Aberration-corrected STEM and EELS of semiconducting nanostructures

K Cui¹², S Hosseini Vajargah¹, S Y Woo¹, M Couillard¹, S Lazar³, R N Kleiman⁴, D A Thompson⁴ and G A Botton¹*

¹ Department of Materials Science and Engineering, McMaster University
1280 Main Street West, Hamilton, Ontario, L8S4L7, Canada
² Department of Electrical and Computer Engineering, McGill University, Montreal, Quebec, H2A3A7, Canada
³ FEI Company, Eindhoven, The Netherlands
⁴ Department of Engineering Physics and Centre for Emerging Devices Technology, McMaster University, 1280 Main Street West, Hamilton, Ontario, L8S4L7, Canada

E-mail: gbotton@mcmaster.ca

Abstract. We review some applications of aberration–corrected electron microscopy for the detailed characterization of semiconducting nanostructures using a combination of high-angle annular dark-field scanning transmission electron microscopy and electron energy loss spectroscopy. The study of self-assembled quantum wires shows that it is possible to determine the composition of the nanostructures with better than 1 nm resolution down to the atomic level while the contrast in the high-angle annular dark-field images is used to determine the presence of wetting layers separating quantum wires and the strain field arising from the local compositional changes. The local measurements of energy loss spectra demonstrate the shift of plasmon peaks consistent with the changes in lattice parameters. High-angle annular dark-field images are also used to study the contrast in GaSb thin films deposited and study the presence of anti-phase domain boundaries. These examples show that aberration-corrected microscopy combined with electron energy loss spectroscopy provide not only enhanced resolution but also increased sensitivity to atomic site compositional changes.

1. Introduction

Group III-V semiconductor nanostructures such as self-assembled nanowires have potential uses in optoelectronic applications such as long wavelength lasers [1] and broad band emitters [2], and have been demonstrated to emit at room temperature in the telecommunication range [3]. Also of great promise for optoelectronic applications such as photodetectors and photovoltaics is the use of III-V compounds grown on Si. While it is clear that in the latter case, the greater range of optical properties of III-V compounds gives much enhanced flexibility in design and tunability of the optoelectronic properties, there is still significant difficulty in growing high-quality films exhibiting low defect densities, due to the intrinsic lattice parameter mismatch between the films and substrate. Although conventional electron microscopy provides excellent insight on the nature of the defects, defect density, and allows to visualize both chemical/compositional contrast, it is now apparent that the new imaging techniques and compositional analysis methods can provide complementary and essential
information that can be used to gain further insight on the very local composition, strain, and local changes in polarity information that could only be obtained with great effort with conventional techniques due to the strong dynamical scattering. The recent development of electron microscopes with aberration correctors has opened the field for such studies by providing enhanced stability of the microscopes, improved resolution and spectroscopic sensitivity.

Here we present some examples of studies related to the characterization of InAs quantum wires grown on quaternary alloys lattice matched to InP. These materials have shown photoluminescence emission at room temperature and tuneability of emission with changes in quantum wire size and growth conditions. We show that the improved resolution of the aberration corrected instruments, together with atomic resolved electron energy spectroscopy mapping, provide key information about the presence of wetting layers connecting the quantum wires as well as the detection energy shifts in the plasmon resonances. We demonstrate that these changes are very localized to the quantum wire structures. We also highlight recent results related to the study of growth mechanisms and defects of GaSb thin films deposited on Si where high-resolution imaging provides key evidence of local polarity change complementing convergent beam electron diffraction information as well as direct visualization of the anti-phase boundaries highlighting the transition between two out-of-phase domains.

2. Experimental methods
Conventional contrast imaging was carried out on a Philips CM12 microscope operated at 120keV. Ultrahigh-resolution microscopy was carried out with a FEI Titan 80-300 “cubed” microscope equipped with a high-brightness source, an electron monochromator, a CEOS hexapole corrector for the probe-forming lens and a similar corrector for the image-forming lens. The system is also fitted with a Gatan 866 energy filter for electron energy loss spectroscopy. This system is also located in an ultrastable environment with demonstrated temperature stability of better than 0.1 °C, low electromagnetic fields (<0.05 mG) and low vibrations [4]. Quantitative elemental maps of the In composition were also obtained with a field emission scanning transmission electron microscope (JEOL 2010F) fitted with a Gatan Tridiem Energy Filter. Samples for conventional imaging were prepared with the conventional mechanical polishing method, followed by dimpling and ion milling to electron transparency (Gatan PIPS system equipped with a liquid N$_2$ cold stage, 4 keV argon ions at an incident angle of ±4°). For ultrahigh resolution electron microscopy, samples were prepared by the tripod method to electron transparency using an Allied Tech Multiprep unit. Samples are then cleaned using a Technoorg Linda Gentle-mill using 300-500 eV ions to remove the residues of polishing contamination from the samples. This last step, as well as the mechanical polishing, has been shown to provide excellent quality samples suitable for the aberration-corrected microscopy work shown here.

3. Self-assembled quantum wires
Self-assembled InAs quantum wires were grown by gas source molecular beam epitaxy on the surface of InGaAlAs which is used as a confinement/capping layer for the electrons (figure 1). This quaternary alloy is also lattice matched to the InP substrate so that the In content is 0.53 of the group III elements. These wires have been shown to grow along the [01\bar{1}] direction of the InGaAlAs surface as a consequence of the anisotropy of the surface reconstruction and nature of surface steps, as demonstrated from AFM imaging [5]. With the deposition on the InGaAlAs of 4-5 monolayer of InAs, quantum wires structures are formed as a result of strain relaxation due to the 3.2% lattice mismatch. As a consequence, the effects of strain, composition and curvature of the quantum wires are intimately linked with each other and ultimately control the chemical potential for group III elements of the surface front during subsequent MBE overgrowth [6]. Due to the complexity of the structures, the effect of strain and the coupling of this strain with the composition also affect the image contrast. Therefore, several imaging methods and chemical analysis tools have been used to characterize these materials. In the following sub-sections we present imaging with conventional contrast transmission electron microscopy, elemental mapping with EELS as well as aberration-corrected imaging so as to
elucidate some of the structural features observed in these materials. Previous work has used this chemical information, combined with finite elements calculations, so as to determine the stress state in the wires and the surrounding areas [6].

Figure 1. General overview of the wire structure with the confinement layers. The cross-section shows the InAs wires enclosed in the InGaAlAs confinement layers, the InP substrate and capping layer. An additional InGaAlAs layer is used as a reference to generate an uncapped array of InAs wires at the very top of the surface used for atomic force microscopy analysis.

3.1. Conventional diffraction contrast imaging
Conventional imaging shows the complexity of the structures and the interplay between the strain and composition in TEM images. By selecting the chemically sensitive $\mathbf{g}=[200]$ reflection in two-beam conditions (figure 2a), the contrast between the wires and the surrounding quaternary InGaAlAs confinement layers is highlighted. Alternatively, the selection of the $[022]$ reflection highlights the strain contrast and the strong strain interaction between the wires (figure 2b). In the case of multiple stacks of wires used for laser applications, it is also clearly possible, when using the chemically sensitive $\mathbf{g}=[200]$ reflection to see the change in the alignment between the subsequent layers of quantum wires. With this imaging method, we have also detected dark bands above the wires and connecting the layer stacks as shown in figure 3 and [7]. Since there is potentially ambiguity in the chemically sensitive images due to possible residual strain contrast, elemental mapping with EELS was used to assess such effect.

Figure 2a. Conventional cross-sectional TEM imaging of quantum wires. Chemically sensitive contrast image using the $\mathbf{g}=200$ reflection

Figure 2b. Conventional cross-sectional TEM imaging of quantum wires. Strain sensitive contrast image using the $\mathbf{g}=022$ reflection.
3.2. Chemical analysis with EELS and EDXS

In order to identify without any ambiguity the origins of the stripe-like contrast observed in the conventional TEM images as seen in figure 3, we used EELS spectrum imaging and the detection of the In $M_{4,5}$ edge in the spectra. Using an energy window from 450 eV to 520 eV energy loss and the power-law background extrapolation, we can extract the In signal and this can be quantified based on the reference InGaAlAs layer (with In content of 0.53) assumed to be of known composition based on X-ray diffraction measurements. The resulting In maps thus provide quantitative composition with a resolution confirmed to be in the 1 nm range when using a non-aberration-corrected field emission TEM [5]. Elemental maps show that In-rich stripes are detected not only on single layer structures [5] but also between multiple stacks of wires (figure 4). Elemental maps of a row of wires (figure 5), suggest relatively uniform composition from wire to wire but, at the resolution of the conventional scanning transmission electron microscopy (STEM) instrumentation, the data does not provide clear evidence on the possible presence of a wetting layer, connecting adjacent wires. This wetting layer is an important factor that determines whether the growth mechanisms follows the Stranski-Krastanov (island formation on wetting layer) or Volmer-Weber (island without wetting layer), In addition, a wetting layer has the potential of affecting the electron confinement and the interactions between the quantum wires and thus any simulations on the optical properties. Therefore, additional high-resolution mapping in combination with high-angle annular dark-field imaging are required.

![Figure 4. EELS SI map with multiple wires. Left: HAADF image and boxed area of interest; Right: In $M_{4,5}$ elemental map and quantitative gray scale lookup table.](image)
In order to ascertain the best possible resolution achievable in elemental maps for the elements of interest here, we have used a reference InGaAlAs material as a test case. Atomic resolution mapping has, for example, been demonstrated in oxides with a relatively large lattice parameter [8] but no atomic resolved mapping on III-V alloys has been reported in the open literature so far. With aberration correction and a probe current approaching 100 pA, elemental maps of In and As demonstrate that atomic resolved mapping is indeed possible and that the peak intensity of the In or As maps is specific to the column corresponding to the related high-angle annular dark field (HAADF) intensify (figure 6). For example, the In map has peaks precisely corresponding to the brightest dumbbell pair coinciding with the In column while the As map has peaks precisely coinciding to the lower intensity dumbbell in the HAADF image. These results demonstrate that, in the conditions used for mapping and in particular for a collection semi-angle larger than 80 mrad, the signal is localized to the atomic column channelling the electron intensity. Since the projected spacing between the In and As columns is in the order of 1.6 Å, this demonstrates a spatial resolution sufficient to distinguish the two atomic columns.

Although the detection efficiency of a standard Si(Li) energy dispersive X-ray spectrometer is significantly lower than what can be achieved with EELS and the newest silicon drift detectors [9] by a factor of about 9 (based exclusively on the solid angle), our measurements show that atomic resolved signals can also be detected in an EDXS map (figure 7).

![Figure 5. EELS SI map with multiple wires with the possible wetting layer.](image)

**Figure 5.** EELS SI map with multiple wires with the possible wetting layer.

![Figure 6a. Atomic resolved EELS maps, for In (M_{4,5} edge) at 450-520 eV loss. Top: HAADF signal, Bottom: In M_{4,5} map. Image quantified following multivariate statistical analysis and noise removal [10]. The thickness of the sample, relative to the total inelastic mean free path at 200keV, (t/\lambda), is approximately 0.2.](image)

**Figure 6a.** Atomic resolved EELS maps, for In (M_{4,5} edge) at 450-520 eV loss. Top: HAADF signal, Bottom: In M_{4,5} map. Image quantified following multivariate statistical analysis and noise removal [10]. The thickness of the sample, relative to the total inelastic mean free path at 200keV, (t/\lambda), is approximately 0.2.

![Figure 6b. Atomic resolved EELS map for As (L_{2,3} edge) at 1300-1400 eV loss. Top: HAADF signal, Bottom: As L_{2,3} map. Image quantified following multivariate statistical analysis and noise removal [10]. The thickness of the sample, relative to the total inelastic mean free path at 200keV, (t/\lambda), is approximately 0.2.](image)

**Figure 6b.** Atomic resolved EELS map for As (L_{2,3} edge) at 1300-1400 eV loss. Top: HAADF signal, Bottom: As L_{2,3} map. Image quantified following multivariate statistical analysis and noise removal [10]. The thickness of the sample, relative to the total inelastic mean free path at 200keV, (t/\lambda), is approximately 0.2.
The raw EDXS data based on the In L and As K peak signals mapped in figure 7 shows in fact a regular pattern centred at the atomic columns. These results show that, in spite of the known effects of beam broadening, the contribution of channelling of electrons on the atomic columns and the highly focused electron beam, significant intensity changes are detected between the different atomic positions, particularly for the In signal which is most intense and has lower noise levels. Further work would be necessary to ascertain whether, for a sharp interface, these results are fully consistent with atomic resolution elemental mapping as demonstrated in EELS experiments [8].

Figure 7. Elemental maps obtained with a Si(Li) drift (0.13 sr solid angle) detector on the aberration-corrected STEM. Top image is the annular dark-field (ADF) signal recorded simultaneously as the EDXS signal. Middle image is the As signal and bottom image is the In. (dwell time 50 msec/pixel). Maximum counts for the In map is 25, for the As map 19. 32x15 pixel map (2.1x1.1nm field of view). The thickness of the sample, relative to the total inelastic mean free path at 200keV, (t/λ), is approximately 0.2.

Because of the demonstration of atomic resolution in EELS maps for In and As, we have focused our attention to the characterization of the wetting layer separating two quantum wires studied previously above at lower resolution (figure 5). The two dimensional EELS map extracted over the area between two adjacent quantum wires shows that the EELS intensity in the In column increases over 3 atomic planes (figure 8) thus confirming that there is a residual wetting layer separating two wires. The implication of the presence of this wetting layer is important in terms of both the growth mechanism and also for the electronic coupling and interaction between the wires. As the nature of a potential well between the wires is intrinsically affected by the local composition, the presence of this wetting layer suggests an anisotropy in the confinement of the wires with a lateral potential well somewhat reduced between adjacent wires. The effect of this feature on the optical properties has been discussed recently [11].

Figure 8. HAADF image of the wetting layer (left) and atomic resolved EELS map of In (right) obtained from the highlighted area surrounded by a white frame. The In map was extracted from the spectrum image following background subtraction of the M_{4,5} edge. Data was processed with multivariate statistical analysis to reduce some of the noise contribution [10].

3.3. Aberration-corrected imaging
The EELS elemental mapping of the wetting layer suggests that there is an increased In concentration within a few atomic planes from the confinement layer. With the enhanced HAADF resolution of a probe-corrected STEM, it is also possible to complement the chemical information from EELS with the intensities of HAADF images (figure 9). From an intensity profile, it is apparent that the intensities of the group III column across the wetting layer increases significantly while the intensity of the group V column is only slightly increased. The background intensity between the dumbbell columns also shows increased intensity, due to the tails of the probe, contributing to an overall increase of the
HAADF signal. With these profiles, it is very clear that the wetting layer consists of 3 to 4 atomic layers with increased In concentration. By combining the EELS maps with the HAADF images, unambiguous confirmation of the presence of the wetting layer is thus possible. The measurement of the localized strain using the geometric phase analysis (GPA) from the HAADF images is also feasible in a very stable instrument and microscope environment. This technique was initially developed for phase contrast HRTEM imaging [12] and demonstrated recently on HAADF images of the GaSb/Si interface [13]. The HAADF image of two stacked wires shows the strong contrast between the confinement layer and the InAs wires while the analysis of strain with the GPA using the [111] and [T11] reflections shows the local strain in the growth direction extending almost exclusively to the wires themselves, at least within the noise level of the observed in the confinement layer (figure 10).

Figure 9a. HAADF image of wetting layer. The vertical thin line indicates the area selected for the intensity profile shown in figure 9b.

Figure 9b. HAADF intensity profile from the same area identified in figure 9a from point A (pixel “0”) to point B.

Figure 10a. HAADF image of two stacked quantum wires. Surrounded by the straight lines are the “boundaries” of the dots as a reference comparison in figure 10b. The selected frame is the same as for figure 10b.

Figure 10b. Strain image εxx obtained from the geometric phase analysis of the HAADF image shown in figure 10a. The straight lines identify exactly the same area shown in figure 10a and delimit the boundaries with increased HAADF signal. The gray scale bar at the bottom of the figure is the strain in % units.
3.4. Low-loss mapping on quantum wires
Measurements of the plasmon energies have been used extensively as an effective way to measure indirectly the composition of samples in alloys [14] and metal hydrides [15]. In the field of semiconductors, plasmon imaging has been used to distinguish Si and SiO\textsubscript{x} based on the strong contrast between the two phases. [16]. More recently, low energy losses have been used to image surface plasmon polaritons in metallic nanowires in the infra-red optical regime and in the visible range [17, 18]. In III-V compounds, bulk plasmon shifts can potentially reveal changes in valence electron density resulting from changes in the composition and volume of the unit cell due to the strain. Two-dimensional maps acquired on the cross-sectional samples reveal very significant changes in the bulk plasmon position and shifts that can be quantified (figure 11). These changes are consistent with the increase in the local lattice parameter as demonstrated from the GPA measurements and further work at lower energy losses is in progress to detect any possible interaction between the valence electrons between stacked quantum wires.

![Figure 11a. HAADF STEM image (top) of stacked quantum wires and plasmon shifts map (bottom) of exactly the same area. Gray scale bar shows the energy shift of the plasmon peak position.](image)

![Figure 11b. Low-loss spectra extracted from the spectrum image dataset used to map the peak positions extracted in figure 11 (a). Spectra taken from confinement layers (spectra 1, 2, 4,5 and 6) and quantum wires (spectra 3 and 7). The vertical dashed line depicts the peak positions of spectra taken from the capping layers as a reference.](image)

4. GaSb on Si layers
In the highly lattice mismatched system of GaSb films grown on Si by molecular beam epitaxy, the nature of the defects is still a subject of significant debate. Even in films grown with an AlSb buffer layer, arrays of dislocations have been observed with conventional imaging [19], high-resolution TEM [20] and HAADF [13] but there have not been conclusive observations of the predicted anti-phase boundaries (APBs). Even though APBs are expected due to the likely presence of quarter-unit cell surface steps [21], direct and unambiguous evidence of APB in films has not been demonstrated so far in spite of the importance of such defects on the electronic properties. By growing GaSb films and a strained-layer-supperlattice (SLS) of AlSb/GaSb to prevent further propagation of threading dislocations and possibly other extended defects, we have thus investigated whether anti-phase
domains (APDs) are indeed present and their structure at the atomic level with the ultimate purpose of understanding the interaction between defects and the potential electrical activity of the range of defects present in the films.

4.1. Conventional diffraction contrast
Using a combination of diffraction contrast imaging, and convergent beam electron diffraction (CBED) with conventional TEM we have identified adjacent APDs separated by boundaries (figure 12). The superlattice reflection dark-field (DF) images show that some of the APBs intersect the SLS and propagate to the surface while some self-annihilate (figure 12). Misfit dislocations, threading dislocations and microtwins are also generated at the Si interface (figure 13) even when an AlSb buffer layer is present. Previous work using cross-sectional TEM and EDXS mapping has shown that the AlSb buffer layer is present as islands [13] and that a network of misfit Lomer dislocations is present. CBED measurements have confirmed that domains separated by the defects shown in figure 12 are indeed of opposite polarity [22].

The SLS also allow clear visualization of the effect of microtwins on the growth of the films with strong contrast enhanced using the chemically sensitive reflections. Already from the diffraction contrast image, it is possible to identify the tilting of the SLS within the microtwin (figure 13) and the fact that the microtwins do not necessarily appear to form right at the interface with the substrate. Due to the finite thickness of the foils, however, this effect does not consider the propagation of the microtwins in the third dimension as discussed further in [22]. These observations are discussed in the following section.
4.2. Aberration corrected microscopy

Imaging with higher resolution and sensitivity to the atomic contrast demonstrates the detection of local deviations from a planar surface due to the presence of the microtwins (figure 14). HAADF imaging also shows the formation of atomically sharp interfaces between the strained layers, with the transition between the AlSb and GaSb just occurring over one single atomic plane (figure 15). HAADF images also demonstrate the detection of lone Ga atoms at the $\Sigma=9$ tilt boundary resulting from intersection of two coherent twin boundaries. Defects are marked by the array of continuous arrows in figure 15.

Figure 13. Conventional TEM DF image of the AlSb/GaSb strained-layer-superlattice and various microtwins intersecting the layers using the $g=\overline{2}0$ fundamental reflection. Pointed by the dotted arrow is evidence of a growth front change from planar to stepped due to the microtwin.

Figure 14. HAADF image of the strained-layer-superlattice intersecting a microtwin. The growth front locally changes from a planar interface. The bright layer is the GaSb component of the SLS while the dark layer is the AlSb component of the SLS.
Figure 15. HAADF image of a defect in the strained-layer-superlattice due to the propagation of a twin from the interface between GaSb and Si. Also visible are missing dumbbell pairs pointed to by the continuous arrows and an atomically sharp interface between the AlSb and GaSb layers. In the AlSb layer, only the Sb column is visible in the dumbbell pair while, on GaSb, both the Ga column (of weaker intensity) and the Sb columns are visible.

Also very clearly visible, based on the intensity of the reversal of the dumbbell pairs in the SLS, are APBs that appear to be about 4-5 atomic planes wide (figure 16). The contrast of the APBs in the AlSb clearly suggest that the Sb column also changes polarity over the thickness of the TEM sample and that the intrinsic width is thus most likely less than the apparent projected dimension. The region where the polarity changes within the APB in the AlSb component of the SLS thus appears as a bright band due to the channelling of electrons on the Sb atoms, now present on both atomic sites over the thickness of the thin film rather than just one of the two sites.

Figure 16. HAADF image of an APB detected in the AlSb/GaSb strained-layer-superlattice as delimited by the two lines. The AlSb layer is identified by the lower intensity at the right of the image (growth direction is from left to right) and the GaSb is on the right of the image. HAADF simulations of AlSb and GaSb images are show as inserts left and right respectively. In the simulations and in the experiments, only the Sb column is visible in AlSb while both the Sb and Ga columns are visible in GaSb. Within the APB in the AlSb layer, both columns are of significant intensity. For clarity of the APB visibility, the image is rotated with respect to the growth direction, which is now from left to right.

At the interface with the Si substrate we have also observed APBs and interactions of these with other defects. Further work detailing the origins of the APBs at the interface with the substrate will be presented elsewhere [23]. These results with high-resolution HAADF clearly show that aberration correction not only provides improved resolution but also enhanced detection of anti-phase boundaries and potentially electrically active lone atomic sites.
5. Conclusions
We have demonstrated the application of conventional and aberration-corrected electron microscopy for the detailed study of semiconducting nanostructures. The application of these two techniques has been shown in the context of the study of structure, strain and composition of InAs quantum wires and in the detailed investigation of GaSb films deposited on Si. Chemical mapping with atomic resolution has been successfully used to determine unambiguously the chemical nature of a wetting layer present between InAs wires while plasmon mapping has been used to identify changes in the local electron density. High-resolution high angle annular dark-field imaging has provided insight on the distribution of strain in quantum wires and the detection of defects in GaSb films grown on Si. The sensitivity to atomic contrast in HAADF imaging has also been exploited to provide insight on the nature of anti-phase boundaries and the interaction between these and strained-layer-superlattices.

Acknowledgements
This work was partially supported by the Ontario Centres for Excellence, the Discovery Grants program of NSERC and ARISE Technologies. The electron microscopy was carried out at the Canadian Centre for Electron Microscopy, a national facility supported by NSERC and McMaster University.

References
[1] Reithmaier J P, Eisenstein G, and Forchel A 2007 Proc. IEEE 95, 1779
[2] Lin Z C, Lu C Y and Lee C P 2006 Semicond. Sci. Technol. 21, 1221
[3] Grundmann M 2002 Nano-Optoelectronics (New York: Springer)
[4] Botton G A 2008 Int. J. Nanotechnology 5 1082
[5] Cui K, Robertson M D, Robinson B J, Andrei C M, Thompson D A and Botton G A 2009 J. Appl. Phys. 105 094313
[6] Cui K, Robinson B J, Thompson D A and Botton G A 2010 J. Appl. Phys. 108 034321
[7] Cui K, Robinson B J, Thompson D A and Botton G A 2010 J. Cryst. Growth 312 2637
[8] Botton G A, Lazar S and Dwyer C 2010 Ultramicroscopy 110 926
[9] Schlossmacher P, Klenov D O, Freitag B and von Harrach H S 2010 Microscopy Today 18 14
[10] Dudeck K, Couillard, M, Lazar S, Dwyer C, and Botton G A 2011 Micron, in press
[11] Cui K, Robinson B J, Thompson D A and Botton G A 2011 J. Appl. Phys. 109 124311
[12] Hýtch M J, Snoeck E and Kilaas R 1998 Ultramicroscopy 74 131
[13] Hosseini Vajargah S, Couillard M, Cui K, Ghanad Tavakoli S, Robinson B, Kleiman R N, Preston J S and Botton G A 2011 Appl. Phys. Lett. 98 082113
[14] Hunt J A 1995 Microbeam Analysis (New York: VCH Publishers) p 215
[15] Woo O T and Carpenter G J C 1992 Microsc. Microanal. 3 35
[16] Carpenter G J C 2004 Microsc. Microanal. 10 435
[17] Rossouw D, Couillard M, Vickery J, Kumacheva E and Botton G A 2011 Nano. Lett. 11 1499
[18] Nelayah J, Kociai M, Stephan O, Garcia de Abajo F J, Tence M, Henard L, Taverna D, Pastoriza-Santos I, Liz-Marzan L M and Colliex C 2007 Nat. Physics. 3 348
[19] Akahane K, Yamamoto N, Gozu S, Ueta A and Ohtani N 2006 Thin Solid Films 515 748
[20] Kim Y H, Lee J Y, Noh Y G, Kim M D, Cho S M, Kwon Y J and Oh J E 2006 Appl. Phys. Lett. 88 241907
[21] Kroemer H 1987 J. of Cryst. Growth 81 193
[22] Woo S Y, Hosseini Vajargah S, Ghanad-Tavakoli S, Kleiman R N and Botton G A, to be submitted.
[23] Hosseini Vajargah S, Kleiman R N, Preston J S and Botton G A, to be submitted.