Nanometer-scale, quantitative composition mappings of InGaN layers from a combination of scanning transmission electron microscopy and energy dispersive x-ray spectroscopy

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Abstract
Using elastic scattering theory we show that a small set of energy dispersive x-ray spectroscopy (EDX) measurements is sufficient to experimentally evaluate the scattering function of electrons in high-angle annular dark field scanning transmission microscopy (HAADF-STEM). We then demonstrate how to use this function to transform qualitative HAADF-STEM images of InGaN layers into precise, quantitative chemical maps of the indium composition. The maps obtained in this way combine the resolution of HAADF-STEM and the chemical precision of EDX. We illustrate the potential of such chemical maps by using them to investigate nanometer-scale fluctuations in the indium composition and their impact on the growth of epitaxial InGaN layers.

(Some figures may appear in colour only in the online journal)

1. Introduction

Indium gallium nitride (InGaN) is used in commercial light-emitting diodes (LEDs), and lasers [1], and is being actively investigated for use in solar cells [2–8]. It is one of many scientifically and economically important materials [9–11] whose quality can potentially be improved if the material were better understood. Detailed chemical mappings of the alloy composition from transmission electron microscopy (TEM) and scanning transmission microscopy (STEM) images can be particularly interesting in such studies, providing useful insights into material growth.

While both TEM and high-angle annular dark field scanning transmission microscopy (HAADF-STEM) give qualitative chemical information, precise, quantitative maps cannot be readily obtained. Indeed, only relative intensity variations in TEM and HAADF-STEM reflect variations in composition. The difficulty lies in the fact that these relative intensity variations cannot be quantitatively interpreted directly from the TEM or STEM image [12]. In order to do
that, one needs a means of establishing an absolute scale of composition.

Recently, Rosenerauer et al [13] implemented a method, originally proposed by LeBeau and Stember [14], to obtain quantitative chemical maps from HAADF-STEM by fitting frozen lattice simulations to experimental images. In this paper we report an alternative for obtaining quantitative chemical maps from HAADF-STEM. In our case the STEM intensity variations are quantified experimentally, using a small set of energy dispersive x-ray spectroscopy (EDX) measurements as a chemical calibration of the observed intensity variations. This approach combines the chemical precision of EDX with the spatial resolution of HAADF-STEM, to obtain quantitative chemical mappings that cannot be obtained independently from either technique. It presents two advantages over direct EDX mappings: for the EDX signal to be reliably quantifiable, exposure times of more 60 s are required, making 1024 × 1024 mappings not only impractical but also subject to error due to small displacements of the sample during the acquisition time; moreover, the resolution of EDX is limited by beam-broadening to, at best, 1–2 nm.

In what follows, we first use elastic scattering theory to show how, within certain limits, the scattering law of electrons in HAADF-STEM can be evaluated experimentally through a small set of EDX measurements. We proceed to describe how to compute this scattering law in practice. Then we demonstrate how to use the resulting function to transform qualitative Z-contrast images of InGaN layers into precise, quantitative maps that reflect the variations of indium composition in these layers with nanometer-scale precision. Using the results from such maps we provide useful insights into the impact of fluctuations in the indium composition on the growth of epitaxial InGaN layers.

2. Experiment

The InGaN epilayers used for this study were all grown on commercial gallium nitride (GaN)/sapphire templates by metal–organic vapor-phase epitaxy (MOVPE), using nitrogen as the carrier gas and tri-methyl indium (TMIn), tri-methyl gallium (TMG) and ammonia (NH3) as precursors for elementary indium, gallium and nitrogen, respectively. The growth temperature was 800°C. Two samples among them are discussed below. Sample A was 70 nm thick, while sample B was 140 nm thick. The TMIn/III, the ratio of TMIn to the sum of TMIn and TMG, for sample A was 15%, while that of sample B was 25%.

X-ray diffraction (XRD) measurements were performed in a Panalytical X’Pert Pro diffractometer in a triple axis configuration. In order to perform accurate composition and strain measurements, reciprocal space mappings (RSMs) were taken along both the symmetric (0 0 0 6) and asymmetric (1 1 2 4) reflections. The effect of tilt was taken into account by averaging the results from measurements along all six asymmetric reflections of the wurtzite lattice, as suggested by Moram and Vickers [15]. The composition was then obtained by numerically solving the third-order polynomial linking the composition to the a and c parameters of wurtzite layers for any strain layers, introduced by Schuster et al [16].

Specimens from the samples were then prepared for STEM using focused ion beam (FIB) thinning and ion milling. The specimens were 80 nm thick. In order to preserve the sample surface during FIB preparation, a surface coating consisting of a 50 nm-thick layer of carbon, followed by 100 nm of silicon nitride (Si3N4) was applied. Prior to STEM imaging, the specimens were cleaned using an argon plasma cleaner.

STEM and EDX were then performed in an aberration-corrected JEOL 2200FS microscope, operating at 200 kV with a probe current of 150 pA, and a probe size of 0.12 nm at the full width at half maximum (FWHM). The convergence half-angle of the probe was 30 mrad and the detection inner and outer half-angles for the HAADF-STEM images were 100 and 170 mrad, respectively. The samples were imaged along the (1 1 2 0) zone axis.

Compared to TEM, which is known to cause image artifacts due to electron beam damage [17], currents of at most a few hundred pA are used in STEM and are expected to leave no residual damage on the specimen. Although the stability of InGaN in HAADF-STEM in particular has been demonstrated previously [13], we ensured that this was the case here also by taking several sets of images and comparing them. No change was observed in the sample between the different sets of measurements.

Quantitative measurements of the indium composition from EDX were obtained from the intensity ratio of the Lα line of indium (3.290 keV) to the Kα line of gallium (9.770 keV). The K line of elementary nitrogen (0.392 keV) was also taken into account and revealed that the alloy is stoichiometric. The acquisition time for each EDX spectrum was 60 s, during which no drift in the position of the electron beam was observed. The EDX spectra were acquired using a JEOL 2300D detector and the accompanying JEOL software. The k-factors used by the software had been previously re-calibrated using GaAs, InP, GaN, AlN, GaP and GaSb, as well as the ternary alloys In0.48Al0.52As and In0.53Ga0.47As. These alloys are lattice-matched to InP, allowing the composition to be precisely determined through XRD.

3. Quantification of the Z-contrast

In this section we explain how the quantification works. We first give a proof of principle from first-principles calculations. We then proceed to present an algorithm to implement it. Finally, we use the algorithm to quantify a HAADF-STEM image of sample A. This example is used to evaluate the spatial resolution and chemical precision of the algorithm.

3.1. Proof of principle

It has been shown that in HAADF-STEM, at detection inner half-angles greater than 60 mrad and for STEM specimens thinner than 100 nm the major contribution to the collected intensity is that of elastically scattered electrons [18]. In
particular it has been shown that a good approximation of the fraction of the incident electron beam intensity scattered towards those angles by an atomic column of the STEM specimen is given by [19]

\[ I_n \propto d \langle Z \rangle^n. \]  

(1)

Here, \( d \) is the length of the imaged atomic column (i.e. the STEM specimen thickness), \( \langle Z \rangle \) is its average atomic number and \( n \) is an exponent that will be discussed in more detail later.

For an atomic column of In\(_x\)Ga\(_{1-x}\)N the average atomic number is given by

\[ \langle Z \rangle_{\text{InGaN}} = xZ_{\text{In}} + (1-x)Z_{\text{Ga}} + Z_{\text{N}}, \]  

(2)

or, equivalently,

\[ \langle Z \rangle_{\text{InGaN}} = xZ_{\text{InN}} + (1-x)Z_{\text{Ga}} + Z_{\text{N}}. \]  

(3)

where \( Z_{\text{InN}} = Z_{\text{In}} + Z_{\text{N}} \).

Thus, one can relate the root \( \alpha \) of \( I_n \) to the concentration \( x \) of an atomic column of In\(_x\)Ga\(_{1-x}\)N by the following equation:

\[ \sqrt[\alpha]{I_n} = \sqrt[\alpha]{Kd} \langle xZ_{\text{InN}} + (1-x)Z_{\text{GaN}} \rangle. \]  

(4)

\( K \) is a proportionality coefficient that depends on the imaging geometry and which is constant during the acquisition of the HAADF-STEM image. If one divides both sides of equation (4) by the reference intensity \( I_{\text{ref}} = Kd \langle Z \rangle^n_{\text{GaN}} \), one finds an intensity ratio \( R \) that is linked to the composition by

\[ \sqrt[\alpha]{R} = (\zeta - 1)x + 1, \]  

(5)

where \( \zeta = \frac{Z_{\text{GaN}}}{Z_{\text{GaN}}} \) is the ratio between the atomic numbers of pure indium nitride (InN) and pure GaN.

From equation (5) it is obvious that if one were to know the value of \( \alpha \), one could readily invert the equation and find

\[ x = \xi(\sqrt[\alpha]{R} - 1), \]  

(6)

where \( \xi = \frac{1}{\zeta - 1} \) is introduced to simplify the notation. Then one could proceed to compute a chemical composition map by applying equation (6) to the HAADF-STEM image.

In elastic scattering theory, the exponent \( \alpha \) is equal to 2. However, it has been argued that a more appropriate modeling of Rutherford scattering would need to take into account electron screening. In such cases, \( \alpha \) is given by

\[ \alpha = 2 - \sigma, \]  

(7)

where \( \sigma \geq 0 \) is a factor modeling electron screening. This factor depends both on the scattering atom and the penetration depth of the incident electrons into the atom’s electron cloud, which in turn depends on the beam acceleration voltage. As a result, the value of \( \alpha \) depends on the material being imaged as well as the imaging conditions. Typical values of \( \alpha \) range between 1.7 and 2 [19].

Although one cannot have an \textit{a priori} knowledge of the value of \( \alpha \), one could estimate it using a set of EDX measurements of the composition, \( \{ x_i \}_{i=1}^M \), measured at the same locations as a set of intensity ratios \( \{ R_i \}_{i=1}^M \), and then using the following estimator \( \hat{\alpha} \):

\[ \hat{\alpha} = \left( \frac{\log R}{\log ((\zeta - 1)x + 1)} \right)_{i=1}^M. \]  

(8)

In section 3.2 we will discuss an algorithm that computes the estimator \( \hat{\alpha} \) and then applies equation (6) to obtain a concentration map.

3.2. Implementation

Here we will discuss how to practically implement the method outlined in section 3.1. The implementation proposed here was carried out in MATLAB, where the HAADF-STEM image is represented by a \( N \times N \) matrix, of typical size \( N = 1024 \). For clarity, we will first introduce a few conventions on notation before giving the details of the implementation.

3.2.1. Notation. As discussed above, each pixel of the STEM image corresponds to an intensity \( I \) collected at position \( (k, l) \) of the STEM probe. In what follows, the intensity \( I_{k,l} \), collected at position \( (k, l) \), will be referred to as \( I \), whereas \( I \) will refer to the matrix of intensities \( \{ I_{k,l} \} \). Furthermore, an \( N \times N \) matrix of ones will be referred to by \( I_N \).

Point-wise, or element-to-element, matrix division will be denoted by ‘/’, i.e. \( C = A / B \) will signify

\[ C_{k,l} = A_{k,l} / B_{k,l} \quad \text{for all couples} \quad (k, l) \]

such that \( k = 1, \ldots, N, \quad l = 1, \ldots, N. \)  

(9)

Finally, a subset of \( M \) intensities \( I_k \), or a set of \( M \) EDX measurements \( x_i \) will be denoted by \( \{ I_i \}_{i=1}^M \) and \( \{ x_i \}_{i=1}^M \), respectively.

3.2.2. Algorithm. A schematic diagram of the proposed algorithm is given in figure 1. In HAADF-STEM, the intensity is collected by an annular scintillation detector combined with a photomultiplier type (PMT), resulting in the useful signal being added on top of a background intensity from the polarization current of the detector. Thus, the first step consists of measuring the average background signal \( \langle b \rangle \) and subtracting it from the collected image intensity matrix \( I_{\text{raw}} \).

We then obtain a normalized intensity image \( I_n \) that can be used for the quantification

\[ I_n = I_{\text{raw}} - \langle b \rangle \cdot I_N. \]  

(10)

In the second step, we evaluate the reference intensity \( I_{\text{ref}} \) defined previously. The reference is taken in the GaN buffer. Since specimen preparation by FIB results in a wedge shaped sample, shown in figure 2, \( d \) is a function of the position \( (k, l) \) in the image. In practice, one observes an intensity gradient, all the more evident at low magnifications. To account for this, \( I_{\text{ref}} \) is obtained through a linear least-squares fit of the GaN part of \( I_n \), to a first-order, two-dimensional polynomial function. This function is then extrapolated to the whole image, and in particular to InGaN. Since InGaN is not too dissimilar from GaN and the gradient created from FIB is not too pronounced, we expect any residual thickness variations to be significantly smaller than the effect of the wedge.

A point-wise division between the background-free \( I_n \) and the reference intensity \( I_{\text{ref}} \) results in the matrix of intensity ratios \( R \):

\[ R = I_n / I_{\text{ref}}. \]  

(11)
Then, a set $\{R_i, x_i\}_{i=1}^N$ of intensity ratios and the associated composition, independently evaluated by EDX measurements, is used to compute the estimator $\hat{a}$ given by equation (8). Since the EDX probe interacts with a volume of the specimen larger than the STEM probe, it would be inaccurate to use the intensity of the center of the probe as the value $R_i$ used in the estimation. Instead, an average ratio was computed for each position of the probe, obtained by using a moving average filter on the image. The mask for the moving average filter was obtained by modeling the EDX probe interaction volume by a truncated cone. The diameter was computed as the average between the entry and exit disks. This typically results in a $5 \times 5$ matrix for the mask.
Finally, the function given in equation (6), combined with the resulting estimator $\hat{\alpha}$, is applied to the matrix of intensity ratios $R$ to compute the quantitative composition map.

3.3. Example

In this section we show an example of how the algorithm works on sample A. The spatial resolution and the chemical precision are estimated and limiting factors for both are discussed. Figure 3 shows an HAADF-STEM image of sample A. One clearly distinguishes the GaN template (bottom right), the GaN layer and the carbon–Si$_3$N$_4$ cap layer. A close inspection of the GaN layer reveals a slight intensity gradient, roughly perpendicular to the GaN/InGaN interface and towards the bottom right of the image. The gradient becomes evident if one extracts the intensity profile from the top left corner of the image to the bottom right corner, shown in the inset of figure 3. As discussed in section 3.2.2, this gradient can be attributed to the gradient in the sample thickness. The dashed blue line represents the section of the extrapolated $I_{ref}$ along the profile line. It shows that a first-order polynomial is a reasonable approximation of $I_{ref}$ in the area of interest, i.e. the GaN and InGaN layers.

Figure 4 shows an image of the intensity ratio $R$, obtained in stage three of the algorithm after subtraction of the background intensity $I_B$ and division by the reference intensity $I_{ref}$. The inset shows profiles of the intensity ratio $R$ along the dashed red and green lines, respectively. Fits in black are given as a guide to the eye. One can see similar oscillations in the intensity along the two lines, proof of a residual lateral modulation of thickness, associated with the FIB preparation.

In order to evaluate $\hat{\alpha}$, a set of 18 EDX measurements was performed. The distribution of the EDX measurements across the HAADF-STEM image is shown in figure 4 (red and blue dotted lines). As indicated in section 3.2.2, the corresponding set of intensity ratios $\{R_i \}_{i=1}^M$ was computed by using a moving average filter with a 5 pixel $\times$ 5 pixel mask. This is the required mask size for a 2 nm EDX disk at the magnification used in for the HAADF-STEM image presented in figure 3. As discussed in section 3.2.2, this gradient can be attributed to the gradient in the sample thickness. The dashed blue line represents the section of the extrapolated $I_{ref}$ along the profile line. It shows that a first-order polynomial is a reasonable approximation of $I_{ref}$ in the area of interest, i.e. the GaN and InGaN layers.

Equation (8) was used on the set $\{R_i, x_i \}_{i=1}^M$, shown in table 1. A value of $\alpha = 2 \pm 0.1$ was found to be the appropriate power model in this case. Similar values for the exponent $\alpha$ in the case of InGaN alloys were reported by Amari et al [20], using the exact ‘sum-of-squares’ scattering model:

$$R = \frac{xZ_{In}^\alpha + (1-x)Z_{Ga}^\alpha + Z_{N}^\alpha}{Z_{Ga}^\alpha + Z_{N}^\alpha}. \quad (12)$$

To check the validity of our approach, we also evaluated the fit of our values to this model. A good fit could not be obtained unless we let $\alpha = 1.7$. This calculation also revealed that, for indium compositions between 0 and 80%, the difference between our model and the sum-of-squares is negligible. Further
Table 1. Summary of the EDX measurements and their corresponding intensity ratios, used for the map presented in figure 5. Numbers in italic were excluded from the calculation, since they were located in regions where FIB artifacts were suspected to be present.

| Line 1 | Line 2 |
|--------|--------|
|        | EDX    | HAADF | EDX    | HAADF |
|        | $x_i$  | $R_i^x$ | $x_i$  | $R_i^x$ |
| 11.5   | 1.115  | 11.4  | 1.068  |
| 12.9   | 1.124  | 13.2  | 1.121  |
| 12.6   | 1.140  | 11.7  | 1.138  |
| 11.7   | 1.143  | 11.9  | 1.168  |
| 13.8   | 1.145  | 12.4  | 1.168  |
| 14.0   | 1.110  | 12.7  | 1.191  |
| 13.2   | 1.145  | 13.0  | 1.179  |
| 14.5   | 1.152  | 12.7  | 1.168  |
| 12.3   | 1.109  | 13.4  | 1.196  |
| Average| 12.9   | 1.131 | 12.5   | 1.155 |
| Standard deviation | 1.0    | 0.017 | 0.7    | 0.040 |

investigation using Ga-poor InGaN alloys may help decide which model is more appropriate. For the purpose of the algorithm presented in this paper either model can be used.

Figure 5 shows the concentration map obtained after application of equation (6). The colorbar indicates the computed indium composition. It has been extended below zero, to better show that the estimation error is randomly distributed around 0 in the GaN layer. The root mean square error (RMSE) between the measured and computed values correspond to a fully relaxed 20% InGaN layer. Similar results have been observed for thick InGaN layers presenting double diffraction peaks in XRD RSMs. Sample B is an example of such a case, where two InGaN diffraction peaks besides the intense GaN peak are apparent in both the asymmetric and the symmetric RSM (cf figure 7). The first diffraction spot corresponds to pseudomorphically strained 13% InGaN. This diffraction spot elongates towards its fully relaxed position, indicating an elastic relaxation of strain. The second diffraction spot corresponds to a fully relaxed 20% InGaN layer. Similar results have been observed for thick InGaN layers in [22–29].

A HAADF-STEM image along the (1 1 2 0) zone axis of sample B is shown in figure 8. It is similar to what has been reported in [26–28]. The GaN, InGaN and C/Si3N4 cap was reported by Rosenauer et al in [13]. Further improving the sample preparation may contribute to removing these artifacts and increasing the precision of the map. For the purpose of the application presented in section 4, however, this value was sufficient.

Finally, the results obtained by the algorithm were compared to the concentration deduced from XRD RSMs. Figure 6 shows an average, tilt-free RSM for the asymmetric (1 1 2 4) (left) and the symmetric (0 0 0 6) (right) reflection, respectively. Table 2 compares the indium concentration obtained from measurements presented in section 4, however, this value was sufficient.

4. Application to InGaN layers presenting double XRD diffraction peaks

In this section, the algorithm is applied to the study of thick InGaN layers presenting double diffraction peaks in XRD RSMs. Sample B is an example of such a case, where two InGaN diffraction peaks besides the intense GaN peak are apparent in both the asymmetric and the symmetric RSM (cf figure 7). The first diffraction spot corresponds to pseudomorphically strained 13% InGaN. This diffraction spot elongates towards its fully relaxed position, indicating an elastic relaxation of strain. The second diffraction spot corresponds to a fully relaxed 20% InGaN layer. Similar results have been observed for thick InGaN layers in [22–29].
Table 2. Average indium composition and the associated standard deviation for both samples, for three different techniques.

|                  | XRD            | EDX            | EQHAADF       |
|------------------|----------------|----------------|---------------|
|                  | Average        | Standard        | Average        | Standard        | Average        | Standard        |
|                  | composition    | deviation       | composition    | deviation       | composition    | deviation       |
|                  | %              | %              | %              | %              | %              | %              |
| Sample A         | 12             | 1              | 12.7           | 0.9            | 14             | 1.4            |
| Sample B         | InGaN 1        | 13.5           | 1              | 13             | 14             | 1.4            |
|                  | InGaN 2        | 22             | 5              | 23             | 23             | 5              |

![Figure 6](image6.png)

Figure 6. XRD reciprocal space mappings of sample A along the asymmetric $\langle 1 \overline{1} 2 4 \rangle$ (left) and the symmetric $\langle 0 0 0 6 \rangle$ reflection (right), respectively.

![Figure 7](image7.png)

Figure 7. XRD reciprocal space mappings of sample B along the asymmetric $\langle 1 \overline{1} 2 4 \rangle$ (left) and the symmetric $\langle 0 0 0 6 \rangle$ reflection (right), respectively.

layers have been labeled. The image shows that the cap layer preserved the rough surface morphology of the InGaN layer and several pyramids are clearly visible all the way to the top. The presence of these pyramids indicates a three-dimensional growth mode, in agreement with results reported in [24–29].

If one inspects the InGaN layer closely, one can see the presence of two distinct regions: a region of homogeneous contrast near the GaN/InGaN interface, labeled InGaN 1, and a second region, where the contrast fluctuates laterally, concentrated at the top of the InGaN layer, labeled InGaN 2. There are two types of fluctuations in InGaN 2: short-scale variations, of the order of a few nanometers, which give InGaN 2 a ‘veined’ aspect, and long-scale fluctuations, of the order of a few tens of nanometers, concentrated essentially near the interface of the two InGaN sublayers. Figure 9 shows the corresponding chemical composition map. The boundary between InGaN 1 and InGaN 2 is clearly identified in the composition map. The average concentration of InGaN 1 computed by the map is 14%, while that of InGaN 2 is 23%. Thus, InGaN 1 and InGaN 2 can be unambiguously identified as the origin of the fully strained and fully relaxed diffraction spots in the RSM, respectively. It should be noted here that, when computing the average composition of InGaN 2, the
The map also reveals that both short-scale and long-scale lateral fluctuations in the HAADF intensity in InGaN 2 correspond to variations in the indium concentration of the order ±5%. This is significantly higher than the uncertainty due to FIB artifacts and also corresponds to the spread in composition determined using XRD. Due to the rapid variations in the HAADF intensity, it has been argued that the short-scale fluctuations in intensity may be linked to the presence of threading dislocations [27]. In order to check whether the algorithm does not misinterpret these fluctuations, atomic scale resolution images of the pyramids, where the threading dislocations are expected to emerge, were taken to verify if this is the case. An example of the results is given in figure 10(a). It shows an atomic scale resolution HAADF-STEM image of the tip outlined by a white dashed rectangle in figure 8. This particular pyramid was chosen because it is entirely within the FIB section and can be observed all the way to the tip. Moreover, it presents very strong short-scale fluctuations that can be easily misinterpreted. While a line of lower contrast is shown to run through the middle of the pyramid all the way to the top, no disruption of the atomic planes is observed. This is confirmed by the corresponding bright field (BF)-STEM image, shown in figure 10(b). Additional EDX measurements along the dotted red line shown in figure 10(a) (see inset) revealed a local decrease from 22% at the edges of the pyramid to 14% near the middle. We can thus conclude that the map accurately reflects fluctuations of the indium composition.

As discussed in [29], we believe the observed process, i.e. the elastic relaxation of InGaN 1, the transition to 3D growth, the increase in the average indium concentration in InGaN 2 and the lateral fluctuations of indium concentration...
in InGaN 2, to be the result of an accumulation of excess indium at the surface during early stages of growth. This in turn leads to a lateral modulation of the incorporation rate of indium into the lattice, which results in a lateral modulation of the growth rate. This eventually leads to the elastic relaxation of the initial InGaN, the appearance of additional facets favorable to growth and the transition to 3D growth. The strong faceting and eventual 3D growth are evidenced by the roughness of InGaN 2. No particular family of planes is favored in the faceting, as revealed by the presence of pyramids with different basal angles and heights. All pyramids, however, present lateral fluctuations of the composition to some extent. Finally, the stark difference between the average compositions of InGaN 1 and InGaN 2, as well as the significant lateral variations of indium composition in InGaN 2, can be explained by either, or both, of the two following mechanisms:

(i) The absence of compositional pulling [30, 31] in the elastically relaxed InGaN 2 sublayer.

(ii) A more favorable binding of indium atoms along the prismatic (1 0 1 ̅) planes [32, 33] that make up the facets of the pyramids.

5. Conclusion

In this paper we have shown that is possible to experimentally evaluate the scattering law of electrons in HAADF-STEM using EDX. In particular, we have evaluated it for InGaN epitaxial layers and used the resulting function to compute quantitative chemical maps of the indium composition. The resulting maps are shown to have nanometric resolution, are precise to 1.5% of the composition and are consistent with results from XRD. Based on these mappings, we have shown that fluctuations in the indium composition are responsible for the spontaneous formation of sublayers with distinct compositions and strain states during the growth of InGaN. Finally, several possible explanations for the origin of these fluctuations are discussed.

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