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Limitation of simple np-n tunnel junction based LEDs grown by MOVPE

Y. Robin^{1,*}, Q. Bournet², G. Avit¹, M. Pristovsek¹, Y. André³, A. Trassoudaine³ and H. Amano¹

¹ Institute of Materials and Systems for Sustainability (IMaSS), Nagoya University, Japan

² Polytech Clermont-Ferrand, Université Clermont Auvergne, France

³ Institut Pascal, Université Clermont Auvergne, CNRS, SIGMA Clermont, F-63000 Clermont-Ferrand, France

* Yoann.robin@crhea.cnrs.fr

We show evidence that tunnel junctions (TJs) in GaN grown by metal-organic vapor phase epitaxy are dominated by defect level-assisted tunneling. This is in contrast with the common belief that highly doped layers ($> 10^{20} \text{ cm}^{-3}$) are required to narrow the TJ space charge region and promote the band-to-band tunneling. Our conclusion stems from the study and the review of the major doping limitations of carefully optimized p^{++} and n^{++} layers. The secondary ions mass spectroscopy profiles of GaN based TJ LEDs show a strong oxygen concentration located close to the p^{++}/n^{++} interface, typical for three dimensional growth. In addition, considering the doping limitation asymmetry and Mg carry-over, our simulations indicate a depletion region of more than 10 nm which is buried in a rough and defective n^{++} layer. However, decent electrical characteristic of the studied TJ based LEDs are obtained, with a low penalty voltage of 1.1 V and a specific differential resistance of about $10^{-2} \Omega \cdot \text{cm}^2$ at 20 mA. This indicates that common TJ could be greatly optimized by using a moderate doping ($\sim 10^{19} \text{ cm}^{-3}$) while intentionally introducing local defects within the TJ.

I. INTRODUCTION

Realization of high-quality p-n homojunction based on wide bandgap semiconductors is challenging due to doping asymmetry problems [1]. GaN, with a bandgap of about 3.4 eV, does not deviate the rule. While doping n-type is rather simple, achieving decent p-type layers is not straightforward [2,3].

Intentional n-type doping of GaN is traditionally realized by substituting gallium by silicon, which allows to reach electron concentration up to 10^{19} cm^{-3} . However, at higher doping levels, Si doped GaN layers tend to roughen or crack under tensile stress [4,5]. Recently, germanium, whose size is comparable to gallium, has been studied as alternative dopant to cover the 10^{19} - 10^{20} cm^{-3} range [6-8]. Oxygen on N-site is an efficient n-type donor, although only few reports exist on intentional oxygen doping [9-11]. N-type doping of GaN by

other chalcogens (sulfur, selenium...) substituting the nitrogen site is little investigated. However, their huge size difference to nitrogen does not bode well for achieving high doping levels and the relatively low vapor pressures of these solid easily cause strong carry-over [12,13].

For p-doped GaN, there are even less suitable acceptors. Zinc, a standard dopant in other III-V materials, is still a relatively deep acceptor in GaN ($E_a \sim 300$ - 400 meV) [14]. Carbon exhibits an amphoteric behavior and beryllium is a shallow acceptor ($E_a \sim 100$ - 140 meV) but suffers from severe self-compensation [15-18]. Other unsuccessful attempts were made using calcium, cadmium and mercury [19]. In the end, magnesium is the only dopant that allows a decent p-type conduction after an activation. Unfortunately, the maximum hole concentration barely reaches 10^{18} cm^{-3} due to the large ionization energy ($E_a \sim 150$ - 200 meV) even though mid 10^{19} cm^{-3}

magnesium can be incorporated before onset of compensating defects [20-22].

Recently, tunnel junctions (TJs) have gained interest in nitrides. Indeed, such TJs reduce optical loss in UV-LEDs, alleviate non-equilibrium hole injection and allow a better current spreading in higher mobility n-type layers [23-24]. Such properties pave the way for inter-cavity contact in vertical cavity surface emitting lasers (VCSELs) and simpler vertical integration of tricolor micro-LEDs [25-28]. Esaki junctions typically require both low or moderate bandgap materials and doping levels around 10^{20} cm^{-3} that is out of reach for GaN [29]. Nevertheless, GaN based TJ based LEDs reported unexpectedly low penalty voltages [30-36]. The common explanation widely found in the literature considers that over-doping the n- and p-type layers, along with the insertion of a thin InGaN layer, reduces the width of the space charge region of the TJ and enhances band-to-band tunneling. The defects assistance is mostly considered as a minor or negligible contribution. In this paper, we show evidence that midgap defect level-assisted tunneling is the main mechanism responsible of the surprisingly high efficiency of TJ grown by metal-organic vapor phase epitaxy (MOVPE). In a preliminary study, we first summarize the general crystal properties of simple p- and n-GaN layers depending on growth conditions. We highlight the morphology changes, the secondary ion mass spectroscopy (SIMS) profiles of dopants and impurities, and the defects typically generated upon heavy doping in TJs. In a secondary part, we compare the electrical behavior of simple np-n structures and corresponding TJ based LEDs grown without using InGaN interlayer. Based on those findings and further band diagram simulation, we conclude that Poole-Frenkel effect is the main tunneling process involved in such TJs while band-to-band tunneling (Fowler-Nordheim) is much less likely given the large band gaps and relatively low carrier concentrations. Finally, we

discuss TJs found in the literature in the light of our observations and results.

II. EXPERIMENTAL DETAILS

All the samples were prepared in a 3x2" EpiQuest vertical showerhead MOVPE reactor. For Si and Mg doping, pure H_2 was used as carrier gas. Both Mg:GaN and Si:GaN layers were grown using a V/III ratio of 7200, at a pressure of 150 Torr and at temperatures ranging from 975 to 1100°C with a growth rate of about 500 nm/h. Bisethylcyclopentadienyl-magnesium (EtCp_2Mg), tetramethyl-silane (TMSi), trimethyl-gallium (TMG) and ammonia (NH_3) were used as metal-organic precursors. For the growth of the quantum wells (QWs), the carrier gas was switched to pure N_2 and triethyl-gallium (TEGa) and trimethyl-indium (TMIn) were used as precursors.

Electrical characterizations were performed by Hall effect measurements (HEM) in Van-der-Pauw geometry. For Si doping levels below 10^{19} cm^{-3} Hall effect measurements and secondary ion mass spectroscopy (SIMS) concentrations agreed. Above that limit, the dopant concentration was measured by SIMS or given as extrapolation of our calibration. Photoluminescence (PL) was performed at room temperature (RT) using a He-Cd laser ($\lambda_{\text{ex}}=325 \text{ nm}$, $P=1 \text{ mW}$). For the activation of the simple Mg-GaN layers, the samples were annealed at 700°C for 5 minutes under N_2 . Surface morphologies were investigated by a Nanocute Hitachi atomic force microscope (AFM) working in tapping mode and by Nomarski optical microscopy (OM).

TJ based LED structures were processed by standard photolithography. L-shape mesa of 0.0675 mm^2 were prepared using Cl_2 dry etching. Then, the devices were annealed for 30 min at 750°C in N_2 ambient in order to activate the buried p-GaN layers (notice it is different from the annealing conditions used to activate simple p-GaN

layers). Finally, Ti/Al electrodes were simultaneously deposited by e-beam evaporation on both n-GaN terminals and annealed 5 min at 625°C in pure N₂. For the reference LED, p-GaN top layer was contacted by a Ni/Au electrode annealed 10 min in air at 500°C.

III. RESULTS AND DISCUSSION

Fig. 1 summarizes the morphological, optical and electrical properties of 500 nm thick Mg doped GaN layers grown at various Mg/Ga ratios, which is in line with current literature. The near band edge (NBE) emission of the PL slowly decreases and a characteristic donor-acceptor pair (DAP) related signal appears while increasing the Mg content [37].

A strong blue-band (BB) dominates the PL spectrum when the hole concentration reaches its maximum around few 10¹⁷ cm⁻³ at RT, corresponding to a Mg concentration of few 10¹⁹ cm⁻³ [38]. Further increase of the Mg content leads to the apparition of various compensating defects and pyramidal inversion domains (IDs) [22]. Simultaneously, the doping level drastically decreases and the p-GaN layers become resistive. At a Mg concentration of a few 10²⁰ cm⁻³ - which is the doping level typically required for TJs - the surface is rough and decorated with large hexagonal N-polar hillocks likely induced by the formation of Mg₃N₂ sub-structures [39,40]. A high density of defects is indicated by the absence of any PL signal.

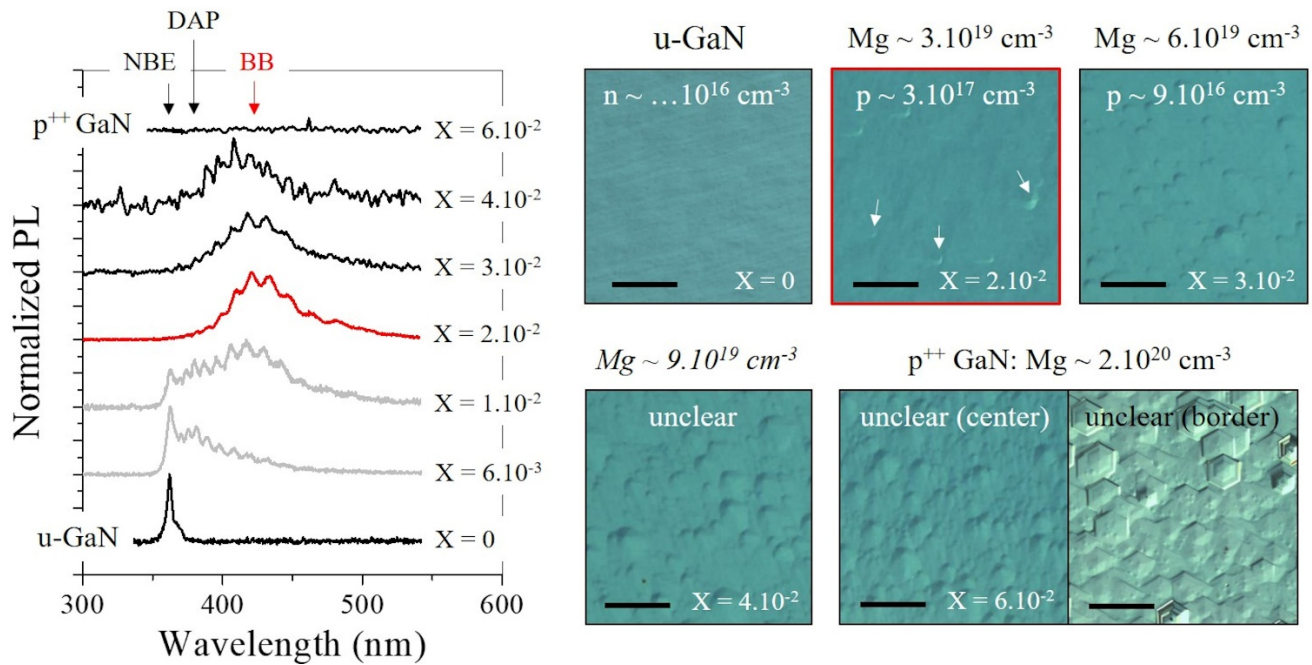


FIG. 1. Normalized PL emission spectra and associated optical microscope pictures of post-annealed GaN grown at 975°C at various X = Mg/Ga ratio. The red PL spectrum highlights the highest doping level achieved. The OM pictures related to the grey PL spectra are omitted. The scale bar corresponds to 50 μm.

Fig. 2 shows a similar study on 500 nm thick (crack-free) Si doped GaN layers grown at various Si/Ga ratio using TMSi. This Si precursor mostly decomposes above 1000°C, which is different from the common silane precursor, hence calibration was performed at two temperatures [41]. The

electron concentration increases with the Si/Ga ratio both for growth temperature of 975 and 1100°C. At Si concentration exceeding 1.5×10¹⁹ cm⁻³, the surfaces show large scale-like islands of different heights separated by pitted boundaries. Further increase of the Si level leads to

extremely rough surfaces characterized by larger pits and smaller grains, similar to previous reports [5,42]. Under typical MOVPE growth conditions and for a concentration higher than 10^{19} cm^{-3} , Si acts as an anti-surfactant. It was suspected to locally form Si-Ga-N substructures, which initiate a SiN_x nano-masking that promotes 3D growth [43,44]. According to high resolution transmission electron microscopy, the actual SiN_x masking is actually a SiGaN_3 monolayer that inhibits Ga nucleation and thus causing islands growth in uncovered areas [45]. Therefore, at the doping levels used for TJ application, i.e. above 10^{20} cm^{-3} , Si behaves as an anti-surfactant rather than a dopant, as it is incorporated inactively in SiGaN_3 . We observed that the 2D-3D transition always happens around a doping level of 10^{19} cm^{-3} regardless the growth temperature. Thus, the 2D-3D transition is driven by the actual Si concentration on the surface and not by the surface kinetics of Ga or the Si precursor we used. The

different Si/Ga ratio is due to the decomposition kinetics of the TMSi, whose decomposition temperature is estimated to be around 1000°C [41]. In other words, the supply of active Si, i.e. the effective Si/Ga ratio in the gas phase, depends with TMSi on the growth temperature. SiH_4 , a more common precursor, does not present such disadvantage since it is fully decomposed around 700°C , and thus usually no temperature dependence of the roughening has been reported.

Fig. 3 (left) shows the SIMS profiles of Si doped GaN layers grown successively on the same template at four different temperatures without varying the other growth parameters, i.e. keeping the NH_3 , TMGa, TMSi flows constant. Fig. 3 (right) shows the similar SIMS analysis performed on Mg-doped GaN grown at four different temperatures on the previous Si-GaN stack. Unlike TMSi, the EtCp_2Mg decomposition and further Mg incorporation is stable over temperatures ranging

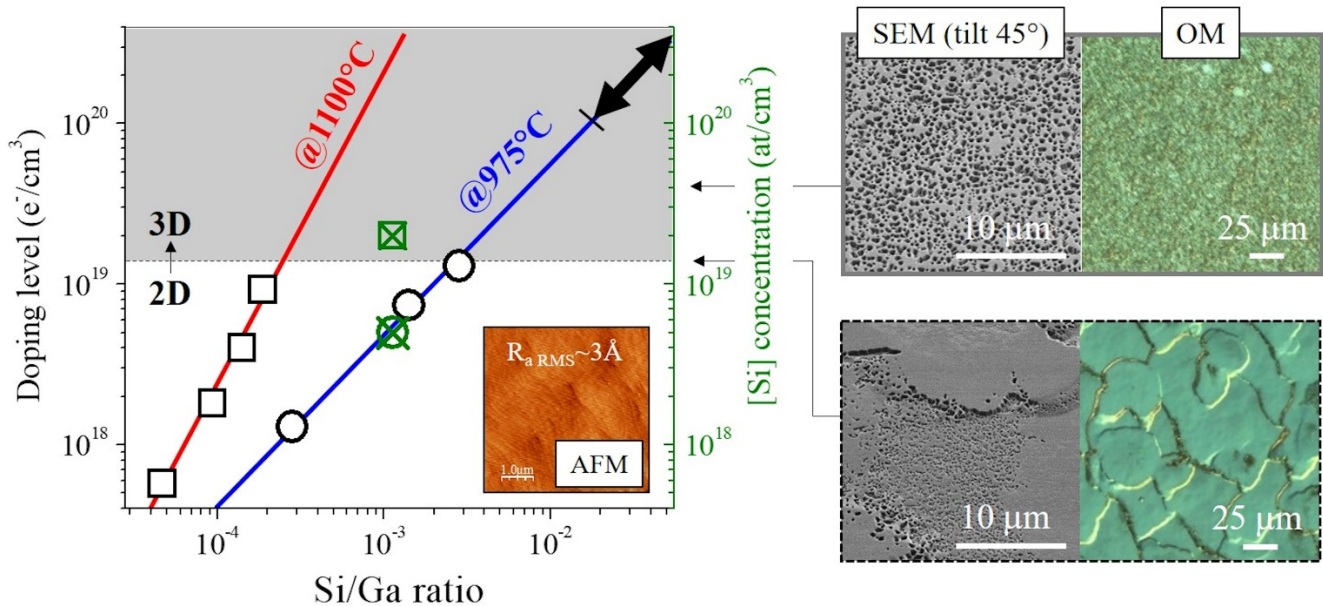


FIG. 2. Doping levels measured by HEM (open black symbols) and dopant concentration (crossed green symbols) measured by SIMS of Si:GaN layers grown at 1100°C (square) and 975°C (circle) for different Si/Ga ratio. The thick left-right arrow indicates the conditions typically required for the growth of TJ. Three kind of surface morphology can be observed depending on the doping concentration: a very smooth surface below 10^{19} cm^{-3} (white area / AFM picture inset), a highly textured surface between $1 \cdot 10^{19}$ and $2 \cdot 10^{19} \text{ cm}^{-3}$ (dashed line / right bottom SEM and OM pictures) and finally a very rough surface above $2 \cdot 10^{19} \text{ cm}^{-3}$ (grey area / right top SEM and OM pictures).

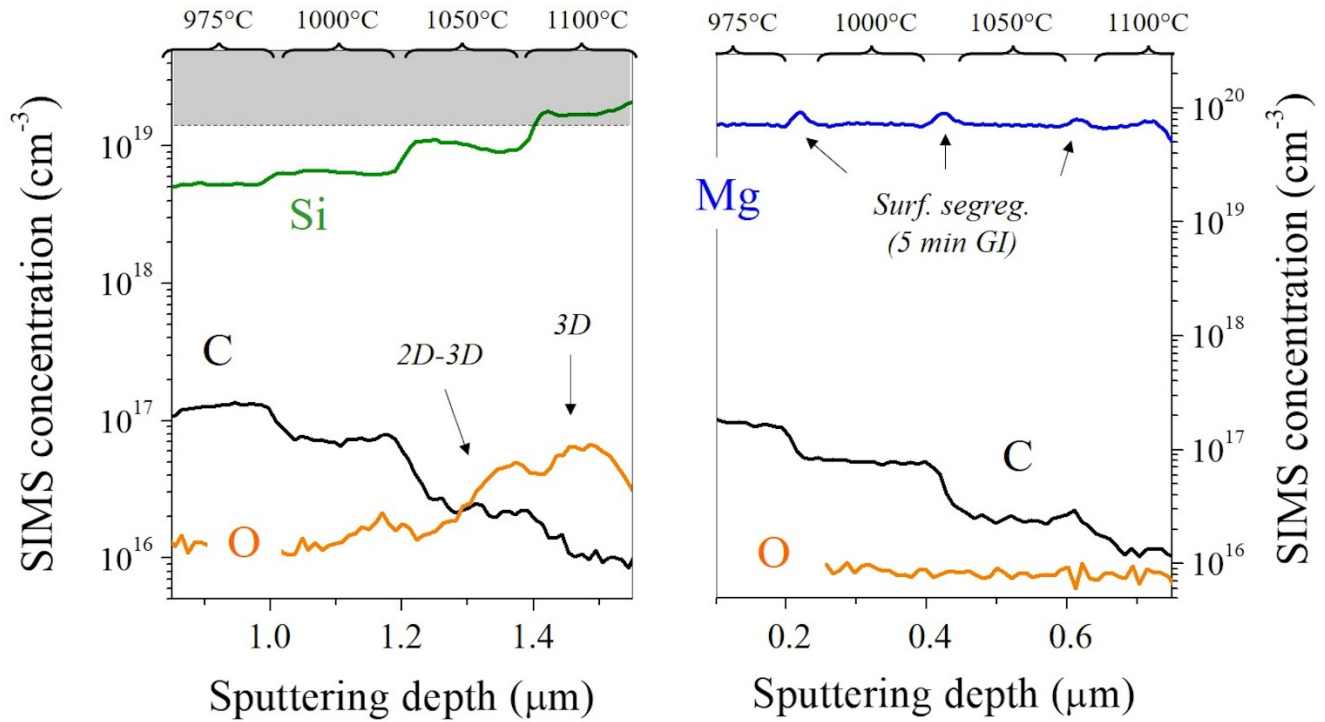


FIG. 3. SIMS analysis of n-GaN (left) and p-GaN (right) layers grown on the same template but at different temperatures. The O signal increases of one order of magnitude when the Si level is higher than $10^{19}/\text{cm}^3$, i.e. when the growth mode changes from 2D to 3D. The Mg incorporation is constant over the full temperature range between 975 and 1100°C but Mg tends to accumulate on the surface during each growth interruption.

from 975 to 1100°C. Mg piled up at the surfaces during the 5 min growth interruptions (GI) used to ramp up and stabilize the temperature between each step, which indicates we doped below the absolute solubility limit. Both for Si:GaN and Mg:GaN, the carbon level increases from 10^{16} to 10^{17} cm⁻³ when the growth temperature decreases from 1100 to 975°C. This is commonly attributed to the lower decomposition rate of NH_3 and less active H readily available to react and etch away hydrocarbons species present on the surface and is also observed at similar level in undoped layers. Such levels do not significantly affect the GaN conductivity. Interestingly, the O signal is very

different for both layers. While the background level remains below 10^{16} cm⁻³ throughout the p-GaN growth, the O concentration increases of one order of magnitude when the n-GaN doping level reaches about 10^{19} cm⁻³. Conversely, once the Si level decreases below 10^{19} cm⁻³, the O concentration slowly returns to the background level. Since 3D growth starts above 10^{19} cm⁻³ as mentioned previously, and 3D growth generates additional semi- and non-polar facets which are usually prone to higher O incorporation, the O signal correlates with the 2D-3D growth transition [46].

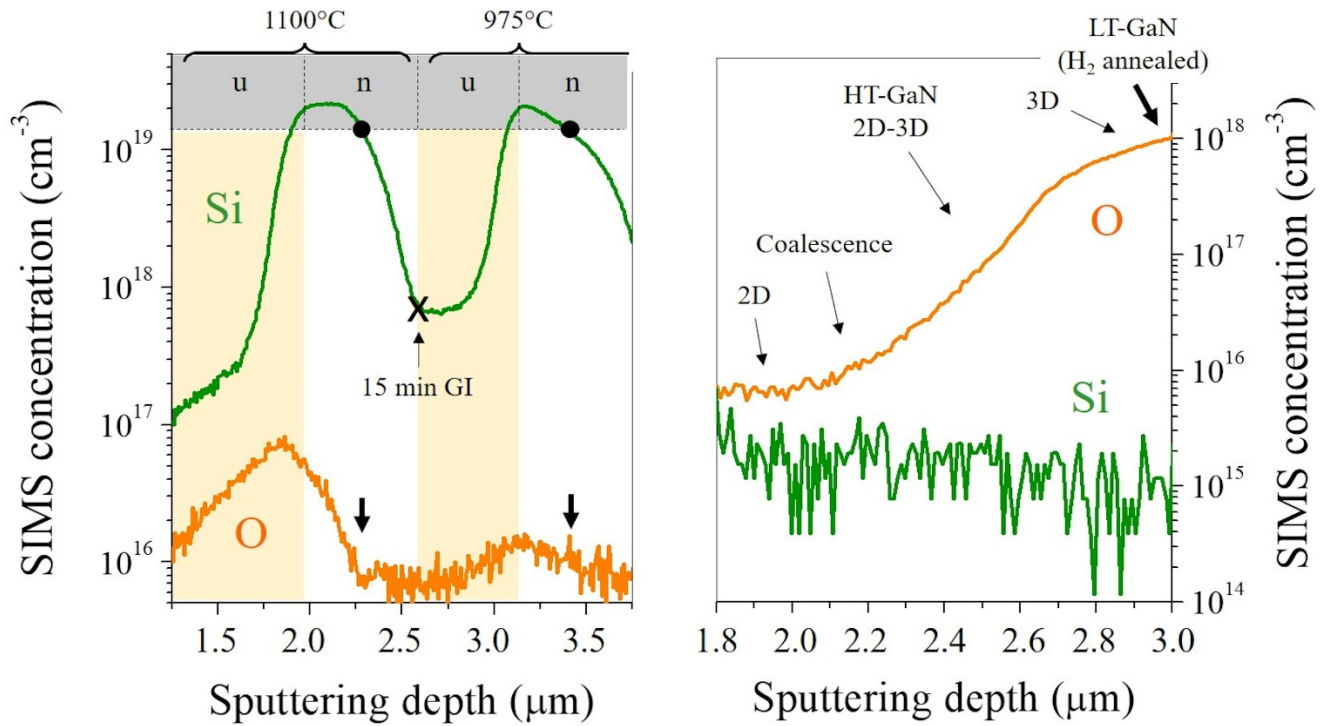


FIG. 4. SIMS analysis of different n-GaN (left) and u-GaN (right) layers. Each time, an increase of the O level is observed when the 3D growth mode is triggered (thick arrows and dots), either by a Si concentration higher than 10^{19} cm^{-3} or by a H_2 annealing of a LT buffer layer. Further GaN overgrowth of the rough surface with proper conditions leads to coalescence and the O level simultaneously decreases to below 10^{16} cm^{-3} in both cases.

Fig. 4 presents two additional evidences for correlation of the O level with 3D growth, triggered with and without SiN_x nanomasking. For the first sample, 600 nm n-GaN is grown at 975°C with a linear increase of the Si concentration up to several 10^{19} cm^{-3} , followed by a 600 nm u-GaN to recover a smooth surface. After a 15 min growth interruption, where no Si surface segregation is observed, the same sequence is repeated at 1100°C . It is worth noting that the O concentration increases as soon as the 3D growth is triggered by the SiN_x nanomasking (black dots and arrows on Fig. 4 (left)). Conversely, the O level slowly decreases

upon u-GaN overgrowth which again smooths the surface. The second sample shows a similar behavior (Fig. 4 (right)) during a u-GaN template growth on sapphire. Here, intentionally, 3D islands are obtained by annealing a nucleation layer deposited at low temperature (LT). At this point, the O level is around 10^{18} cm^{-3} . Further regrowth of the 3D nuclei at high temperature (HT) smooths the surface. Since this means that the semipolar facets are vanishing, the O concentration also decreases until it reaches its background level of about 10^{16} cm^{-3} for O incorporation on a flat (0001) surface.

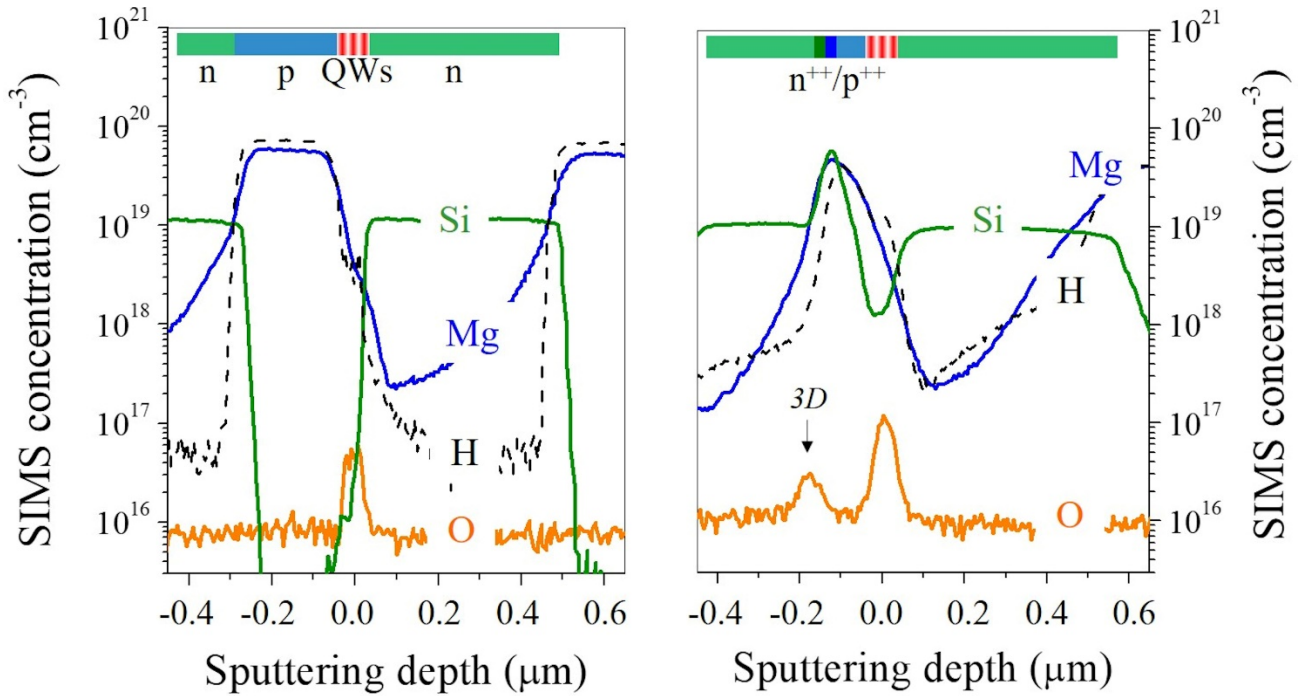


FIG. 5. SIMS analysis of the np-n (left) and TJ (right) based LEDs. In the p-type layers, the Mg is still passivated, i.e. $[Mg] \sim [H]$, since the structures have not been annealed at this stage. Note the O peak and the Mg tail buried in the n^{++} side of the TJ. The x-axis origin has been set at the QWs position for better comparison.

Fig. 1, 2, 3 and 4 indicate a rather low doping limit on both Si:GaN and Mg:GaN, which would be theoretically unsuitable for a TJ. However, driven by many successful reports found in the literature, we compared the SIMS profile of a “np-n” LED (no n^{++}/p^{++} layers at the p-n junction) and one with n^{++}/p^{++} layers “TJ based” LED. Fig. 5 (left) shows the results of the “np-n” LED structure featuring the highest doping level for smooth surfaces as determined from Fig. 1 and 2 ($\sim 10^{19}$ and few 10^{19} cm^{-3} for Si and Mg respectively). The O level exhibits a peak around the QWs position, most likely due to the residual O found in the N_2 carrier gas required for the InGaN growth (n- and p-GaN are grown under pure H_2 carrier gas, which is easier to purify and present lower contamination levels). Since Mg has a strong carry-over into subsequent layers, it is also incorporated in the Si:GaN top layer [47-49]. Even though Mg decreases drastically, the first ten nanometers of intended n-type layer is rather p-type since there is

still a higher Mg than Si concentration. In conclusion, the electrical interface, i.e. the dopants profiles of the maximum doped np top junction are not as well defined as intended. The right side of fig. 5 shows the same structure but with a 15 nm/15 nm n^{++}/p^{++} layers typical for TJ inserted between the n- and p-GaN top layers. The Si and Mg levels were aimed to be around 2×10^{20} and $1 \times 10^{20} \text{ cm}^{-3}$ respectively. The SIMS analysis in Fig. 5 (right) shows that an additional O peak appears in the n^{++} region, indicating a rough interface due to 3D growth. Unfortunately, the SIMS profiles lack depth resolution, i.e. the concentration of the different elements is averaged over a larger volume, depending on how flat the sputter crater is at this position. This underestimates the concentration in these layers. Still, when reaching appropriate Si:GaN growth conditions, the rough surface can be smoothed by a proper overgrowth. Thus, as expected from Fig. 2, the 15 nm thin n^{++} layer has a lot of nanomasking, is very rough, and likely

discontinuous. It is worth noting the electrical interface and the metallurgical are not located at the same position. Due to the strong Mg carry-overs, the electrical interface is buried in the n-side of the junction, around 10 nm away from the rough metallurgical interface. In short, both interfaces are not the abrupt interfaces typically assumed for TJs.

At this point, one has to remember that it is not straightforward to activate buried p-GaN layer due to the impossible vertical diffusion of H through the n-type top layer (note the LED structures presented in Fig. 5 were not activated at this stage as it can be seen by the Mg/H ratio close to 1). Therefore, we remind that the TJ LEDs had to be annealed 30 min at 750°C in N₂ for further electrical investigations. This has been done right after the mesas etching, allowing the H to out-diffuse laterally through the sidewalls. Such Mg activation is mostly limited by the size of the mesas since the diffusion length of the H seems proportional to the square root of the annealing time [33]. In our case, we considered the most favorable situation for the next discussion, i.e. we assumed the p-type layer of our L-shape TJ LED fully activated since the expected H diffusion length (~ 100 μm) is larger than half the typical dimension of our mesas.

Fig. 6 (left) compares the IV characteristics of three TJ based LEDs with different p⁺⁺ and n⁺⁺ doping levels. The reference is a standard cyan LED emitting at a wavelength of 490 nm (red curve). In good agreement with the results found in the literature, the TJ based LEDs show typical LED behavior and emit light once both Mg and Si expected levels reach about 10²⁰ cm⁻³ [31]. A further increase of either dopant concentration leads to a significantly decrease of the penalty voltage. Si seems more efficient than Mg at lowering the junction barrier, as previously observed by Kaga *et al.* [32]. Eventually, for Mg and Si levels of 10²⁰ and 2×10²⁰ cm⁻³ respectively, the TJ based LED penalty voltage is reduced to as

low as 1.1 V with a specific differential resistance of about 10⁻² Ω.cm² at 20 mA. These values are surprisingly low considering the present structure does not involve any (p-)InGaN interlayer that is often introduced between the p⁺⁺ and n⁺⁺ layers to reduce the nominal tunnel barrier height and facilitate the tunneling process [29,50]. It is worth noting at these doping values, the TJ starts to leak at negative bias (Fig 6. dashed circle). A further increase of the Mg and Si levels led to an electrical shortcut of the junction, i.e. the I-V analysis become characteristic of a pure resistor.

Based on the rather poor morphological, crystalline, and electric properties of the overdoped doped n⁺⁺- and p⁺⁺-GaN layers in TJs as shown in the previous sections, the results of Fig. 6 (left) seriously challenge a pure tunneling mechanism for the following reasons:

- i) Assuming no Mg-H passivation (100% H free), which is the most favorable case, the maximum free carrier density achievable in the p⁺⁺ and n⁺⁺ side of the junction is limited around 3×10¹⁹ and 2×10¹⁹ cm⁻³ respectively. Therefore, the space charge region (SCR) spreads over a distance of 20 nm which is not compatible with a Fowler-Nordheim emission, given the large bandgap of GaN. If one further considers that the buried Mg:GaN layers might not be fully activated, the SCR is even larger and the tunneling process even less likely.
- ii) The Mg carry-over causes fairly high amount of active, i.e. not passivated, Mg (>10¹⁹ cm⁻³) to be incorporated in the n⁺⁺ side. This leads to a significant compensation of the free electron density, especially at the vicinity of the interface, which widens the SCR and so further reduces the electric field across the junction.
- iii) Due to the effective doping asymmetry p⁺⁺>n⁺⁺ deduced from i) and ii), the SCR is shifted inside the n⁺⁺ layer. This part of the

junction is characterized by a low structural quality and a high O contamination at grain boundaries inherited from a SiN_x induced 3D growth. Worse, the presence of Mg likely induces additional defects such as IDs, or Mg-O and Mg-Si complexes. Thus, the field of the SCR is strongly affected by (various) deep centers that prevent the free carriers to directly tunnel from one band to the other one, and could even pin the Fermi level at their midgap states.

- iv) Any attempt to increase the Mg and Si concentrations up to 10^{20} cm^{-3} exacerbates the problems discussed in points ii) and iii) even though the penalty voltage decreases and the TJ starts to leak.

These considerations imply that a defect level-assisted tunneling, i.e. Poole-Frenkel emission, is the main mechanism responsible for the TJs in GaN [51]. The band diagram of such TJ and the proposed tunneling mechanisms are sketched in Fig. 6 (right). For this band alignment, the most favorable situation has been considered in the p⁺⁺ type layer: $p = N_a - N_d \approx \text{Mg}$ (i.e. we assumed all the Mg atoms are in substitution of Ga, fully activated by the electric field and free of H passivation). For the n⁺⁺ type layer, we took a donor concentration of

$2 \cdot 10^{19} \text{ cm}^{-3}$ and we included a Mg carry over decreasing from $3 \cdot 10^{19} \text{ cm}^{-3}$ at the metallurgical interface down to $5 \cdot 10^{18} \text{ cm}^{-3}$ at 50 nm (as observed from Fig. 5 and references [47-49]). Again, this corresponds to an ideal case, the most favorable for a TJ “free of defects”. In a more realistic case, one would have to consider the whole 10 nm p⁺⁺ layer mostly self-compensated (quasi-insulating) and a larger SCR buried in the n-side characterized by various crystal defects such as Mg clusters, IDs, SiN_x nanomasking, O complex, etc. Finally, instead of the low probability of direct band to band tunneling on such a long distance, the carriers could tunnel through defect states. So even a less abrupt band transition can still allow relatively efficient tunneling, because the tunnel current is then rather limited by the distance, both in space and energy, between the intermediate states. More states could sustain higher tunnel currents. It is worth pointing out a similar mechanism should occur with AlGaN. Here the problems are even worse since the larger gap would require even higher dopant concentrations and associated issues. In addition, at higher levels, the Si forms DX centers in AlGaN [52]. Hence, overdoping would lead to more midgap states. This explains the recently reported relatively high conductivity of AlGaN TJ in deep UV LEDs [21].

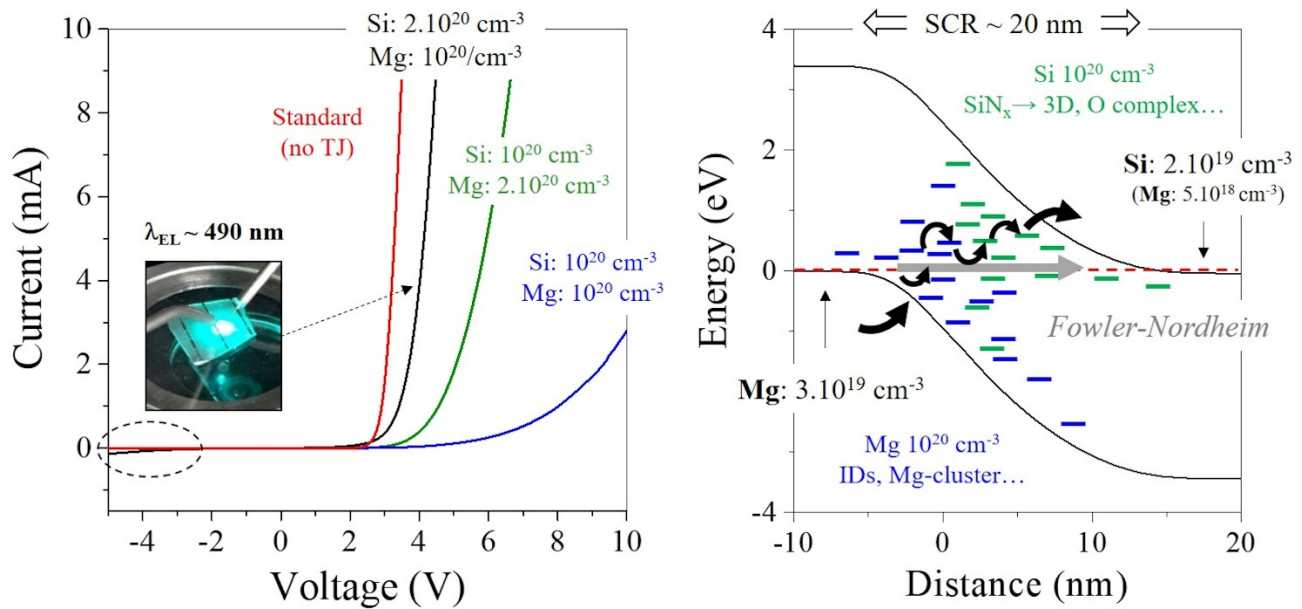


FIG. 6. I-V characteristics of standard and TJ based LEDs with different p^{++} and n^{++} doping levels. The dashed circle highlights the current leakage at negative bias for the most heavily doped TJ (left). Band diagram of unbiased TJ assuming that the p and n doping levels saturate at 3×10^{19} and $2 \times 10^{19} \text{ cm}^{-3}$ respectively, and considering the carry-over of Mg (right). The curved black arrows sketch the defect level-assisted tunneling (Poole-Frenkel emission). The straight grey arrow represents a pure tunneling (Fowler-Nordheim emission). Note the position of the defect levels is arbitrary.

It is worth pointing out our explanation contrasts within the dominant thinking where defects only play a secondary or even negligible role. Therefore, it seems interesting to discuss various results of TJs found in the literature in the light of our observations and results. For instance, Young *et al.* observed the TJ formed by molecular beam epitaxy (MBE) overgrowth of a MOVPE grown p-n LED by atomic probe tomography (APT). They reported a high O concentration at the TJ interface, most likely originating from the residual Ga_xO_y formed during the sample transfer from the MOVPE reactor to the NH_3 -MBE chamber [53]. In contrast to our conclusion that O introduces midgap states in the SCR, the authors concluded the O rather establish a delta doping that strongly reduces the SCR. However, regrowth on a partly Ga_xO_y covered template can easily lead to 3D growth due to different attachment rates on GaN and Ga_xO_y . More recently, Akatsuka *et al.* reported that an overlap of Mg and Si doping at the TJ interface enhances the tunneling current in MOVPE grown

TJs [54]. This is counterintuitive to a sharp TJ, so the authors speculated that either GaN bandgap narrowing due to heavy doping or an advanced “2Mg+1Si” p-GaN co-doping (proposed by Katayama-Yoshida *et al.* [55]) would explain the lower resistivity of their device. However, their results are very reasonable when assuming that midgap states e.g. from defects like Mg or Si or MgSi clusters are needed to enhance tunneling. Indeed, the strong crystal disorder introduced at the TJ interface most likely favors the creation of defect states that can assist the tunneling of the carriers. A similar argument applies to the so-called polarization enhancement TJ structures. Theoretically, the insertion of a thin InGaN interlayer between the p^{++} and n^{++} GaN allows a narrower SCR (due to the local bandgap reduction) and a higher electric field across the junction (due to built-in spontaneous and piezoelectric polarization), which is beneficial for carrier tunneling [56]. Practically, one should also consider that InGaN is grown at low temperature,

and thus has a much higher defect density than GaN. In the case of MOVPE, InGaN must also be grown under N₂ ambient and features a much higher O background (as seen in Fig. 5) while high quality n- and p-GaN require H₂ and higher temperature. Thus, one has to choose between growing the whole TJ at same temperature under N₂ (resulting in poorer crystal quality and higher background O levels) or interrupting the growth to properly switch the carrier gas and increase the temperature (leading to Mg and eventual contaminations piling up at each interface). Furthermore, In-rich InGaN easily introduces 3D growth after few nanometers [57]. Whatever approach is chosen, either overdoping, MBE regrowth or InGaN interlayers, all of these introduce a fairly large amount of disorder inside the TJ, leading to a tunneling current mainly supported by the defects. Following this assumption, one may enhance the tunneling current by deliberately introducing deep levels (e.g. doping with carbon, europium, or iron...) [58].

IV. CONCLUSION

We systematically studied the electrical properties, the surface morphology and the Mg, Si, H, C and O concentrations of p-GaN and n-GaN layers grown by MOVPE at different temperatures for dopant levels ranging from 10¹⁸ to 10²⁰ cm⁻³. SIMS results show the C concentration drastically increases from 10¹⁶ cm⁻³ at 1100°C to about 10¹⁷ cm⁻³ at 975°C without significantly affecting the electrical properties of both p- and n-type layers. However, O is drastically increased once the Si concentration exceeds 10¹⁹ cm⁻³ and the growth mode simultaneously changes from 2D to 3D due to SiN_x nanomasking. Overdoping with Mg to 10²⁰ cm⁻³ causes inversion domains and defect rich layers which show no longer PL. Thus, at the high Si/Ga and Mg/Ga ratios typically used for p⁺-n⁺ tunnel junctions, both type of layers are resistive and present rough surfaces or inversion domains. SIMS analysis of a np-n LED indicates a strong Mg carry-over. The pinning of the Fermi level, close to

the conduction band in the n-GaN, destabilizes the Mg-H complex (Mg/H≪1) and allows the unintentionally incorporated Mg to efficiently compensate the doping level of the first nanometers of the n-GaN top layer. However, introducing more defective n⁺⁺ and p⁺⁺ layers lead to the desired tunneling. Increasing dopant concentration (and thus defects) reduces the penalty voltage. Based on these results, we conclude that tunnel junctions grown by MOVPE present a poor structural quality and mostly work in defects assisted tunneling.

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