Strain, composition and disorder in ADF imaging of semiconductors

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Abstract: The effect of strain, composition and disorder in ADF images is systematically studied as a function of detection angle in order to understand the main contrast mechanisms. We demonstrate that the complex phenomenology in ADF images can be accounted for by accurate simulations and modelling. The advantage of an accurate modelling on the image interpretation will be demonstrated in the case of dislocations, chemical analysis of InGaP/GaAs and most noticeably the measurement of both In and N content in quaternary InGaAsN.

1. Introduction

Whereas the initial hopes on scanning transmission electron microscopy (STEM) with a high-angle annular dark-field (HAADF) detector were for an easily interpretable chemically sensitive technique it is becoming now clear that many effects determine the contrast with this technique, especially when imaging along high symmetry zone axes [1]. We mention in particular here strain and static disorder effects [2–4]. On one hand these contrast effects may complicate compositional analysis but on the other hand new interesting information can be gained from more detailed experiments. The most recent microscopy studies have demonstrated a high interest in STEM based techniques. It is becoming clear that new and more reliable information can be gained with different kinds of detector settings beyond the “classical” HAADF setting. This is the case of LAADF (low angle annular dark field) [5,6] and even ABF (annular bright field) that pushes the interest of analysis in the region of low angle scattering [7,8].

We aim to demonstrate in this work that new insight into materials properties can be gained by the comparison of imaging with different camera lengths. We will show that it is possible to investigate the presence of surface strain relaxation, gauge the static displacement, evaluate quantitatively the composition in quaternary alloys and perform defect analysis. A qualifying point of a successful analysis of the different contrast mechanisms is the comparison of imaging with different experimental conditions that permit to distinguish the different contrast contributions.
We will concentrate in particular on the analysis of the angular distribution of scattering as a function of different structural properties of the material, namely composition, static disorder and long range strain. In order to understand the possible phenomena and contrast mechanisms we used both approximate modelling and rigorous multislice simulation with frozen lattice algorithms [9,10]. The latter has been implemented using an updated version of the STEM_CELL parallel computing algorithm [11,12].

The first part of this work will be dedicated to developing a kinematical model of diffuse scattering and study how it relates to chemical effects, static displacement and strain. In the second part a few experimental examples of applications are shown. These examples will regard dislocations in GaN, the effect of alloy disorder on the contrast between InGaP and GaAs and the composition in InGaAsN quaternary alloy. These examples have been chosen to highlight the role of each single contrast mechanism.

2. Theory

Diffuse scattering arises from the displacement of atoms from the periodic ideal position. This may occur because of

1) thermal agitation,
2) the local atomic static displacement (SD) to reach equilibrium configuration. This typically occurs on a very local scale as a form of disorder in alloys or because of point defects.
3) long range strains field are imposed on the analyzed structure.

Since diffuse scattering is dominant in HAADF and very important in medium and low angle dark field it is relevant to pinpoint the mechanism of diffuse scattering production.

A first common characteristic of diffuse scattering, especially at medium-high angle, is the incoherent character, i.e. the diffuse scattering produced by an atomic column is to a good level of approximation the sum in amplitude without phasing of the contributions from all atoms. These waves can then undergo additional elastic scattering that are at the origin of, for example, Kikuchi lines or additional diffuse scattering.

If this diffuse multiple scattering is neglected and the elastic rearrangement is azimuthally averaged in the integration over the detector it is possible to introduce a kinematical approximation typically used for high angle scattering that consider the HAADF image as the convolution of a probe and an object function [13]. We refer in particular to the notation of Voyles [14] that considers a depth (hereafter $z$) dependent probe $J(z)$ which can be identified with the forward propagating part of the electron wave that has undergone only elastic scattering

$$I_{HAADF} = \int J(z) \otimes O(\vec{r},z)dz$$

$\vec{r}$ here is the coordinate vector in the plane $(x,y)$ orthogonal to electron propagation and $O$ is then the object function that contains the scattering from single atoms (at position $\vec{r}_i$) and that can be written to some extent as

$$O(\vec{r}) = \sum_{\text{atoms}} \delta(\vec{r} - \vec{r}_i)\sigma$$

For the cross section $\sigma$ of the diffusely scattering single atom a common expression that is valid for scattering angles $\theta$ much larger than the typically low order Bragg angles is [15]

$$\sigma \propto \int_{\theta_{\text{min}}}^{\theta_{\text{max}}} \frac{f^2(\theta)}{1 - e^{-2M(\theta^2/\lambda^2)}} d^2\theta$$

This expression has been introduced to describe the thermal diffuse scattering in the Einstein model. It contains the single atom scattering factor $f$ that produces at high angles the well known $Z^2$ dependence at high angles. However it is customary to extend this to the case of static displacement, although not being perfectly applicable. An effective Debye Waller factor is therefore introduced [4] as

$$M = M_{\text{TDS}} + M_{\text{Stat}}$$
accounts for the average static displacement. This simple quasi-thermal modelling permits to give a qualitative account of the characteristics of static displacement, namely that it produces an increase of low angle scattering. This is graphically represented in figure 1a showing the average cross-section according to eq. 3 for pure GaAs and for GaAs after the introduction of a SD of the same magnitude as what would be produced by 3% of N. This small amount of N has a relatively low chemical effect (here neglected) but produces a large disorder due to the strong difference in covalent/ionic radius between As and N [16]. A similar effect was already found on B-doped Si [17].

![Image of diagrams](attachment:image.png)

**Figure 1.** a) Comparison of the cross-section for GaAs and GaAsN as a function of the scattering angle for a beam voltage of 200kV: the difference is more pronounced at low angle. b) Angular dependence of the Huang scattering (see text). The solid line refers to rigorous simulations for the GaAs$_{1-y}$N$_{y}$ (y=3%)GaAs case. The dotted line is a quasi-thermal modelling. The dashed line refers to the effect of surface strain (structure as in ref. [2]) and has been obtained by comparison between strained and unstrained GaAs. All profiles have been smoothed by three point lateral averaging to reduce the fluctuation inherent to the limited numbers of frozen lattice configurations.

According to our valence force field (VFF) simulations the additional non-thermal displacement $\nu_{SD}$ for 3% N in GaAsN is 0.0087nm for group III element and 0.0051 nm for group V elements. The most important finding is that the difference between the two cross-sections concentrates in the low angle regime. Following a nomenclature typical of X-rays we will call this additional static diffuse contribution “Huang scattering” [18] to distinguish it from the thermal contribution. Another characteristic that emerges from this model is that diffuse scattering tends to saturate for very large disorder since for very large $M$ (which would make the crystal similar to an amorphous alloy) the cross-section would tend to a constant.

In Figure 1b the angular distribution of Huang scattering has been plotted according to rigorous simulations (solid line) and according to the quasi-thermal model (dotted line). The details of the rigorous model are described in the experimental and simulation section below. The distribution in the quasi-thermal model has been translated laterally by a constant value to fit experimental data. The best fit is obtained for a translation by 6 mrad and accounts for the convergence of the impinging beam (11.9 mrad). Both solid and dotted profiles are obtained as the difference between the GaAsN and GaAs case. In spite of some differences the two profiles are quite similar. The additional dashed line refers to the case of long-range strain that will be considered below. All profiles are normalized to their maxima.

Previous work [4] has demonstrated that a correction of the cross-section as above is not sufficient to account for experimental phenomenology as, in fact, SD produces important effects on the forward propagating wave. In particular in zone axis conditions the most relevant part of $J$ in equation (1) is the 1s Bloch wave component of $J$. The 1s state is responsible for the localization of most of the probe current on the atomic column (channelling) and for this reason its excitation is more directly bound to the generation of diffuse scattering.

Simulations have demonstrated that the introduction of SD produces a faster damping of the 1s excitation and therefore the $J$ factor in eq. 1 for the diffuse scattering [3]. This phenomenon goes under the name of dechannelling and since it affects only the term $J$ in eq. 1 it produces a similar effect at both low and high angle. However it is interesting to note that at low angle SD has two competitive effects: dechannelling reduces the intensity (through the decrease of $J$) while Huang scattering increases it (through the increase of $\sigma$). This produces a non monotonic dependence of the LAAFD intensity on the size of SD [4]. Conversely, SD through dechannelling produces always a decrease of intensity in HAADF (even at very high detection angles). A way to summarize the above effects is to remark that SD translates diffuse scattering from high angle to low angle.
If the above discussion permits to define a framework for the thermal and static local effects this does not allow to treat the case of long-range strain. The use of an effective Debye Waller factor with average displacement in one direction $v_i$ equal to a component or the module of elastic displacement vector $\mathbf{u}$, for example, is not satisfactory since $\mathbf{u}$ depends on which part of the crystal is used as a reference and can be different from 0 even at distances far from an extended defect.

It has been highlighted that long-range strain and local static displacement have a different effect on the propagation of the forward wave function $J$ [19]. We will try to highlight also some characteristics of the additional Huang scattering produced by the strain.

In fact, the dashed curve in figure 1b is the plot of the angular distribution of the Huang scattering produced by the surface relaxation at the barriers of a strained QW (as in ref. [2]). As in the precedent cases the Huang scattering has been deduced as the difference between the strained and unstrained case for a specimen thickness of 40nm. This deduction is based on the fact that the other effect, namely dechannelling, is only a multiplicative factor in the whole angular distribution. This does not mean that strain related dechannelling is negligible: Walther and Humphreys [20] for example have found strain effects extending to detection range out of 200 mrad where Huang scattering is certainly negligible. Conversely, what we want to highlight here is that the simulation shows that the angular distribution of strain related Huang scattering is quite different, namely more oriented in the forward direction with respect to the case of SD. This can be an effect of the more localized nature of the SD.

A qualitative but effective description of the spatial location of this Huang scattering contribution comes from a recent simplified model that describes the forward propagating wave $J$ in terms of its sole 1s component [21]. This approach permits to calculate a qualitative image able to reproduce the main experimental features and to clarify that the important quantities for ADF contrast are the surface tilt of the specimen and the curvature of the atomic planes, i.e. a variation of the column inclination. When such a variation occurs the states at the two parts of the crystal need to be matched. With increasing curvature the overlap between the two states (and in particular the 1s state) is reduced but an additional diffuse scattering is observed. This has been summarized in the equation system in ref. [2] and is ere extended to the 2D case with the addition of the Huang scattering contribution $I_{\text{Huang}}$ which is just proportional to the curvature $J$ through an adjustable factor $H$.

$$\Gamma^2 = \left( \frac{\partial^2 u_x}{\partial z^2} \right)^2 + \left( \frac{\partial^2 u_y}{\partial z^2} \right)^2$$

$$\frac{d^2 \Phi_{1s}(\rho, z)}{dz^2} = \left[ -2\mu_1 \frac{d\Phi_{1s}(\rho, z)}{dz} - 2 \left( \frac{\Gamma^2}{\sigma_\theta^2} \right) \Phi_{1s}(\rho, z) \right]$$

$$\frac{dI_{\text{Huang}}(\rho, z)}{dz} = 2H \left( \frac{\Gamma^2}{\sigma_\theta^2} \right) \Phi_{1s}(\rho, z)$$

$$I = \int (\Phi_{1s} + C) \sum_{i=\text{atoms}} \sigma_i \delta(z - z_i) dz + I_{\text{Huang}}$$

Here $\Phi_{1s}$ is the 1s excitation factor, $\mu_{1s}$ is the 1s absorption factor, $\sigma_i$ is the size of 1s in reciprocal space, $C$ is an adjustable parameter to account for the non-1s states contribution. The non trivial initial condition for the 1s state is

$$\Phi_{1s}(z = 0) = \frac{1}{\Omega} \left[ 1 - \frac{1}{\sigma_\theta^2} \left( \frac{\partial \mu_{1s}}{\partial z} \right)_{z=0} - \frac{1}{\sigma_\theta^2} \left( \frac{\partial u_x}{\partial z} \right)_{z=0} \right]$$

where $\Omega$ is the area of the unit cell and the specimen tilt $\theta_t$ has been set for simplicity along $x$ direction.
3. Experimental and simulations

Three kinds of sample have been analyzed for this work. The first is nominally undoped hexagonal GaN grown at 1150°C on a GaN buffer grown at 560°C on (0001)Al2O3 substrates in an AIXTRON MOVPE system equipped with a rotating substrate [22]. This sample has been prepared in plan-view to observe dislocations end-on. Conventional two-beam analysis has been also conducted. As second sample we used an In0.5Ga0.5P/GaAs (x=51%) structure grown by MOCVD at 700 °C. The InGaP was capped with other layers obtaining the following layer sequence along [100]: GaAs substrate/GaAs buffer /InGaP/GaAs QW/AlGaAs cap. As third sample we used a 4 period multiple quantum well of InGaAsN/GaAs. The nominal compositions are [x]=20% In and [y]=2.4 % N. The structure has been grown by MOCVD on GaAs. The last two samples have been prepared in cross section along the <110> direction. For (S)TEM experiments the InGaAsN has been prepared in cross-section by FIB. As for GaN and InGaP samples the specimen preparation has been performed by mechanical polishing and ion milling using a GATAN DUO MILL.

The STEM observation for these two samples has been performed using a JEOL 2200 FS operated at 200keV and with an objective lens spherical aberration Cs=0.5 mm implying an optimal convergence of 11.9 mrad. In the case of InGaP the specimen thickness has been measured by energy-loss thickness mapping and convergent beam imaging. The experiments on InGaAsN samples have been performed on a Titan 80/300 TEM/STEM operated at 300 keV with an objective lens spherical aberration Cs=1.2 mm implying an optimal convergence of 9 mrad. The detector response in this case has been calibrated for different camera lengths according to the well established procedure. This has permitted to obtain a direct evaluation of sample thickness along the zone axis in the GaAs region [23].

As for simulations, rigorous calculations have been performed based on STEM_CELL software, a parallel computer simulation which is largely based on Kirkland’s routines [10]. The input cells have been calculated by VFF or, in the case of surface strain relaxation (see fig. 1b), by decorating Finite Elements calculation with atomic positions as described in ref. [2]. The detection in the diffraction plane has been divided into circular rings (every segment about 2.5 or 5mrad wide). The results for intermediate thickness have been stored every 4 or 8 slices up to the maximum thickness of 200nm. The slices were 0.2 nm thick. The advantage of the segmentation of the detector in small angular regions is that this permits to introduce after the simulation the efficiency of the detector and different camera lengths. The post processing of data has been performed within the STEM_CELL user interface [12]. An optimal sampling [24] has been used for a single GaAs <110> cell of 12 x 8 pixels and 24 x 16 for alloys where 4 unit cells were simulated to reduce compositional fluctuations. The number of frozen lattice runs was set to 20. The slices have been sampled on 1024 x 1024 pixels while thermal Debye Waller factors were taken from [25]. For simplified simulation the solution of equation (8) has been calculated within STEM_CELL graphic user interface that implements both an Euler and a Runge Kutta solver for the equation.

4. Results and discussion

4.1 Long-range strain

Grillo et al. [5] have studied the HAADF and LAADF contrast from dislocations oriented in a plane orthogonal to the electron propagation direction. In this contribution we will demonstrate the case of ADF contrast from end-on screw dislocation in GaN. The theme has been addressed by different authors and worth mentioning is in particular the work of Cosgriff et al. who carefully studied the effect of surface relaxation by accurate multislice simulations [26].

The first interesting result is that using eq. (6-8) we can reproduce a similar contrast. The strain field of a screw dislocation has been calculated with the classic analytical formula accounting for surface relaxation [26]. The result of simulation for exact on-axis conditions is visible in figure 2a showing both the simulated profile and the expected 2D image. The simulation produces a result qualitatively similar to the rigorous simulation within 1 minute calculation time on a PC. It is, in particular, possible to notice the dip at 1 nm from the core and the slow rise of the intensity with increasing distance. As for the absolute value of the contrast in the dip eq. (6-8) are not able to produce a quantitative simulation as they depend on the parameter choice for the simulation and on the specimen thickness. However, using a standard choice [2] of the main parameters a good agreement is obtained with rigorous simulations.

Using the same equations it is possible to extend the simulations to the case of a tilted sample. In particular figure 2b is obtained with a tilt of 10 mrad. It shows a very peculiar dipole contrast oriented along
the tilt direction with dipole sign that depends on the orientation of the Burgers vector. This shows some similarity with the contrast in 2-beam dark field, however, the dipole orientation is not related to the Bragg vector direction (in this case orthogonal to the TEM foil plane) but depends only on the actual sample mistilt.

![Figure 2.](image)

This has been checked by experiments on a GaN sample prepared in plan-view. Figure 2c shows a typical ADF image with a detection acceptance semi-angle of 46–128 mrad. The image shows a few screw dislocations with a bright core and the predicted dipole-like contrast. In particular, in the top left part of this figure two dislocations with opposite dipole orientations, and therefore opposite dipole signs, are visible. However, except for the sign all dipole have similar orientation. Moreover, the apparent dipole orientation could effectively be changed in the experiment by appropriate tilting.

**Figure 2.** a) Simulation according to eq. (8) of the dependence of the intensity on the distance from the core for a screw dislocation. Specimen thickness was set to 20 nm. The inset shows a 2D rendering of the dislocation. b) simulation of the ADF contrast of a screw dislocation for a mistilt of 10 mrad. c) experimental ADF image (46 – 128 mrad) of screw dislocations.

4.2 Static displacements in alloys: InGaP/GaAs

The demonstration of the importance of the SD in HAADF images has already been presented in [3]. We want to study here, by a comparison between simulations and experiments, the more complicated effect in the middle–low angular regime. To this aim we use the In$_x$Ga$_{1-x}$P/GaAs system. When the In molar fraction approaches $x=51\%$, the two lattice parameters are lattice matched. Moreover, the expected HAADF intensity on the basis of $Z^2$ considerations in the two systems is quite similar. With $a=1.7$ the expected intensity ratio $I_{\text{InGaP}}/I_{\text{GaAs}}$ is 0.896. In this situation static displacement is expected to be dominant since chemical differences are relatively low. The GaAs is a binary system and therefore presents no static displacement while InGaP is an alloy with composition close to 50% where SD is maximum. According to VFF calculation static displacements in InGaP are 0.0044 nm on group-III elements and 0.0086 nm on group-V elements.

A proof of the importance of SD in this system is found in the experiment of figure 3a for different camera lengths on a sample with a rapid and non-linear thickness increase in the horizontal direction. For comparison simulations for a wedge shaped sample are also reported. In this example a direct quantitative comparison with experiments is complicated due to the selective milling of the InGaP layer that can be easily observed in the experimental image. We will perform only a qualitative analysis of the contrast inversion. Simulations demonstrate that a slightly (i.e. 5%) thinner InGaP layer only shifts the point of contrast inversion to larger thickness (in the case of the range 48-128 mrad the shift would amount to about 50 nm). For detection within 30-84 mrad and at very low thickness it is certainly possible to observe that the GaAs is effectively slightly brighter than the InGaP. However, InGaP takes over quite quickly, at a thickness of about 40 nm (according to EELS measurements). The approximate thickness of contrast inversion is indicated by a line in figure 3a.
At small camera length (larger detection angle) the InGaP/GaAs contrast inversion occurs at very thick sample regions. Eventually, at very small camera length (in this case detection angles of 170-470 mrad) no inversion is observed. The Huang scattering due to SD is at the origin of the additional InGaP intensity in the 34-80 mrad detector that builds up within a typical channelling length of some tens of nm. As discussed in the theory section the Huang scattering is concentrated in the region between 30-50 mrad. It is therefore natural that for a detector starting from 30 mrad this contribution is very important while for detection beyond 46 mrad (second row) this contribution is lower.

The simulated distribution of the Huang scattering for this specific system for two different thicknesses is shown in figure 3b that confirms this conjecture. To evaluate the Huang scattering distribution, as in the case of GaAsN, we used the fact that the chemical contribution is very similar, therefore we just calculated the difference in the radial scattering distribution.

It is in particular interesting to notice that the radial distribution of the Huang scattering in the diffraction plane is slightly broader in the case of thicker sample. This is due to multiple scattering of the Huang scattered electrons [27]. The multiple scattering of diffuse waves brings the additional Huang contribution to larger angles and this explains the fact that even with larger detection angle an inversion of the InGaP/GaAs level occurs. It is worth mentioning that this phenomenology can be explored only with the frozen lattice approach since other approximate methods cannot account for multiple scattering.

**Figure 3.** a) Qualitative comparison of simulation and experiment for the interface InGaP/GaAs for different angular detection ranges. The thickness increases in both cases from left to right. The contrast inversion, indicated by the vertical bars, is quite sudden for LAADF in the range 30-84 mrad but becomes slower for increasing detection angle. At very high angle (170-470 mrad) no inversion is observed. The maximum thickness in the simulation is 250nm while it is roughly 350 nm in experiments. Faint striations in the simulations are graphical artefacts due to the discretisation of the grey scale on 255 levels. b) Simulation of the Huang scattering angular distribution (calculated as the difference between InGaP and GaAs) for two thickness values.

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### 4.3 Chemical analysis of quaternary alloys

Having in mind the above effects it is now the turn to tackle the most complex case, the InGaAsN system. This is a complicated system since both In and N content can vary and need to be determined independently. The experiments reported here regard the case of InGaAsN QW on GaAs. The presence of GaAs barriers is useful to have a safe reference.
The large interest in this system is due to the fact that the introduction of In has a real “chemical effect” that should be well visible in HAADF. Conversely, N is a very light element and should be visible mainly through the static displacement it introduces in the lattice. A convincing proof of this fact is visible in simulations in figure 1b showing that practically no effect should be visible in the low angle regime. It is therefore reasonable to believe that a comparison between low angle and high angle ADF experiments should permit to separate In and N contribution.

A systematic experimental study on the effect of detection angles on the chemical contrast has been published for ternary InGaAs/GaAs and binary SiGe/Si systems and used to derive the In or Ge composition [28, 29]. We will treat here the information at different angles as independent in order to separate the In and N contribution. In order to study quantitatively the above phenomenon simulations have been performed for 26 sets of In and N compositions. The resulting dependence on \( x \) and \( y \) of LAADF (26<\( \theta \)<161 mrad) and HAADF (50<\( \theta \)<301 mrad) has been plotted in figures 4a and 4b. The data have been fitted to a 2D polynomial to obtain smooth interpolation.

Figure 4. Outcome of simulations for LAADF (a) and HAADF (b) imaging of \( \text{In}_{x}\text{Ga}_{1-x}\text{As}_{1-y}\text{N}_{y} \). The simulations refer to a detection range of 26-161 mrad and 50-300 mrad respectively. The intensity levels are normalized to GaAs.

Figure 5. a) Experimental profiles of the InGaAsN quantum well for different detection range. The underlying image is an experimental image for the detector range 50-300 mrad. b) Compositional profiles as deduced from figure 5a and comparison with simulations in figure 4a. The intensity at interfaces is strongly affected by surface strain artefacts.

The simulations refer to a thickness of 88 nm, relevant for comparison with experiments. All intensities have been normalized to the GaAs level. The simulations confirm our original guess that low angle ADF is much more sensitive than HAADF to the N content; in fact, in HAADF the intensity remains the same for any N content up to at least 7% (on Y-axis); it increases monotonically for increasing In content (on X-axis). The situation is completely different for low angles where N has an important effect. The method to extract the In and N content is based on a point to point comparison of normalized LAADF and HAADF intensity between simulation and experiment.
Figure 5a shows an experimental image under HAADF conditions (50-300 mrad); superimposed are normalized profiles for different detector conditions. The inner angle is directly indicated while the outer is about 6 times larger. Both the image and the profiles show the characteristic dip in the barrier and the smooth profile that are typical artefacts of surface strain relaxation [2]. It is clear that this kind of effect reduces the reliability of the composition map.

The rise of the intensity in the centre of the QW for decreasing detection angle is conversely the effect of an increase of Huang scattering due to SD which is in larger part contributed by N and to a minor extent by In. A reduced detection angle produces an increase of the intensity also in the barriers where both In and N are absent. This is clearly due to the Huang scattering produced by the long-range strain according to equation (1). Figure 5b is an attempt to reconstruct In and N profiles from the profiles obtained with inner detection angles of 26 and 50 mrad. The profiles are clearly affected by the artefacts related to the strain and additional Huang scattering in the regions close to the interface. The most reliable region remains the centre of the QW where the atomic column tilt due to surface relaxation should be 0. Here the best fit between simulations and experiments is for $[x]=19\%$ and $[y]=1\%$, which is to be compared with nominal values of 20.5% and 2.4%. This result can be considered as an important progress toward quantitative compositional evaluation of quaternary alloys. The relative error on In evaluation is low as the quantitative analysis is related to the well established high-angle scattering [3, 28]. The larger relative error on the apparent content of N that is more related to low-angle scattering. Therefore, a more accurate description of LAADF scattering is necessary [5].

Finally, figure 5a permits to outline a possible strategy to remove the strain effects from the composition profile at least in simple cases. In fact we notice that the intensity dip due to dechanneling can be partially compensated by Huang scattering and the compensation increases with decreasing detection angle. Unfortunately, it can be estimated that a satisfactory compensation would occur for very low detection angle possibly close to annular bright field conditions where eventually other effects take place. Alternatively, it is possible to extrapolate the Huang scattering at very low angle from data in the allowed range. The result is just a linear combination of HAADF and LAADF profiles. In figure 6 we found that a combination of the profiles at 50 and 26 mrad in the form of $I_{50}+1.5(I_{26}-I_{50})$ shows a good compensation of strain effects. This profile can be considered a good qualitative demonstration of the abruptness of the interface which could not be deduced from figure 5a alone, but no separate composition information on In or N can be deduced from this plot.

![Figure 6. Combinatorial profile obtained as $I_{50}+1.5(I_{26}-I_{50})$. This linear combination of profiles permits to minimize the strain effect.](image)

5. Conclusions
We have analyzed systematically for screw dislocations, InGaP and InGaAsN the contrast mechanism in ADF imaging under different detection conditions. In the examples we demonstrated that it is possible to account for the strain contrast in dislocations and the contrast inversion as a function of static displacement, however, the most interesting application of a multiple detection condition approach is in the case of InGaAsN where the In and N compositions can be independently estimated. In spite of the demonstration of principle, the uncertainty on N composition and the problem with surface relaxation indicate that the research has to be oriented toward a more detailed description of the low angle scattering. In conclusion, it is becoming evident that advanced experiments with multiple detection ranges and simulations can help solving the complex interplay among composition, strain and static disorder.
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