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Abstract

Gallium arsenide (GaAs) growth by metal-organic vapor phase epitaxy (MOVPE) is typically achieved by using arsine (AsH_3) and trimethylgallium (TMGa) as precursors. However, this study explores the growth of high-quality, undoped homoepitaxial GaAs thin films using triethylgallium (TEGa) and tertiarybutylarsine (TBAs) as alternative precursors. The thin films were characterised using Nomarski optical microscopy, high-resolution X-ray diffraction (HRXRD), photoluminescence (PL) and temperature-dependent Hall measurements.

The growth temperature significantly influenced the properties of the epitaxial films. A smooth surface morphology was observed in undoped GaAs epilayers grown at temperatures above 600°C , whereas films grown at or below 575°C exhibited a rougher surface morphology. All layers displayed *n*-type conductivity, except those grown at 550°C , which showed p-type conductivity, with a carrier concentration of $6 \times 10^{16} \text{ cm}^{-3}$ and a mobility of approximately $298 \text{ cm}^2/\text{V} \cdot \text{s}$ at room temperature.

The lowest carrier concentration ($\sim 1.5 \times 10^{15} \text{ cm}^{-3}$) and highest mobility ($\sim 5147 \text{ cm}^2/\text{V} \cdot \text{s}$) at room temperature were achieved at growth temperatures of 600°C and 650°C , respectively. Additionally, low-temperature (5.4 K) PL measurements showed a reduction in defect-related emissions with increasing growth temperature, further supporting improved material quality at higher temperatures.

Keywords: Gallium Arsenide, tertiarybutylarsine, triethylgallium, MOVPE

1. Introduction

Gallium arsenide (GaAs) remains a semiconductor of significant interest due to its unique electrical and optical properties. It is structurally compatible with other semiconducting materials, enabling the fabrication of a wide range of technologically relevant heterostructures, including high-electron-mobility transistors and near-infrared detectors and emitters [1,2]. Metal-organic vapour phase epitaxy (MOVPE) is widely used to grow high-quality GaAs epitaxial layers. Typically, GaAs is grown using trimethylgallium (TMGa) and arsine (AsH_3) as precursors. However, arsine (AsH_3) [3], while effective as a Group V arsenic source, is toxic and highly flammable, raising significant safety concerns.

These concerns may be addressed by exploring alternative arsenic precursors such as trimethylarsine (TMAs) [4], triethylarsine (TEAs) [5], and tertiarybutylarsine (TBAs) [6–11]. TBAs (tBuAsH_2) is of particular interest due to its higher thermal decomposition efficiency and significantly lower toxicity compared to AsH_3 . TBAs is more than 50% decomposed at a susceptor temperature above 650°C and a gas temperature ($T_{\text{decomp}}^{\text{TBAs}}$ above 360 ± 10) $^\circ\text{C}$ based on unimolecular decomposition [12], making it more suitable for low-temperature MOVPE growth compared to AsH_3 . Furthermore, TBAs has been shown to result in lower carbon contamination than other arsenic precursors, such as TMAs and TEAs, which is crucial for producing high-quality GaAs films [13]. Consequently, TBAs is considered a highly promising alternative to AsH_3 as an arsenic source in MOVPE [14].

Concerning Ga, triethylgallium (TEGa) offers better control over layer thickness when growing ultrathin structures, such as superlattices and quantum dots, due to its lower vapour pressure than TMGa. This improved thickness control is particularly advantageous for precision applications in advanced semiconductor structures. Moreover, TEGa has been reported to reduce carbon contamination [9] and is more than 50% decomposed at susceptor temperatures above 500°C .

and at gas temperatures ($T_{\text{decomp}}\text{TEGa}$) above 250 ± 10 °C (further enhancing its suitability for high-quality GaAs growth). The optimal substrate temperature for producing high-quality GaAs using TEGa and TBAs has been reported to range from 645°C to 690°C [11,12,15] and 580°C to 640°C [16] , respectively. However, a systematic study examining MOVPE growth parameters for these precursors used together has not been published previously.

In this paper, TBAs and TEGa are utilised as precursors for the MOVPE growth of homoepitaxial GaAs layers. While this precursor combination has been successfully used in fabricating GaAs-based structures, there is limited research on the specific MOVPE growth parameters required to optimise film quality. Typically, deposition parameters such as the V/III ratio and growth temperature play a critical role in the quality of epitaxial layers. Here, the focus is on the effect of growth temperature on the morphological, structural, electrical, and optical properties of GaAs thin films, aiming to provide insight into optimising these parameters to achieve high-quality films.

2. Experimental Procedure

The undoped epitaxial GaAs layers were grown on a semi-insulating (SI) (001) GaAs substrate with orientation (100) 2° off $\langle 110 \rangle$ (± 0.3), supplied by Sumitomo. Growth was conducted at atmospheric pressure in a Thomas Swan Epitor 4 MOVPE system with a horizontal quartz tube at predefined temperatures in 25°C increments from 550°C to 675°C. TEGa and TBAs were used as the gallium and arsenic precursors, respectively. Hydrogen at 2175 sccm served as the carrier gas. The V/III ratio was maintained at 7.5 throughout the growth study to provide sufficient arsenic overpressure, thereby minimising arsenic desorption and promoting stable stoichiometric growth conditions, while allowing the effect of growth temperature to be investigated independently. Prior to growth, 1 cm² substrates were cleaved from a 2-inch wafer and cleaned by blowing with dry nitrogen. No further substrate cleaning was performed after the

dry nitrogen treatment. The substrates were then loaded onto a molybdenum susceptor in the horizontal quartz reactor, and the system was purged with hydrogen for 10 minutes. Subsequently, the substrates were annealed in a TBAs overpressure at 700°C for 10 minutes to remove surface oxides.

Following this, the susceptor temperature was adjusted to the required growth temperature. Depending on the growth rate, determined in a separate experiment, the growth period lasted approximately 4 hours. The layers grown on SI-GaAs were used to assess basic electrical properties via Hall-effect measurements in the Van der Pauw configuration. The surface morphology of the layers was examined via Nomarski optical microscopy. Photoluminescence measurements were conducted using a 532 nm laser line (Nd: YAG) as the excitation source. The luminescence was dispersed by a 1 m-focal length monochromator and detected by a liquid nitrogen-cooled Ge diode. The structural integrity of the layers was evaluated using high-resolution X-ray diffraction (HRXRD).

3. Results and Discussion

3.1 Morphology and crystallinity

Figure 1 presents Nomarski interference-contrast micrographs of undoped epitaxial GaAs layers grown at temperatures ranging from 550 to 675 °C. The surface morphology exhibits a strong dependence on growth temperature. At 550 and 575 °C, Fig. 1(a, b), the layers display relatively rough surfaces characterised by a high density of inverted-pyramid-like depressions, hillocks, and elongated boat-shaped defects. With increasing growth temperature, the surface morphology progressively improves, and smooth, mirror-like surfaces are obtained between 600 °C and 650 °C, Fig. 1(c–e). This temperature dependence is consistent with previous MOVPE studies of GaAs, indicating that the optimum morphology occurs within an intermediate temperature

regime where precursor decomposition and surface kinetics are sufficiently favourable for stable step-flow growth [6,17].

The rough morphology observed at 550–575 °C is associated with kinetically limited growth. At these temperatures, incomplete decomposition of TEGa can result in the accumulation of carbon-containing surface species, which reduce adatom mobility and hinder uniform incorporation at surface steps [1,2,4,14] . Consequently, three-dimensional nucleation is enhanced, leading to the formation of hillocks, pits, and boat-shaped defects. In contrast, between 600 and 650 °C, precursor decomposition is more efficient, and surface diffusion is enhanced, promoting stable step-flow growth and resulting in significantly smoother surfaces [2,5] .

At temperatures of 675°C, the surface morphology deteriorates again Fig. 1(f). This degradation is likely associated with increased thermal desorption of arsenic from the GaAs surface, which promotes Ga-rich growth conditions [8–12,18]. Under such conditions, local Ga accumulation and enhanced surface decomposition can lead to non-uniform growth, resulting in the reappearance of hillocks and pit-like defects. Thus, the observed morphology indicates an optimum growth window of approximately 600–650 °C, where precursor decomposition, surface diffusion, and arsenic stability are favourably balanced.

Although some defects, particularly elongated boat-shaped features and hillocks, may originate from substrate-related imperfections such as dislocations or residual surface damage [13,14,16] Their pronounced temperature dependence indicates that the dominant factor governing the surface morphology is the growth temperature. The observed anisotropic defect shapes are also consistent with the known anisotropic surface evolution of GaAs (001) during epitaxial growth [19] . As is evident from Fig. 1, epitaxial layers with the best morphology were obtained at a growth temperature of 650°C.

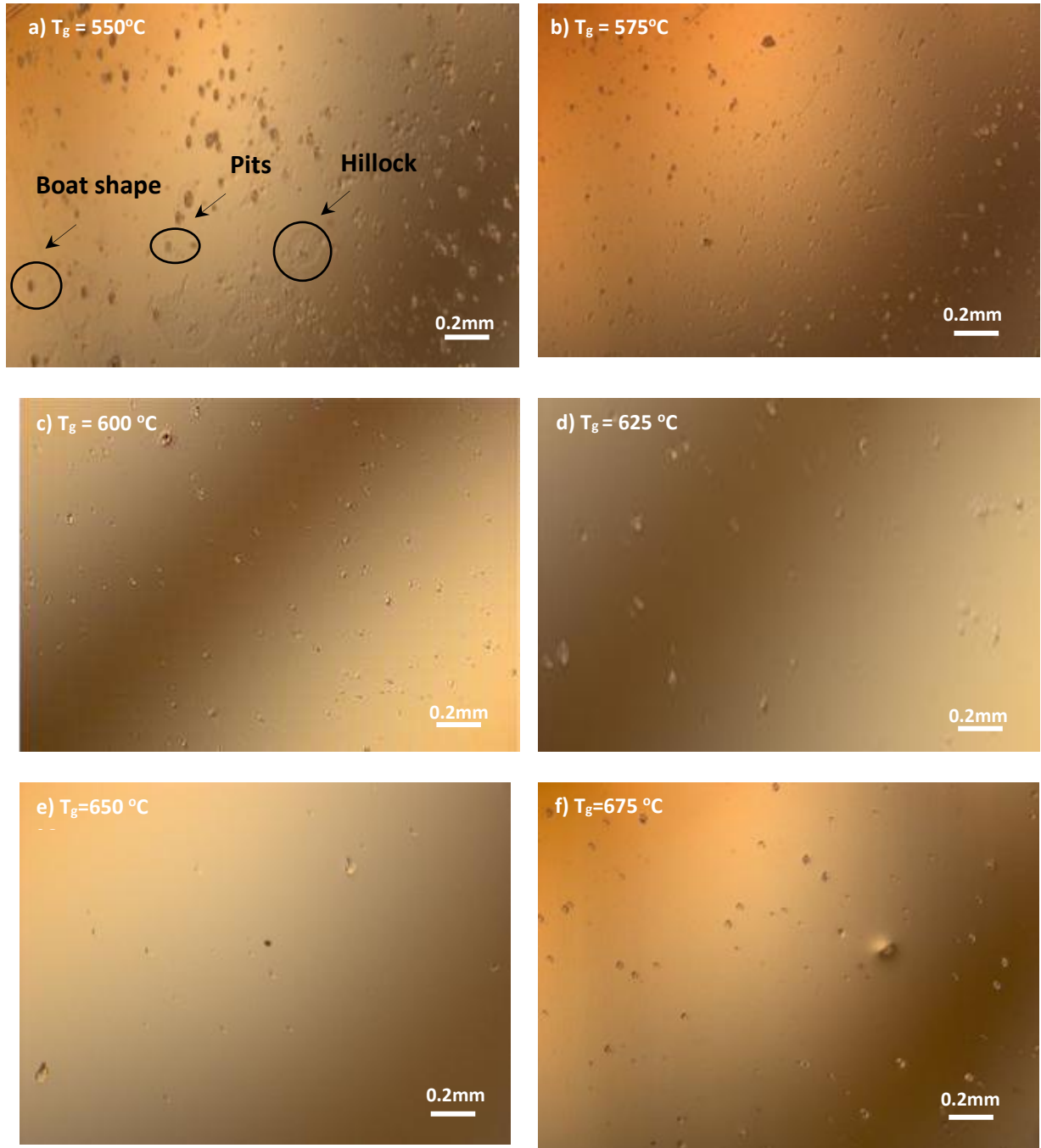


Figure. 1. Nomarski microscopy images of undoped GaAs layers grown at the specified growth temperatures.

High-resolution X-ray diffraction analysis was performed to assess the crystallographic orientation and structural quality of the layers. Fig. 2 (a) presents the normalised HRXRD patterns and (b) the FWHM of the (004) diffraction peak for undoped epitaxial layers grown at the predefined temperatures. The narrow scans, centered on the (004) reflection at approximately 66.05° , confirm that the epitaxial layers are single-crystalline, while the absence of any other reflections, apart from the GaAs (002) peak (not shown), in accordance with similar studies [2,4–6,14], highlights the crystalline quality of the material. Additionally, the small full-width-at-half-maximum (FWHM) values of 0.04° – 0.06° indicate minimal crystalline defects. In this study, the measured FWHM of 0.04 – 0.06° (≈ 144 – $216''$) using TEGa+TBAs lies at the low end of reported values for MOVPE, indicating high crystal quality. In general, narrower FWHM implies fewer mosaic tilts/dislocations [4,20]. This material is therefore expected to exhibit good electronic and optical properties. In summary, the low FWHM values achieved in this study are strong evidence of excellent crystal quality and minimal defects, consistent with consensus in the literature. Notably, the structural integrity of the layers improves with increasing growth temperature, as evidenced by the narrowing of the (004) Bragg peak.[21]

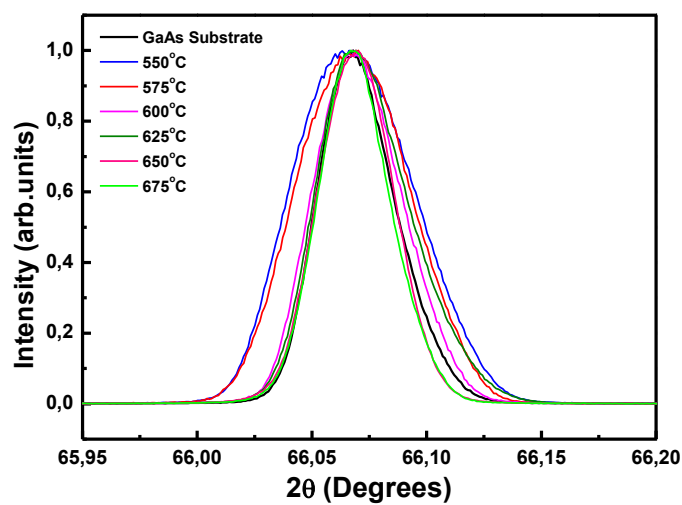


Fig. 2 (a). Normalised HRXRD patterns of the GaAs (004) peak for the SI substrate and the subsequently grown thin films.

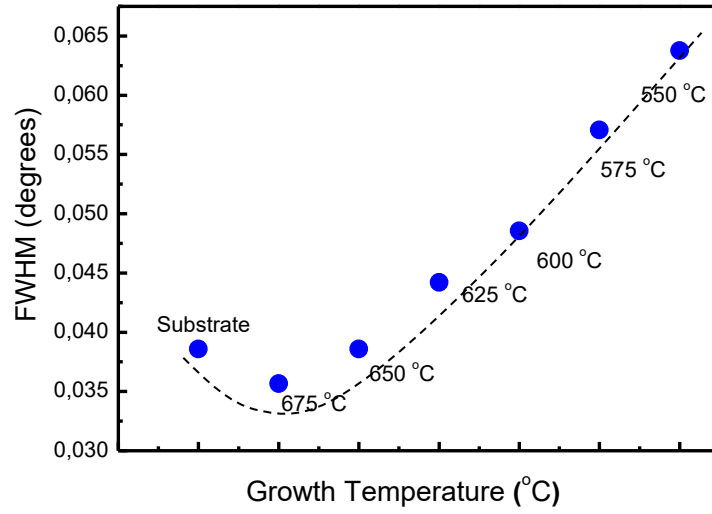


Figure 2 (b). HRXRD FWHM values of the GaAs (004) peak as a function of the growth temperature. The line serves as a guide to the eye.

Hall measurements

Fig. 3 shows the room temperature mobility of undoped GaAs layers as a function of growth temperature. Samples grown at 550°C exhibited *p*-type conductivity, likely due to unintentional carbon incorporation. It should be noted that the metal-organic molecules used in the MOVPE process contain carbon, and incomplete decomposition can lead to carbon doping, resulting in *p*-type behaviour (carbon acts as an acceptor in GaAs, located 26-28 meV above the valence band). However, all layers grown at temperatures above 575°C exhibited *n*-type conductivity.

The lowest electron mobility for *n*-type material was observed in samples grown at 575°C. As expected, electron mobility increases with growth temperature (and improved crystallinity), reaching $\sim 5147 \text{ cm}^2/\text{V}\cdot\text{s}$ at 650°C. This layer also yielded one of the smallest HRXRD FWHM values (Fig. 2) and a smoother surface morphology (Fig. 1).

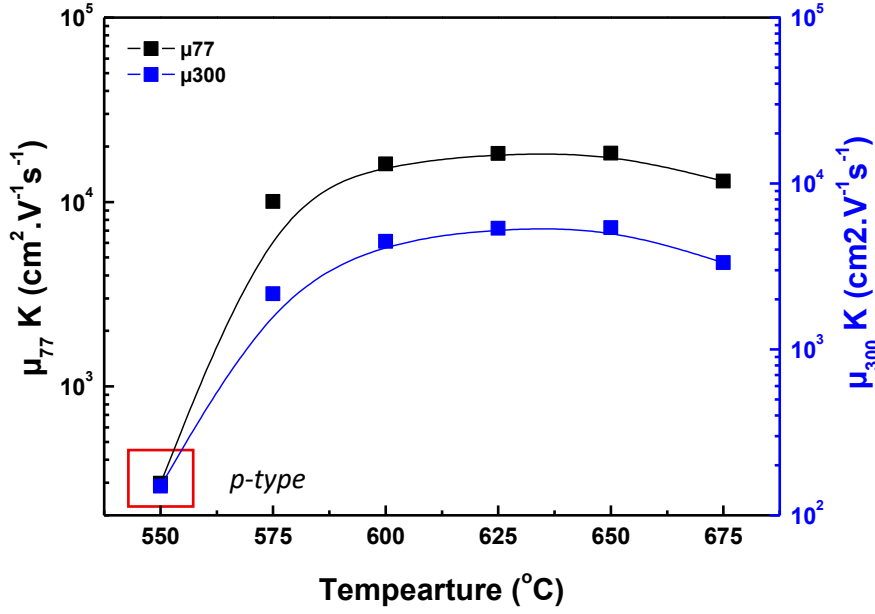


Figure 3. Room temperature Hall mobility of the undoped GaAs epitaxial layers as a function of growth temperature. All layers are *n*-type, except the one grown at 550°C (data points encircled in red).

Table 1. 77 K and 300 K electrical properties of undoped GaAs *n*-type epilayers grown with TEGa and TBAs.

Tg (°C)	n_{77} (cm ⁻³)	μ_{300} (cm ² /V·s)	n_{300} (cm ⁻³)	μ_{77} (cm ² /V·s)	N_D (cm ⁻³)	N_A (cm ⁻³)	θ (N_A / N_D)
575	6,36E+15	2 161	8,33E+15	10053	1,91E+16	1,27E+16	0,66
600	2,52E+15	4 469	4,50E+15	16089	3,65E+15	1,13E+15	0,31
625	2,44E+15	6 365	3,78E+15	18321	3,89E+15	1,45E+15	0,37
650	1,51E+15	8 127	2,62E+15	18364	2,25E+15	7,40E+14	0,33
675	3,06E+15	6 338	5,79E+15	12974	5,66E+15	2,60E+15	0,46

Table 1 summarises the electrical properties (as determined from Hall measurements) of the layers grown at the indicated temperatures. Hall measurements at 300 K and 77 K were used to evaluate the background donor (N_D) and acceptor (N_A) concentrations, and the compensation ratios for *n*-type material. The electrical characteristics show a significant dependence on the growth temperature (T_g). The transition from low-temperature growth at 550 °C to higher temperatures reveals a distinct "purification window" where background impurities and

compensation levels are minimised. The Hall effect measurements demonstrate that as T_g increases from 575 °C to 650 °C, the material purity improves significantly, and the 77 K electron mobility (μ_{77}) increases from a low of 2161 cm²/V. s at 575 K to 18 364 cm²/V. s at 650 °C. The net carrier concentration (n_{77}) reaches a minimum of 2.62×10^{15} cm⁻³ at 650 °C, indicating the lowest background doping level for this growth temperature. The sample grown at 550 °C exhibits exceptionally low mobility (even for *p*-type material) and a high carrier concentration, characteristic of high acceptor concentrations, possibly due to incomplete precursor decomposition at low growth temperatures. In *n*-type GaAs, the relationship between donor (N_D) and acceptor (N_A) concentrations is essential for quantifying material compensation. These are derived using the charge-neutrality condition, combined with ionised-impurity scattering analysis [21]:

$$n = N_D - N_A \quad (1)$$

At 300 K, shallow donors (Si) are fully ionised in GaAs due to their low activation energy (~5–6 meV) [21]

$$n_{300} \approx N_D - N_A \quad (2)$$

At 77 K, donors remain largely ionised in moderately doped GaAs (10^{15} – 10^{16} cm⁻³ range) [21], hence:

$$n_{77} \approx N_D - N_A \quad (3)$$

To extract the total ionised impurity concentration, Matthiessen's rule was applied:

$$\frac{1}{\mu_{77}} = \frac{1}{\mu_L} + \frac{1}{\mu_{ii}} \quad (4)$$

where, μ_L is lattice mobility and μ_{ii} is ionised impurity mobility. For ionised impurity scattering, the Brooks–Herring model [22,23] yields:

$$\mu_{ii} \propto \frac{T^{3/2}}{N_{ii}} \quad (5)$$

Solving simultaneously:

$$N_D = \frac{N_{ii} + n_{300}}{2} \quad (6)$$

$$N_A = \frac{N_{ii} + n_{300}}{2} \quad (7)$$

where N_{ii} = total concentration of ionised scattering centres. The compensation ratio is defined as:

$$\theta = \frac{N_A}{N_D} \quad (8)$$

for undoped GaAs layers that show p -type conductivity, N_A and N_D from single room temperature hall measurement in the simple way, as for an uncompensated semiconductor, for a compensated p -type, charge neutrality is:

$$p + N_D = n - N_A \quad (9)$$

For the p -type, where $p \gg n$, This becomes approximately:

$$N_D - N_A \approx p \quad (10)$$

If acceptor and donor are fully ionised:

$$p \approx N_A - N_D \quad (11)$$

Therefore, Hall data gives the hole concentration p determines the net acceptor concentration, not $N_A - N_D$ and N_A and N_D individually.

The electrical properties of the undoped GaAs layers show a pronounced dependence on growth temperature. The sample grown at 550 °C exhibits an unusually high carrier concentration of ($\sim 10^{17}$) cm^{-3} and a transition to p -type conductivity, indicating substantial impurity

incorporation at the lowest growth temperature. This behaviour can be attributed to the incomplete decomposition of TEGa under kinetically limited growth conditions. At 550 °C, incomplete pyrolysis of the metalorganic precursor may result in the incorporation of carbon-containing species, which can introduce compensating acceptor-related defects and complexes [24,25]. Such impurity incorporation can modify the balance between native donors and acceptors, thereby accounting for the observed conversion conductivity-type.

In contrast, the samples grown between 575 and 650 °C exhibit n-type conductivity, with carrier concentrations of $(2 - 4) \times 10^{15} \text{ cm}^{-3}$ and compensation ratios of approximately 30–35%. These values are within the range reported for unintentionally doped MOVPE-grown GaAs using metalorganic precursors [24–27]. The higher compensation compared with ultra-high-purity commercial GaAs (<5%) [28] is consistent with the residual impurity incorporation typically associated with carbon-containing group-III precursors. In this temperature range, more complete TEGa decomposition and improved surface reaction kinetics promote efficient incorporation of Ga and As while reducing the incorporation of carbon-related impurities [25–27]. The relatively low compensation observed in this temperature window, therefore, suggests that 600–650 °C provides favourable conditions for obtaining high-quality unintentionally doped GaAs.

At 675 °C, the compensation ratio increases to approximately 0.46, accompanied by deterioration in electrical properties compared with the 600–650 °C samples. This behaviour may associate with increased defect formation and/or impurity incorporation at higher growth temperature. Enhanced thermal desorption of arsenic and parasitic gas-phase reactions can modify the surface stoichiometry, thereby promoting the formation of electrically active defects [8,12]. These, result suggest that, the higher temperature can enhance precursors decomposition may adversely affect the electrical properties of epitaxial GaAs.

Overall, the electrical measurements indicate an optimal growth temperature window approximately 600–650 °C, within which n-type conductivity, relatively low compensation, and favourable surface morphology are simultaneously achieved. The results are consistent with previous reports on unintentionally doped MOVPE GaAs and further highlight the important of controlling the growth temperature to regulate impurity incorporation and electrical compensation [24–27].

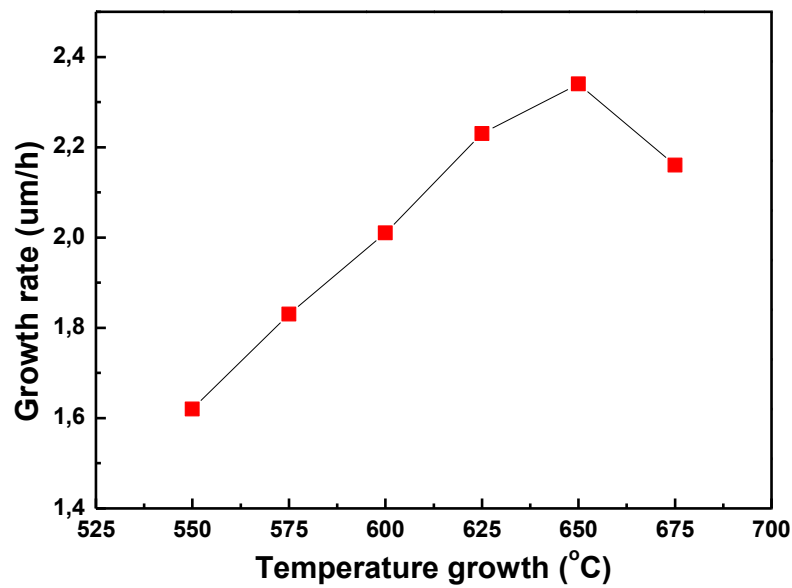


Figure 4. Growth rate based on the growth temperature of undoped GaAs

Figure 4 shows the growth rate as a function of growth temperature for undoped GaAs. The growth rate of undoped GaAs layers grown by MOVPE using TEGa and TBAs increases as the temperature rises from 550 °C to 675 °C. At lower temperatures (~550 °C), incomplete precursor decomposition limits growth and incomplete pyrolysis, but the sticking probability of Ga and As species is high, leading to greater incorporation. As the temperature rises, precursor decomposition improves, yet enhanced surface desorption and increased adatom mobility reduce

the residence time of reactive species. Parasitic gas-phase reactions and boundary-layer depletion further lower the effective flux to the surface.

This trend reflects a shift from a surface-reaction-limited regime at lower temperatures to a desorption and transport-influenced regime at higher temperatures. The observation aligns with prior MOVPE studies of undoped GaAs, where growth rates peak at intermediate temperatures and decline at higher temperatures due to reduced incorporation efficiency [22–24] . These results underscore the importance of precise temperature control when using TEGa and TBAs to achieve a uniform, high quality epilayer.

Photoluminescence

Figure 5 shows the normalised near-band-edge photoluminescence (PL) spectra measured at 5.4 K for the series of layers. The spectra were recorded in the wavelength range 825–840 nm, and photon energies were calculated using $E(eV) = \frac{1239.47}{\lambda (nm)}$. All samples exhibit sharp emission peaks characteristic of high-quality GaAs, however, systematic variations in acceptor-related transitions are observed as a function of growth temperature and directly correlate with the electrical properties extracted from Hall measurements.

In the excitonic region (1.512–1.515 eV), the dominant peaks at 1.515 eV, 1.514 eV, and 1.513–1.512 eV correspond, respectively, to the free exciton (FX), donor-bound exciton (D^0X), and acceptor-bound exciton (A^0X). These assignments are consistent with established low-temperature GaAs spectra reported in the literature [17,19] . The resolution and relative intensity of these excitonic features improve significantly between 550 °C and 650 °C, indicating enhanced crystalline uniformity and changes in the relative incorporation rates of donor and acceptor species.

In the lower-energy region (1.486–1.493 eV), three distinct acceptor-related transitions are observed. The 1.493 eV emission is attributed to the free-to-bound ($e - A^0$) transition associated with carbon acceptors C_{As} , as reported by D. J. Ashen et al. [25] and confirmed in MOVPE-grown GaAs by Leitch et al. [26]. The emission observed near 1.490 eV is attributed to shallow acceptor-related emission. Based on the low temperature GaAs band gap and reported binding energies of common shallow acceptors, this spectral region is consistent with either free to bound recombination, ($e - A^0$), involving shallow acceptor or, depending on the compensation level, donor-acceptor pair recombination (D^0, A^0). Previous PL studies of epitaxial GaAs have established characteristic free-to-bound and donor-acceptor pair transitions for several residual acceptors, including C, Si, and Zn. In particular, Si is amphoteric in GaAs, with Si_{As} acting as an acceptor with a reported binding energy of approximately 34.5 meV, consequently, its free-to-bound transition is expected in the vicinity of 1.485 eV at low temperature. Therefore, the emission observed at 1.486 eV is more appropriately considered consistent with (e, A_{Si}^0) transition rather than being assigned uniquely to Si. The possible contribution of Si cannot be excluded, however, because Si was not intentionally introduced, such an assignment would require independent compositional evidence. The coexistence of several acceptor-related emissions, nevertheless, the presence of more than one electrically active acceptor or acceptor-related defect, which may contribute to carrier compensation in the AP-MOVPE grown GaAs layers. The sample grown at 550 °C exhibits the strongest emission in the 1.486–1.493 eV region, accompanied by broadened excitonic features. Hall measurements show p-type conductivity with $\mu_{300K} \approx 150 \text{ cm}^2/\text{V} \cdot \text{s}$, indicating substantial coexistence of donors and acceptors. The pronounced carbon-related (1.493 eV) and two-hole (1.490 eV) transitions confirm significant incorporation of background carbon under low-temperature growth conditions. Incomplete decomposition of TBAs and TEGa at this temperature likely enhances carbon incorporation, leading to strong ionised-impurity scattering and reduced mobility. The strong correlation

between intense acceptor-related PL and low mobility demonstrates that impurity scattering dominates transport in this regime.

At 575 °C, the relative intensity of the impurity related PL features changes compared with the 575 °C layer. The Si related transition at 1.486 decreases, while the carbon-related 1.493 eV and two-hole 1.490 eV emissions remain evident. At the same time, the near band edge excitonic features become sharper and more pronounced, indicating an improvement in optical quality and reduction in relative contribution of non-radiative recombination.

The electrical measurement shows corresponding change in the transport properties with the layer exhibiting *n*-type conductivity and substantial increase in carrier mobility. The simultaneous reduction donor-acceptor related contribution. The improvement maybe associated with more efficient precursor decomposition at higher growth temperature, which can modify the incorporation of electrical active impurities and native defects. Although carbon related emission remains detectable, the electrical result indicate that net acceptor contribution reduced relative to donor contribution, resulting *n*-type conductivity and improve carrier mobility. ($\mu_{300K} = 2161 \text{ cm}^2/\text{V} \cdot \text{s}$). The PL and hall measurement indicate that increasing growth temperature from 550 °C to 575 °C modifies both the impurity related recombination spectrum and electrical compensation of the epitaxial layers. The sharper excitonic emission and the increased mobility suggest improvement of material quality, while the persistence of carbon related features indicate that impurity incorporation remains relevant at this growth temperature.

For the 600 °C sample, the excitonic peaks at 1.515 eV and 1.514 eV become dominant and well resolved, while the acceptor-related emissions at 1.493 eV, 1.490 eV, and particularly 1.486 eV are markedly suppressed. Hall analysis shows a reduced acceptor concentration and a minimum compensation ratio ($\theta = 0.31$), accompanied by a significant increase in mobility ($\mu_{300K} = 4469 \text{ cm}^2/\text{V} \cdot \text{s}$). This behaviour indicates that carbon incorporation is significantly reduced at

this temperature due to improved surface kinetics and more efficient precursor decomposition. The suppression of DAP-related transitions directly correlates with reduced ionised-impurity scattering.

At 625 °C, as the growth temperature increase from the low regime, the relative contribution of excitonic features becomes significantly more pronounced, whereas the broad impurity related emission is progressively suppressed. This evaluation indicates that elevated temperatures promote more complete precursors decomposition and improve the incorporation of gallium and arsenic into crystal lattice. Across the temperature range (600-650 °C) near band edge excitonic peaks including the bound exciton (D^0X), transition at 1.514 eV and free exciton (FE) structural and optical quality of the GaAs epilayer is successfully maintained at this growth temperature. Acceptor related region for the band – donor to acceptor region highlight systematic evolution of impurity transitions, specifically the 1.493 eV (carbon-related), 1.490 eV (two-hole), and 1.485 eV (silicon related) features at lower temperature related incomplete precursors cracking promotes carbon incorporation, within the optimal window (600-650 °C): Acceptor related transition is effectively suppressed and carrier mobility reaches a maximum. At higher temperatures (>650 °C) a renewed emergence of carbon related emissions and surface degradation is observed, indicating that excessive thermal energy disturbs the impurity balance. However, growth above 650 °C results in renewed carbon-related emission and surface degradation, indicating that excessive temperatures disturb the impurity balance. These findings demonstrate that precise temperature control is essential for minimising impurity incorporation during MOVPE growth.

Electrical transport measurement correlates strongly with this optical observation. While mobility remains high at moderate temperature. Hall data reveal a rising compensation ratio reaching 0.46 at 675 °C and corresponding drop in mobility drop to $3338 \text{ cm}^2/\text{V} \cdot \text{s}$. This non monotonic mobility trend demonstrates the onset of thermally activated defect scattering and

arsenic desorption at elevated temperatures, even though HRXRD indicates good structural quality Fig. 2.

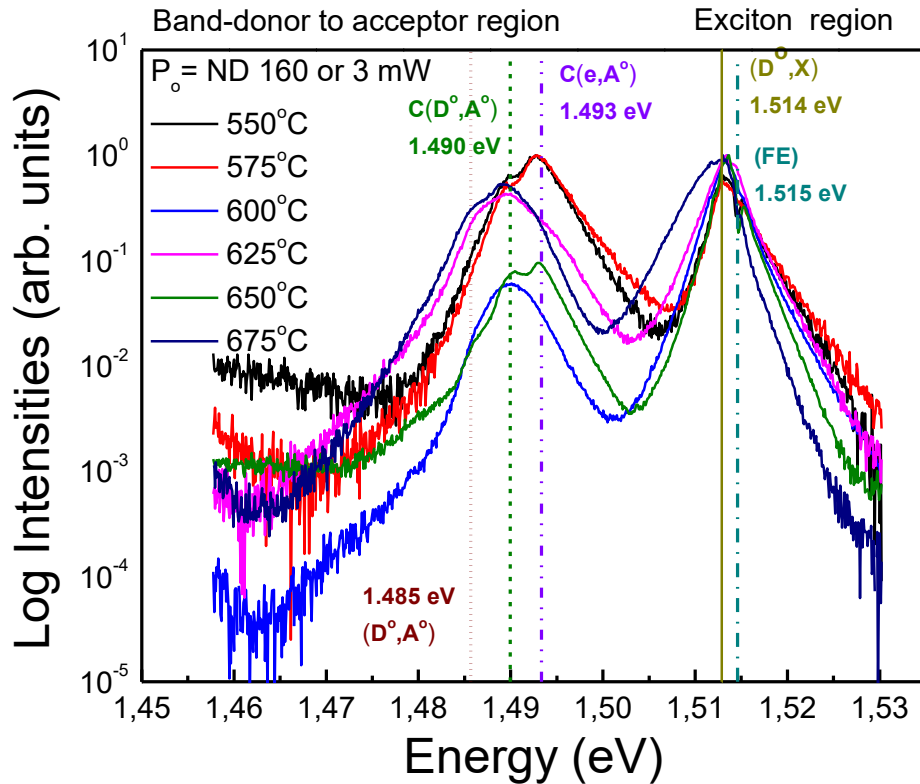


Figure. 5 5.4 K normalised photoluminescence spectra for the exciton and acceptor-related regions of *p*-type (550 °C) and *n*-type layers.

Conclusion

In this study, homoepitaxial GaAs layers were successfully grown using triethylgallium (TEGa) and tertiarybutylarsine (TBAs) by atmospheric-pressure metal-organic vapour-phase epitaxy (AP-MOVPE) at temperatures between 550 and 675 °C. The structural, morphological, optical, and electrical properties were investigated using Nomarski microscopy, high-resolution X-ray diffraction (HRXRD), low-temperature photoluminescence (PL), and Hall-effect measurements. The combined results indicate that the most favourable overall properties were obtained at growth temperatures of 600–650 °C, with a constant V/III ratio, where the layers exhibited

smooth morphology, good crystalline quality, strong near-band-edge emission, and relatively high carrier mobility at low carrier concentration.

These results are consistent with previous studies showing that TBAs is a suitable arsenic precursor for high-quality GaAs growth by MOVPE. Notably, TBAs based MOVPE has been associated with reduced impurity incorporation compared with conventional arsenic precursors, while the growth temperature and V/III ratio remain important parameters controlling the electrical and optical properties of GaAs [27–30]. Studies of intrinsic carbon incorporation in GaAs grown by MOVPE have shown significant dependence on growth temperature and V/III ratio, the lowest carbon incorporation reported near 640 °C under optimised conditions [29].

The p-type conductivity observed in the sample grown at 550 °C is likely attributable to the increased to enhanced incorporation of acceptor impurities at the lower growth temperature. Carbon is a well-established acceptor impurity in GaAs, and previous MOVPE studies have identified carbon-related acceptor transitions near 1.49 eV [30,31]. However, carbon was not directly measured in the present study.

The low temperature PL spectra further confirm excellent optical quality of the layers. The emission around 1.515 eV is consistent with the free-exciton transition of GaAs, while emission at approximately 1.490 eV is indicative of recombination associated with impurities or acceptors [31,32]. The combination of strong near the band edge, high mobility, and low carrier concentration indicates that the layers grown between 600 and 650 °C provide a suitable GaAs reference material.

Although TEGa and TBAs have previously been used for GaAs growth, the present study provides a systematic assessment of their performance under the specific AP-MOVPE conditions employed here. The resulting optimisation establishes a suitable growth window for GaAs and

provides a well-characterised reference platform for the subsequent growth and investigation of GaAs_{1-x}N_x/GaAs structures.

References

- [1] M.T. Crowley, N.A. Naderi, H. Su, F. Grillot, L.F. Lester, GaAs-Based Quantum Dot Lasers, *Semiconductors and Semimetals* 86 (2012) 371–417. <https://doi.org/10.1016/B978-0-12-391066-0.00010-1>.
- [2] C.A. Broderick, S. Jin, I.P. Marko, K. Hild, P. Ludewig, Z.L. Bushell, W. Stolz, J.M. Rorison, E.P. O'reilly, K. Volz, S.J. Sweeney, GaAs_{1-x}Bix /GaNyAs_{1-y} type-II quantum wells: novel strain-balanced heterostructures for GaAs-based near- & mid-infrared photonics, *Sci. Rep.* 7 (2017). <https://doi.org/10.1038/srep46371>.
- [3] W. Metaferia, A.K. Braun, J. Simon, C.E. Packard, A.J. Ptak, K.L. Schulte, Control of surface morphology during the growth of (110)-Oriented GaAs by hydride vapor phase epitaxy, *Cryst. Growth Des.* 21 (2021) 3916–3921. <https://doi.org/10.1021/acs.cgd.1c00235>.
- [4] D.W. Vook, S. Reynolds, J.F. Gibbons, Growth of GaAs by metalorganic chemical vapor deposition using thermally decomposed trimethylarsenic, *Appl. Phys. Lett.* 50 (1987) 1386–1387. <https://doi.org/10.1063/1.97865>.
- [5] D.M. Speckman, J.P. Wendt, Triethylarsenic and arsine as co-reagents: The novel manipulation of in-situ GaAs OMCVD growth chemistry to improve growth efficiency and safety, *Journal of Crystal Growth*, Volume 93, Issues 1–4, 1988, Pages 29-33, ISSN 0022-0248, [https://doi.org/10.1016/0022-0248\(88\)90501-5](https://doi.org/10.1016/0022-0248(88)90501-5).
- [6] C.H. Chen, C.A. Larsen, G.B. Stringfellow, Use of tertiarybutylarsine for GaAs growth, *Appl. Phys. Lett.* 50 (1987) 218–220. <https://doi.org/10.1063/1.97666>.
- [7] R.M. Lum, J.K. Klingert, M.G. Lamont, Use of tertiarybutylarsine in the metalorganic chemical vapor deposition growth of GaAs, *Appl. Phys. Lett.* 50 (1987) 284–286. <https://doi.org/10.1063/1.98226>.
- [8] S.P. Watkins, G. Haacke, H. Burkhard, M.L.W. Thewalt, S. Charbonneau, Magnetophotoluminescence characterisation of residual donors in GaAs grown by metalorganic chemical vapour deposition, *J. Appl. Phys.* 64 (1988) 3205–3209. <https://doi.org/10.1063/1.341537>.
- [9] T.F. Kuech, R. Potemski, Reduction of background doping in metalorganic vapor phase epitaxy of GaAs using triethylgallium at low reactor pressures, *Appl. Phys. Lett.* 47 (1985) 821–823. <https://doi.org/10.1063/1.95995>.
- [10] G. Haacke, S.P. Watkins, H. Burkhard, Metalorganic chemical vapor deposition of high-purity GaAs using tertiarybutylarsine, *Appl. Phys. Lett.* 54 (1989) 2029–2031. <https://doi.org/10.1063/1.101182>.
- [11] G. Haacke, S.P. Watkins, H. Burkhard, Control of residual impurity incorporation in tertiarybutylarsine-grown GaAs, 1991. G. Haacke, S.P. Watkins, H. Burkhard, Control

- of residual impurity incorporation in tertiarybutylarsine-grown GaAs, *Journal of Crystal Growth*, Volume 107, Issues 1–4, 1991, Pages 342–347, ISSN 0022-0248, [https://doi.org/10.1016/0022-0248\(91\)90481-J](https://doi.org/10.1016/0022-0248(91)90481-J).
- [12] Sundaram, V.S., Fraas, L.M. & Samuel, C.C. Group III alkyl source purity effect on the quality of GaAs grown with tertiarybutylarsine. *J. Electron. Mater.* 20, 567–569 (1991). <https://doi.org/10.1007/BF02666019>.
 - [13] Kejihidaka Low growth GaAs using TBAs and TEGa, (n.d.).
 - [14] W. Metaferia, A.K. Braun, J. Simon, C.E. Packard, A.J. Ptak, K.L. Schulte, Control of surface morphology during the growth of (110)-Oriented GaAs by hydride vapor phase epitaxy, *Cryst. Growth Des.* 21 (2021) 3916–3921. <https://doi.org/10.1021/acs.cgd.1c00235>.
 - [15] G. Haacke, S.P. Watkins, H. Burkhard, Epitaxial growth of high-mobility GaAs using tertiarybutylarsine and triethylgallium, *Appl. Phys. Lett.* 56 (1990) 478–480. <https://doi.org/10.1063/1.102771>.
 - [16] Kim, M.H., Bose, S.S., Skromme, B.J. et al. Hall effect analysis of high-purity p-type GaAs grown by metalorganic chemical vapour deposition. *J. Electron. Mater.* 20, 671–679 (1991). <https://doi.org/10.1007/BF02654537>
 - [17] Bogardus, E. H., & Bebb, H. B. (1968). Bound-exciton, free-exciton, band-acceptor, donor-acceptor, and auger recombination in GaAs. *Physical Review*, 176(3), 993. <https://doi.org/10.1103/PhysRev.176.993>.
 - [18] O. Maßmeyer, J. Haust, T. Hepp, R. Günkel, J. Glowatzki, C. Von Hänisch, W. Stolz, K. Volz, Revealing the Significance of Catalytic and Alkyl Exchange Reactions during GaAs and GaP Growth by Metal Organic Vapor Phase Epitaxy, *ACS Omega* 6 (2021) 28229–28241. <https://doi.org/10.1021/acsomega.1c04316>.
 - [19] A.C. Beye, J.C. Garcia, G. Neu, J.P. Contour, J. Massies, Shallow-acceptor isoelectronic-impurity complexes in GaAs, *Solid State Communications*, Volume 67, Issue 12, 1988, Pages 1239–1243, ISSN 0038-1098. [https://doi.org/10.1016/0038-1098\(88\)91091-5](https://doi.org/10.1016/0038-1098(88)91091-5)
 - [20] P. Wanarattikan, S. Sanorpim, S. Denchitharoen, K. Uesugi, S. Kuboya, K. Onabe, Influences of two-step growth and off-angle Ge substrate on crystalline quality of GaAs buffer layers grown by MOVPE, *J. Cryst. Growth* 414 (2015) 15–20. <https://doi.org/10.1016/J.JCRYSGRO.2014.11.027>.
 - [21] P. Modak, M. Kumar Hudai, S. Hardikar, S.B. Krupanidhi, Structural and Electrical Properties of Undoped GaAs Grown by MOCVD, n.d.
 - [22] R. Lang, F. Habib, M. Dauelsberg, F. Dimroth, D. Lackner, MOVPE growth of GaAs with growth rates up to 280 $\mu\text{m/h}$, *J. Cryst. Growth* 537 (2020) 125601. <https://doi.org/10.1016/J.JCRYSGRO.2020.125601>.
 - [23] M.A. Ladugin, I. V. Yarotskaya, T.A. Bagaev, K.Y. Telegin, A.Y. Andreev, I.I. Zasavitskii, A.A. Padalitsa, A.A. Marmalyuk, Advanced AlGaAs/GaAs heterostructures grown by MOVPE, *Crystals (Basel)*. 9 (2019). <https://doi.org/10.3390/cryst9060305>.

- [24] W.G.J.H.M. van Sark, M.H.J.M. de Croon, G.J.H.M. Janssen, L.J. Giling, Analytical models for the growth by metal organic vapour phase epitaxy: III. Applications, *Semicond. Sci. Technol.* 5 (1990) 291–298. <https://doi.org/10.1088/0268-1242/5/4/002>.
- [25] D.J. Ashen, P.J. Dean, D.T.J. Hurle, J.B. Mullin, A.M. White, P.D. Greene, THE Incorporation and Characterisation Of Acceptors In Epitaxial GaAs, N.D.
- [26] H.L. Ehlers, A.W.R. Leitch, J.S. Vermaak, scanning photoluminescence study of the spatial uniformity of the growth rates of OMVPE grown GaAs Quantum wells, *J. Cryst. Growth* 96 (1989) 101–106. [https://doi.org/10.1016/0022-0248\(89\)90279-0](https://doi.org/10.1016/0022-0248(89)90279-0).
- [27] Kobayashi, N. Inoue, Carbon-doped GaAs grown by metalorganic vapor phase epitaxy using TMAs and TEG, *Journal of Crystal Growth*, Volume 102, Issues 1–2, 1990, Pages 183–186, ISSN 0022-0248, [https://doi.org/10.1016/0022-0248\(90\)90900-6](https://doi.org/10.1016/0022-0248(90)90900-6).
- [28] W.S. Hobson, S.J. Pearton, F. Ren, Y. Cheng, D.M. Kozuch, M. Stavola, M. Geva, Carbon doped GaAs and AlGaAs grown by OMVPE: doping properties, oxygen incorporation, and hydrogen passivation, *Materials Science and Engineering: B* 20 (1993) 266–270. [https://doi.org/10.1016/0921-5107\(93\)90239-J](https://doi.org/10.1016/0921-5107(93)90239-J).
- [29] M.C. Hanna, Z.H. Lu, E.G. Oh, E. Mao, A. Majerfeld, Intrinsic carbon incorporation in very high purity MOVPE GaAs, *J. Cryst. Growth* 124 (1992) 443–448. [https://doi.org/10.1016/0022-0248\(92\)90498-8](https://doi.org/10.1016/0022-0248(92)90498-8).
- [30] M. Longo, R. Magnanini, A. Parisini, L. Tarricone, A. Carbognani, C. Bocchi, E. Gombia, controlled intrinsic carbon doping in MOVPE-grown GaAs layers by using TMGa and TBAs, *J. Cryst. Growth* 248 (2003) 119–123. [https://doi.org/10.1016/S0022-0248\(02\)01846-8](https://doi.org/10.1016/S0022-0248(02)01846-8).
- [31] M. Ozeki, K. Nakai, K. Dazai, O. Ryuzan, Photoluminescence Study of Carbon Doped Gallium Arsenide, *Jpn. J. Appl. Phys.* 13 (1974) 1121–1126. <https://doi.org/10.1143/JJAP.13.1121>.
- [32] Weyers M, Sato M, Ando H, Weyers_1992_Jpn. _J. _Appl. _Phys. _31_L853, *Japanese Journal of Applied Physics* 31 (1992) 853–855. <https://iopscience.iop.org/article/10.1143/JJAP.31.L853/pdf> (accessed July 12, 2024).