MOVPE grown self-assembled Sb-based quantum dots assessed by means of AFM, TEM, and PL

P. Möck, G.R. Booker, N.J. Mason, E. Alphandéry, R.J. Nicholas

Abstract: Self-assembled Sb-based quantum dots (QDs) were grown by metal-organic vapour phase epitaxy and assessed by means of atomic force microscopy, transmission electron microscopy and photoluminescence. Two series of InSb QDs in a GaSb matrix were grown at 490 ± 10 °C and luminesced in the mid-infrared at about 1.7 µm. Reductions in the \( \frac{\text{III}}{\text{V}} \) ratio and growth rate as used for the second series resulted in a change of the morphology of the InSb islands from hillocks without facets and a low level of order to dumbbell shaped islands with distinct facets and a higher level of order. Self-assembled GaSb islands were grown on GaAs at 550 °C and assessed for comparison purposes by means of AFM.

1. Introduction

In recent years, there has been an increasing interest in self-assembled semiconductor quantum dots (QDs) [1], leading to “paradigm changes in semiconductor physics [2]”. In order to make QDs useful in opto-electronic devices, they should simultaneously satisfy the following conditions: dimensions smaller than the three-dimensional Bohr radius of the semiconductor material (which leads to quantised energy levels inside the QDs), sufficiently high size uniformity and number density, and no defects such as misfit dislocations [3]. A sufficiently high size uniformity of self-assembled QDs is thought to be achievable by means of self-ordering processes [1], [3].

Heteroepitaxy in the Stranski-Krastanow growth mode is considered to be one of the most promising routes towards the fabrication of QDs in many material systems. Little work has been done so far to extend the range of wavelength at which QD based opto-electronic devices could operate in the mid-infrared region of the electromagnetic spectrum [4]. In that wavelength region, Sb-based III-V semiconductors are the material of choice, although kinetic effects complicate the growth of such systems by means of metal-organic vapour phase epitaxy (MOVPE) [5].
Self-assembled GaSb islands on GaAs were first grown in our group as early as 1989 in an attempt to produce very thin quantum wells [6], as happened four years earlier at the research labs of the Centre National d’Etudes des Telecommunications, Paris, in the case of InAs islands on GaAs [7]. Here we report on Sb-based QDs that were grown by MOVPE and characterised by means of atomic force microscopy (AFM), transmission electron microscopy (TEM) and photoluminescence (PL). We will be mainly concerned with comparing results from our previous reports on InSb QDs in a GaSb matrix to new results that have been obtained for different growth conditions. This will require some quoting from our previous papers [8-10].

In addition, some results on GaSb islands on GaAs substrates will be included for further comparisons. These results will be mainly used to compare the self-ordering levels of GaSb on GaAs with those of InSb on GaSb grown under different conditions. We will employ the phenomenological classification scheme of the hierarchy of QD self-ordering mechanisms as given in ref. [1] in a slightly modified form for this purpose. Since this is only a qualitative scheme with not well defined boundaries, one should not be surprised that there may be different interpretations of the AFM images we will present.

2. Experimental

Two series of InSb QDs samples were grown by MOVPE using trimethyl sources in an atmospheric pressure reactor [11-12] using nominal on-axis (001) substrates of either GaAs or GaSb. An overview of the growth parameters of the two InSb on GaSb sample series is given in Tables 1a and 1b.

While the first sample series, A to F, was grown at a rate of approximately one to two monolayers per second InSb deposition time, the growth rate was somewhat lower for samples G1 and G2. The other samples of the second series, H to K, were grown at half the rate of samples A to F. This was achieved by halving the trimethylindium flow rate. The GaSb on GaAs sample (X) was grown by depositing GaSb directly on an “Epi-ready™” GaAs substrate at 550 °C for 5 s, resulting in a nominal epilayer thickness of about 2.5 nm.

While uncapped samples were used for AFM analyses, plan-view TEM was best performed on samples with a cap thinner than 100 nm, such as specimen A, but was also done on uncapped samples. Conventional plan-view TEM chemical jet thinning with Cl₂ in methanol of samples on GaAs substrates benefited from the different etch rates of GaAs and the GaSb buffer layer, resulting in wider areas that were transparent at 200
kV acceleration voltage. PL required a GaSb capping layer in order to give a reasonably high signal at a probing temperature of 4 K. Further growth and structural parameters of the samples as well as details on the experimental assessments were given previously [8-10].

<table>
<thead>
<tr>
<th>sample code</th>
<th>substrate / buffer layer</th>
<th>trimethylindium (TMIn) flow [cm$^3$ / minute]</th>
<th>InSb growth temperature [°C]</th>
<th>InSb deposition time [s]</th>
<th>GaSb capping layer</th>
</tr>
</thead>
<tbody>
<tr>
<td>A (2573)</td>
<td>GaSb / 0.08 µm GaSb</td>
<td>500</td>
<td>480</td>
<td>2</td>
<td>0.08 µm GaSb</td>
</tr>
<tr>
<td>B (3178)</td>
<td>GaAs / 1.95 µm GaSb</td>
<td>500</td>
<td>480</td>
<td>3</td>
<td>0.48 µm GaSb</td>
</tr>
<tr>
<td>C (3182)</td>
<td>GaAs / 1.95 µm GaSb</td>
<td>500</td>
<td>480</td>
<td>4</td>
<td>0.48 µm GaSb</td>
</tr>
<tr>
<td>D (3201)</td>
<td>GaAs / 1.95 µm GaSb</td>
<td>500</td>
<td>480</td>
<td>1</td>
<td>0.48 µm GaSb</td>
</tr>
<tr>
<td>E (3202)</td>
<td>GaAs / 1.95 µm GaSb</td>
<td>500</td>
<td>480</td>
<td>2</td>
<td>0.48 µm GaSb</td>
</tr>
<tr>
<td>F (3262)</td>
<td>GaAs / 1.95 µm GaSb</td>
<td>500</td>
<td>480</td>
<td>3</td>
<td>none</td>
</tr>
</tbody>
</table>

**Table 1a:** Growth parameters for the first series of InSb on GaSb samples.

The trimethylantimony (TMSb) flow was kept constant at 100 cm$^3$ per minute (sccm), resulting in a $\text{III}/\text{V}$ ratio of 5.

<table>
<thead>
<tr>
<th>sample code</th>
<th>substrate / buffer layer</th>
<th>trimethylindium (TMIn) flow [cm$^3$ / minute]</th>
<th>InSb growth temperature [°C]</th>
<th>InSb deposition time [s]</th>
<th>GaSb capping layer</th>
</tr>
</thead>
<tbody>
<tr>
<td>G1 (3376)</td>
<td>GaAs / 0.45 µm GaSb</td>
<td>*</td>
<td>500</td>
<td>4</td>
<td>none</td>
</tr>
<tr>
<td>G2 (3376)</td>
<td>GaSb / 0.45 µm GaSb</td>
<td>*</td>
<td>500</td>
<td>4</td>
<td>none</td>
</tr>
<tr>
<td>H (3508)</td>
<td>GaAs / 2.67 µm GaSb</td>
<td>250</td>
<td>500</td>
<td>8</td>
<td>0.18 µm GaSb</td>
</tr>
<tr>
<td>I (3512)</td>
<td>GaAs / 2.67 µm GaSb</td>
<td>250</td>
<td>500</td>
<td>8</td>
<td>0.18 µm GaSb</td>
</tr>
<tr>
<td>J (3571)</td>
<td>GaAs / 2.67 µm GaSb</td>
<td>250</td>
<td>490</td>
<td>8</td>
<td>0.18 µm GaSb</td>
</tr>
<tr>
<td>K (3573)</td>
<td>GaAs / 2.67 µm GaSb</td>
<td>250</td>
<td>490</td>
<td>8</td>
<td>none</td>
</tr>
</tbody>
</table>

**Table 1b:** Growth parameters for the second series of InSb on GaSb samples.

The trimethylantimony (TMSb) flow was kept constant at 100 cm$^3$ per minute (sccm), resulting in a $\text{III}/\text{V}$ ratio of 2.5 for samples H to K. Samples G1 and G2 were grown simultaneously with the GaAs and GaSb substrates in close proximity. * As a result of the gradual decrease in the output of the trimethylindium (TMIn) bubbler, the TMIn mole fraction in the gas phase for samples G1 and G2 was lower than that for 500
3. Results and Discussion

3.1. InSb on GaSb, sample series A to F

While the height of nm sized islands can be measured quite accurately by means of AFM, the base widths estimations are less accurate due to non-linear mixing of the sample topography with the tip geometry [13-14]. AFM examinations on sample F revealed the co-existence of three different types of three-dimensional InSb islands on GaSb as shown in Fig. 3 of ref. [9] and enabled accurate measurements of the islands' heights as well as estimations of their base widths. The key to our classification scheme was a proposal by Seifert et al. [15], according to which a barrier to the chemical potential surrounds each strained island. This enabled us to distinguish between fully strained and partly relaxed islands on the basis of AFM analyses alone [9-10]. Table 2 classifies the different island types in terms of their dimensionality and multiplicity and gives the numerical results of the AFM analysis for sample F, i.e. the uncapped counterpart of sample B which showed the highest PL intensity [9] from the samples of the series A to E.

Note that without any kinetic barrier to the growth of the islands, island arrays with a reasonably uniform size distribution, i.e. predecessors of QDs, could not exist due to Ostwald ripening. A recent high resolution TEM study and finite element simulation [16] confirmed the hypothesis by Seifert et al. [15], although it was not referred to in that paper.

We observed orientation contrast haloes in the vicinity of medium sized islands by means of TEM, Fig. 2 in ref. [10]. The general trend was that the larger the medium sized island was, the smaller the orientation contrast halo around them would be [10]. This observation is compatible to the hypothesis by Seifert et al. [15]. We will refer to capped objects as particles and to uncapped objects as islands. The term QDs will only be used for fully strained particles that have the capacity to luminesce.

<table>
<thead>
<tr>
<th>Island Dimensionality / Multiplicity</th>
<th>Measured heights (nm)</th>
<th>Estimated base widths (nm²)</th>
<th>Estimated number density (cm⁻²)</th>
<th>Elastic status</th>
</tr>
</thead>
<tbody>
<tr>
<td>2D / single</td>
<td>0.3 - 0.6</td>
<td>10 x 10</td>
<td>6 x 10⁹</td>
<td>not known</td>
</tr>
</tbody>
</table>
### Table 2: Result of AFM analyses for sample F.

<table>
<thead>
<tr>
<th>3D / single</th>
<th>≤ 4</th>
<th>≤ 30 x 30</th>
<th>3 × 10^9</th>
<th>fully strained</th>
</tr>
</thead>
<tbody>
<tr>
<td>3D / single</td>
<td>&lt; 10</td>
<td>≥ 50 x 100</td>
<td>6 × 10^9</td>
<td>partially relaxed</td>
</tr>
<tr>
<td>3D / double</td>
<td>≥ 10</td>
<td>≥ 75 x 200</td>
<td>4 × 10^9</td>
<td>partially relaxed</td>
</tr>
</tbody>
</table>

The strained state of the smallest three dimensional islands/particles of the series A to F was confirmed by conventional plan-view TEM, such as Figs. 1a,b where a typical “black-white” contrast (marker BW) is observed under two-beam diffraction contrast conditions for both islands [10] and particles. The direction vector of the black-white contrast was not uniformly oriented with respect to the diffraction vector. This implies that the island/particle shape is not uniform [10], i.e. that there is essentially no shape order. Some of the small particles did not show black-white contrast, indicating that although still strained, they might be highly alloyed. QDs occurred usually in isolation in samples of the series A to F, Fig. 1a, but occasionally they form conglomerates of two to three QDs, Fig. 1b, top-right corner.

Partially relaxed particles/islands (Table 2) were larger than fully strained QDs and showed moiré fringes in TEM diffraction contrast images. Structural defects such as threading dislocations and stacking faults were frequently associated with partially relaxed particles, Fig. 1b. These particles’ degree of relaxation was estimated to be typically about 80 % [8]. As the low magnification overview, Fig. 1c shows, the number density of the particles was variable on the area range that can more typically be assessed by one plan-view TEM sample, i.e. a few µm². Areas of the order of magnitude 100 µm² as accessible by AFM, were considered to be roughly representative for most samples.

PL spectra of samples such as E, B and C (2, 3, 4 s InSb deposition times) showed one peak with QD characteristics at around 1.7 µm with a full width at half maximum (FWHM) as low as about 15 meV, while sample D (1 s InSb deposition time) showed only a peak from the wetting layer (WL) at around 1.6 µm [9]. The QD emission characteristics were confirmed by means of magneto-PL, applying magnetic fields of up to 15 T [9]. The dimensions of luminescent QDs in sample C (4 s InSb deposition time) at a low excitation power were deduced from magneto-PL to be about 4 nm in height with 28 nm lateral diameter [9]. This dimension estimation of the QDs compares well with the AFM data (Table 2, line 2) from sample F (3 s InSb deposition time) and the TEM data from samples A (2 s InSb deposition time) and F, but gives more representative values since areas of about one mm² were probed by means of the PL equipment.
3.2. InSb on GaSb, sample series G1 to K

For the three capped samples of the series with the lower \( \frac{\text{III}}{\text{V}} \) ratios and growth rates (H, I and J), two peaks are observed in the PL spectra at typically 1.6 \( \mu \text{m} \) and 1.75 \( \mu \text{m} \). The full width at half maximum (FWHM) of the longer wavelength peaks were as low as about 10 meV [17-18].

Magneto-PL and temperature dependent PL were performed on samples I and J and led to an assignment of the longer wavelength peak to QDs, but did not permit a conclusive assignment of the shorter wavelength peak [17]. The size of the QDs, as deduced from magneto-PL [17-18], was similar to that obtained before on sample C [9].

With regard to the morphology, we observed by means of AFM, Figs. 2a,b and 3 that the uncapped samples of the lower \( \frac{\text{III}}{\text{V}} \) ratios and growth rates series (G1, G2, and K) showed “dumbbell” shaped islands. When the specimens were rotated 90° in the AFM, the dumbbells in the images rotated through 90°, showing that the shapes of the islands were not an AFM-tip artefact. While Fig. 2a was taken with the AFM-tip scanning direction parallel to the longer axis of the dumbbell shaped islands, Figs. 2b and 3 were taken with the AFM-tip scanning in the direction perpendicular to this axis.

In plan-view diffraction contrast TEM, medium sized and larger dumbbell shaped islands were revealed to be partially relaxed as indicated by the occurrence of moiré fringes, Figs. 4a,b. The edge of the regions containing the moiré fringes also outline the shape of the islands as being bounded by facets. Thus, the dumbbell shape of the islands as derived from AFM is confirmed by the independent TEM method.

The presence of larger and smaller dumbbell shaped islands on samples G2 and G1, Figs. 2a,b; 4a,b, (Table 3a,b) suggests that these islands did not form as a result of the coalescence of two faceted islands. While “chain like” and “symmetrically shaped” morphologies of InAs QDs clusters in GaAs have previously been observed by means of plan-view TEM [19-20], we are not aware that similarly shaped dumbbell islands have been observed in any other material system.

Tables 3a-c classify the different island types for samples G1, G2 and K in terms of their shapes and gives the numerical results of the AFM analyses. Since only a standard tip with a nominal radius of 20 nm was used for these experiments, the AFM resolution was inferior to the experiments that are represented in Table 2.
Consequently, two-dimensional islands have not been resolved for either of these samples. The shapes of the smallest islands could for the same reason not be resolved in all three samples. While sample G2, which was grown on a GaSb substrate, possessed a smooth surface, Fig. 2a, there were pits of up to an order of magnitude wider than the dumbbell shaped islands in samples G1, Fig. 2b, and K, Fig. 3, which were grown on GaAs substrates. In comparison to samples G2 and G1, the size of the dumbbell shaped islands is more homogenous in sample K, probably as a result of the lower $\frac{\text{III}}{\text{V}}$ ratio and growth rate.

<table>
<thead>
<tr>
<th>Island shape</th>
<th>Measured heights (nm)</th>
<th>Estimated base widths ($\text{nm}^2$)</th>
<th>Estimated number density ($\text{cm}^{-2}$)</th>
<th>Elastic status</th>
</tr>
</thead>
<tbody>
<tr>
<td>not resolved</td>
<td>$\leq 5$</td>
<td>$\leq 40 \times 60$</td>
<td>$1\cdot 2 \times 10^7$</td>
<td>fully strained</td>
</tr>
<tr>
<td>small dumbbell</td>
<td>5 - 8</td>
<td>$\geq 50 \times 100$</td>
<td>$8 \times 10^7$</td>
<td>partially relaxed</td>
</tr>
<tr>
<td>large dumbbell</td>
<td>10 - 15</td>
<td>$\geq 130 \times 240$</td>
<td>$6 \times 10^8$</td>
<td>partially relaxed</td>
</tr>
<tr>
<td>double-dumbbell</td>
<td>$\geq 15$</td>
<td>$\geq 140 \times 480$</td>
<td>$1 \times 10^7$</td>
<td>partially relaxed</td>
</tr>
</tbody>
</table>

Table 3a: Result of AFM analyses for samples G1.

<table>
<thead>
<tr>
<th>Island shape</th>
<th>Measured heights (nm)</th>
<th>Estimated base widths ($\text{nm}^2$)</th>
<th>Estimated number density ($\text{cm}^{-2}$)</th>
<th>Elastic status</th>
</tr>
</thead>
<tbody>
<tr>
<td>not resolved</td>
<td>$\leq 4.5$</td>
<td>$\leq 40 \times 70$</td>
<td>$1\cdot 2 \times 10^7$</td>
<td>fully strained</td>
</tr>
<tr>
<td>small dumbbell</td>
<td>5 - 8</td>
<td>$\geq 60 \times 130$</td>
<td>$8 \times 10^7$</td>
<td>partially relaxed</td>
</tr>
<tr>
<td>large dumbbell</td>
<td>10 - 15</td>
<td>$\geq 120 \times 240$</td>
<td>$6 \times 10^8$</td>
<td>partially relaxed</td>
</tr>
<tr>
<td>double-dumbbell</td>
<td>$\geq 15$</td>
<td>$\geq 150 \times 500$</td>
<td>$1 \times 10^7$</td>
<td>partially relaxed</td>
</tr>
</tbody>
</table>

Table 3b: Result of AFM analyses for samples G2.

<table>
<thead>
<tr>
<th>Island shape</th>
<th>Measured heights (nm)</th>
<th>Estimated base widths ($\text{nm}^2$)</th>
<th>Estimated number density ($\text{cm}^{-2}$)</th>
<th>Elastic status</th>
</tr>
</thead>
<tbody>
<tr>
<td>not resolved</td>
<td>$\leq 3.5$</td>
<td>$\leq 30 \times 40$</td>
<td>$1\cdot 2 \times 10^7$</td>
<td>fully strained</td>
</tr>
<tr>
<td>large dumbbell</td>
<td>$\leq 15$</td>
<td>$\geq 90 \times 350$</td>
<td>$2 \times 10^8$</td>
<td>partially relaxed</td>
</tr>
</tbody>
</table>

Table 3c: Result of AFM analyses for samples K.

As the AFM and TEM analyses of samples G1 and G2 showed, there seems to be no significant influence of the choice of substrates (i.e. either GaSb or GaAs) underneath the buffer layer on the structural and morphological properties of InSb islands. As Tables 3a,b show, the island number densities, sizes and shapes were similar. This agrees well with the observation that the surface roughness is not an important parameter.
in the deposition of QDs [21], [10].

The dumbbell shaped islands of the lower \( \text{III/V} \) ratios and growth rates series showed a higher level of order than sample F, for which there was essentially no shape order (as derived from TEM and described above). While we observed for samples G2 and G1, Fig. 2a,b, only shape order, sample K, Fig. 3 showed some size order in addition to shape order. There were even first indications of some linear alignment order for sample K.

The fact that we observed the occurrence of faceted islands in conjunction with signs of self-ordering of the islands suggests that facet induced self-limiting growth of fully strained and partially relaxed islands [22-23] may be important for self-ordering processes. For the growth of islands with facets, two-dimensional seeds are required that preferentially nucleate at the edges between the WL and the island facets [22-23], i.e. at strained areas.

There is always an energetic barrier to the nucleation of two-dimensional seeds. The strain, however, provides an additional kinetic barrier to the nucleation of two-dimensional seeds which is for the same strain level larger for larger islands, resulting in more uniformity of the island sizes [22-23]. This self-limiting growth argument of faceted islands is valid even for low strain levels [22-23], such as the bigger and partially relaxed islands may possess. It is, thus, justified to infer self-ordering for the smaller and fully strained islands from the observation of self-ordering for the bigger and partially relaxed islands, which is experimentally less demanding to observe.

### 3.3. GaSb on GaAs, sample X

As Fig. 5 indicates, self-ordering processes seem to be more readily achievable in the GaSb on GaAs system. For sample X, there are large “pyramid/cone like” \((2 \times 10^8 \text{ cm}^{-2})\) islands and many more small “trapezoidal-prism like” islands \((4 \times 10^9 \text{ cm}^{-2})\). All of these islands are probably relaxed to some extent. There are also still smaller islands \(10^7 \text{ cm}^{-2}\) that may be fully strained and could luminesce when excited. GaSb QDs in a GaAs matrix photo-luminesce in the near infrared [24].
3.4. Self-ordering for InSb on GaSb and GaSb on GaAs

In the following we will make use of the phenomenological classification scheme of the hierarchy of QD self-ordering mechanisms by Bimberg et al. [1] in a slightly modified form. We are aware that this classification scheme is only qualitative and that the boundaries between the hierarchy levels are not well defined. We are not aware of any alternative classification scheme of self-ordering in self-assembled QDs systems.

According to ref. [1] there are four levels to the hierarchy of self-ordering mechanisms: 1st - orientation order, 2nd - shape order, 3rd - size order, and finally 4th - alignment order. Any higher level of hierarchy includes the feature(s) of the lower level(s). At the highest level, i.e. alignment order, we distinguish between one and two dimensional alignment order. While we observed indications of linear (i.e. one dimensional) alignment in some of our samples, two dimensional alignment as an array was (so far) not observed in the Sb-based material systems we investigated. It is, however, well documented in other materials systems such as InAs on GaAs [1], [3], [15], and SiGe on Si [25].

The InSb islands on GaSb of the uncapped sample F that represents the sample series with the higher $\text{III}/\text{V}$ ratio and growth rate (A to F) possessed only orientation order that may be due to preferential island nucleation at terrace steps, Fig. 3 in ref. [9], as reported in the literature several years ago for early work on InAs islands on GaAs [26]. There is, however, orientation, shape, size, and some linear-alignment order for the uncapped sample K, Fig. 3, that represents the samples with the lower $\text{III}/\text{V}$ ratio and growth rate (H to K). The samples G2 and G1, Figs. 2a,b, show only orientation and shape order, and their intermediate status relates well to their intermediate growth conditions, as given in Tables 1a,b. The higher level of self-organisation of sample series H to K may also be the explanation for the smaller FWHM of the QDs PL peaks as mentioned in the previous section.

Although non-linear mixing of the AFM-tip geometry and the sample topography can lead to changes in the apparent surface periodicity, the presence of peaks in a power spectrum is known to be a good indication of surface periodicities [14]. We used, therefore, the power spectrum of AFM line scans at various sample locations as a figure of merit for the occurrence of alignment order of the islands.

Power spectra of AFM line scans perpendicular to the long axis of the trapezoidal-prism like shaped GaSb
islands on GaAs (sample X) showed peaks of up to 20 dB (Fig. 6a). AFM line scans parallel to this axis, however, did not show such peaks in the power spectra (Fig. 6b). We, thus, conclude that GaSb islands on GaAs possess linear alignment order. Since this is a feature of the 4th level of hierarchy, there should according to the phenomenological model [1] also be orientation, size, and shape order. Since a phenomenological model with not well defined boundaries has been used, one should not be surprised that there may be different interpretations of AFM images such as Fig. 5 which we consider as showing these features to some extent.

The island number densities for our InSb on GaSb and GaSb on GaAs systems are about two to three orders of magnitude lower than the ones typically observed for InAs on GaAs [1], [3], [15] and SiGe on Si [25]. As demonstrated experimentally in ref. [25] and theoretically based on elastic interactions of the islands via the underlying substrate/buffer-layer/WL combination [3], [27], a high island number density seems to be required for self-ordering processes of the highest level of hierarchy to occur.

4. Conclusions

Two series of InSb QDs in a GaSb matrix have been grown by MOVPE at 490 ± 10 °C and luminesce in the mid-infrared at about 1.7 µm. AFM and TEM analyses showed that fully strained and partially relaxed islands/particles co-exist. Lower \( \text{III/}V \) ratios and growth rates led to dumbbell shaped islands with distinct facets as observed by TEM. The facetting of the dumbbell shaped islands may trigger self-ordering processes. We expect that still lower growth rates will lead to higher QD number densities, a higher degree of self-organisation and hence higher and narrower QD PL peaks. Further optimisation of the growth conditions will be needed to investigate this.
5. References


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from InAs Quantum Dots in a GaAs matrix”, Proc. 24th Intern. Conf. on Physics of Semiconductors, Jerusalem, 1998, in press


Fig. 1a: TEM 220 dark-field diffraction contrast image close to [001] zone axis; InSb particles in GaSb matrix, sample A. QDs show “black-white” contrasts (marker BW).
**Fig. 1b:** TEM 220 dark-field diffraction contrast image close to [001] zone axis; InSb particles in GaSb matrix, sample A. QDs in the top-right corner show black-white contrasts and relaxed particles are frequently associated with defects (marker D). The moiré fringes inside the larger particles were lost by the photographic reproduction.
**Fig. 1c:** TEM 220 dark-field diffraction contrast image close to [001] zone axis; InSb particles in GaSb matrix, sample A, lower magnification overview, showing about 200 $\mu$m$^2$. 
Fig. 2a: Plan-view AFM image of InSb islands on GaSb buffer layer on GaSb substrate, sample G2.

The third dimension of the image is represented by 256 grey levels that cover the whole range from black to white within 50 nm. The AFM tip was scanned parallel to the [110] direction and the long axis of the dumbbell shaped islands were parallel to this direction. A plan-view representation was chosen for this figure in order to facilitate easy comparison with Fig. 2b, on the one hand, and the TEM plan views of sample G1, Fig. 4a,b, on the other hand.
**Fig. 2b:** Plan-view AFM image of InSb islands on GaSb buffer layer on GaAs substrate, sample G1.

The third dimension of the image is represented by 256 grey levels that cover the whole range from black to white within 100 nm. The AFM tip was scanned parallel to the [-110] direction and the short axis of the dumbbell shaped islands were roughly parallel to this direction. A plan-view representation was chosen for this figure in order to facilitate easy comparison with Fig. 2a, on the one hand, and the plan-view TEM images from the same sample, Fig. 4a,b, on the other hand.
Fig. 3: 3D AFM image of InSb islands on GaSb, sample K.

The AFM tip was scanned parallel to the [-110] direction and the long axis of the dumbbell shaped islands was perpendicular to this direction. A 3-D representation clearly reveals the dumbbell shape of the islands. Smaller islands that are probably fully strained are revealed as well by their shadow images (marker S).
**Fig. 4a:** TEM 220 bright-field diffraction contrast image close to [001] zone axis of a medium sized InSb island on GaSb, sample G1. The edges of this image are parallel to <110>. The straight edges of the regions containing the moiré fringes indicate that the island is bounded by facets.

**Fig. 4b:** TEM 220 bright-field diffraction contrast image close to [001] zone axis of a large InSb island on GaSb, sample G1. The edges of this image are oriented parallel to <110>. 
Fig. 4c: Diffraction pattern to Fig. 4a,b.
1 x 1 \( \mu m^2 \) scan area, AFM tip scanning direction \([110]\)

**Fig. 5:** 3D AFM image of GaSb islands on GaAs, sample X. The AFM tip was scanned parallel to the \([110]\) direction and the long axis of the trapezoidal-prism like island were parallel to this direction.
**Fig. 6a:** Height diagram (height in Å versus distance in µm) and power spectrum (relative frequency in dB versus reciprocal distance in µm⁻¹) of a 3.1 µm long AFM scan across GaSb on GaAs, sample X, perpendicular to the long axis of the trapezoidal-prism like island. The spectrum is noisy but representative of several taken indicating the reproducibility of the phenomena responsible.

**Fig. 6b:** Height diagram (height in Å versus distance in µm) and power spectrum (relative frequency in dB versus reciprocal distance in µm⁻¹) of a 3.06 µm long AFM scan across GaSb on GaAs, sample X, parallel to the long axis of the trapezoidal-prism like island.