The results obtained in [33] indicate new possibilities for designing low-misfit heteroepitaxial growth including self-organized formation of quantum structures.

9. Conclusions

The perfect single-crystalline lattice-matched substrates for III nitrides will give new technological possibilities and will stimulate further development of III–N physics and technology. The epitaxial growth on such substrates, especially on their surfaces vicinal to the low-index crystallographic planes, should result in much more uniform layers regarding composition of ternary or quaternary compounds and both intentional and unintentional doping. It also should lead to the reduction of thickness fluctuations of thin layers constituting a quantum structure. Perfect interfaces are especially important for nitrides grown in polar (0001) directions, since then the fluctuations of the polarization-induced internal electric fields can be minimized. The availability of bulk GaN and AlN substrates oriented in non-polar directions will open a new way for determination of the role of polarization-related effects in nitrides. The progress in both AlN [40] and GaN [19] bulk crystallization suggests that this possibility will appear in the very near future.

The application of dislocation-free GaN substrates allows us to limit dislocation densities in the epitaxial GaN-based structures to very low values (lower than 10^2 cm^{-2}). In particular the full structures of InGaN-based blue lasers can be grown without creation of misfit dislocations. Thus the important centres of non-radiative recombination can be eliminated from both In-free and In-containing quantum structures, which is of crucial importance for efficient UV emitting devices and for both UV and visible GaN-based high-power lasers.

The absence of TDs from the substrates should allow a bandgap engineering based on the typical elastic and plastic strain relaxation processes as for classical low-misfit heteroepitaxial systems.

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