Growth Nature of InSb Channel Layer on Heteroepitaxial Films of InGaSb Layer on GaSb/Si(111)-√3×√3-Ga Surface Phase

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The single crystalline InSb and GaSb thin films on the same Si(111) substrate is an essence of CMOS device fabrication. In this study, the growth of InSb has been observed on InGaSb intermediate layer grown on high quality (HQ) GaSb buffer layer of single crystalline nature. In this case, In$_{0.45}$Ga$_{0.55}$Sb was initially grown on HQ GaSb/Si(111)-√3×√3-Ga surface following a two-step growth method. The controlled and precise flux ratio of Ga and Sb enabled HQ GaSb layer to grow without twins. For In$_{0.45}$Ga$_{0.55}$Sb intermediate layer, In ratio x ranged from 0.90 to 0.75 for samples S1–S4 with a step 0.05. To analyze the growth nature and surface morphology, reflection high-energy electron diffraction (RHEED), scanning electron microscopy (SEM), x-ray diffraction (XRD) have been performed and studied. It is concluded that InSb epitaxial surfaces have shown better smoothness and fewer defects towards Gallium increase. The growth nature in this case for both InSb epitaxial layer and InGaSb intermediate layer were coherent where HQ GaSb buffer layer was effective in decreasing twins. Some complementary samples were prepared to support the assumptions made before starting the experiments. Finally, the possibility of single crystalline InSb and GaSb thin films on the same Si(111) substrate has been demonstrated.

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I. INTRODUCTION

The semiconductor world is continuously in search of low-power, high-performance CMOS devices of III-V materials. Through heavy research on III-V materials for the betterment of CMOS devices, different aspects have been disclosed. Among those issues, minimizing the leakage currents have challenged the researchers most. To overcome this, wider bandgap materials are there. It has already been reported that HQ buffer layer has great impact in growing other thin film layers on it. Such as HQ GaSb buffer layer have a positive impact in growing smoother, defectless and single crystalline InGaSb intermediate or barrier layer [1–5]. Si being cheaper substrate material and group III-V semiconductors having attractive properties, the better growth III-V films are expected on Si. Complex III-V multilayer grown directly on Si substrates have been demonstrated for n-FETs in the last decades [6]. Large lattice mismatch with low defect densities has also been reported for III-V growth on Si [2,7]. A directional mismatch is observed when a polar III-V material is deposited on a non-polar material like Si resulting in defects. Those defects affect material quality, electrical properties and device performance when III-V thin film layers are grown on silicon. Anti-phase domains are another after effect. This is why reduction in dislocation densities is necessary which in turn reduces anti-phase domains and provides high mobility for group III-V layers. These factors have inspired us to grow a buffer layer of GaSb on Si to provide a good base for other layers to grow. This effort was proven through better III-Antimonides growth on silicon substrate with the parallel factor of large lattice mismatch [8]. Formation of InSb quantum dots has been investigated on GaSb substrates [9–12]. Apart from lucrative III-V properties III-Antimonides have superior electron and hole mobilities [13,14]. High-performance Antimonide p-MOSFETs with peak hole mobility more than 900 cm$^2$/Vs with strained InGaSb p-MOSFETs have been reported in recent years [6,15,16]. The hole mobility around 1,500 cm$^2$/Vs in InGaSb [17], more than 1,300 cm$^2$/Vs in GaSb [18], and over 1,200 cm$^2$/Vs in InSb [19] has also been reported. InGaSb p-FETs with effective hole mobility of 1,230 cm$^2$/Vs has also been reported [20] despite the difficulty in growing high quality InGaSb crystals for wide range of indium ratio [21,22]. III-Antimonides have been widely accepted as a channel material due to its superior responses toward ultrafast and very low power devices of CMOS technology [3]. Al$_2$O$_3$/InSb/Si(111) nMOSFETs have been descriptively reported by our group [23] and has shown several growth methods and conditions to overcome the lattice mismatch issue of InSb precisely 19.3% and GaSb precisely 12.3% with Si through MBE growth technique describing their superior electronic properties, grown surfaces and patterns on Si(111) [24–27]. Heteroepitaxial growth of GaSb on Si(111)-√3×√3-Ga with single crystalline nature has been reported by our group with the possibility of narrow Full Width at Half Maximum (FWHM) [28]. XRD scan has shown the possibility of better epitaxial layer as well. The large lattice mismatch between Si(111) and GaSb causes Volmer-Weber growth pattern to form GaSb islands. To reduce this difficulty, Si(111)-√3×√3-Ga template has been proved effective [29]. Si(111) substrate has superior property as it avoids anti-phase domains for III-V semiconductors and inter-facial misfit dislocations have the ability to reduce strain for heteroepitaxial films[30]. High performance III-V MOSFETs have some large barriers like interface state density between Al$_2$O$_3$/InSb interface, leakage current to Si substrate and high resistance for source to drain. To overcome the leakage issue a III-V composition of InGaSb was assumed to be a special factor having peak high hole mobility of nearly 900–1200 cm$^2$/Vs introducing a wider bandgap. The assumption has been*

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positively proved recently with In$_x$Ga$_{1-x}$Sb growth on Si(111)-$\sqrt{3}\times\sqrt{3}$-Ga on GaSb buffer layer through two-step heteroepitaxial growth [1]. This report focuses on the advancement of the InSb channel layer’s growth on InGaSb layer.

Figure 1 shows the schematic diagram of growth layers of InSb on InGaSb on HQ GaSb buffer layer with Si(111)-$\sqrt{3}\times\sqrt{3}$-Ga template. All the new samples were analyzed to observe the growth nature of the InSb layer on the intermediate InGaSb layer. The primary focus remains on reducing substrate and interface leakage currents with GaSb layer having a wider bandgap. To investigate the coherency of InSb channel layer growth on InGaSb/GaSb compared to the InGaSb intermediate layer growth on HQ GaSb [1], new samples of some specifications were prepared. The effectiveness of HQ GaSb buffer layer in growing single crystalline nature GaSb, InGaSb epitaxial layers with narrow FWHM is found similar to the previous report [1,28]. GaSb has been deployed to remove the substrate leakage current by its wider bandgap, superior electrical properties with better surface and single crystalline thin film. The HQ GaSb layer acts as a barrier layer as its wider bandgap allows it to create a quantum well barrier between GaSb and Si substrate to eliminate substrate leakage current. On the other hand, InGaSb acts as an intermediate layer between HQ GaSb and InSb channel layer to provide a wider bandgap for both the adjacent layers. This report describes the growth nature of InSb thin film on the In$_x$Ga$_{1-x}$Sb thin film grown on HQ GaSb.

II. EXPERIMENTAL

The OMICRON molecular beam epitaxy (MBE) chamber supported by a base pressure of about $2 \times 10^{-8}$ Pa with an attached RHEED unit was used for the deposition process. Substrates with the dimension of $15\times5\times0.6$ mm$^3$ were cut from mirror polished p-type Si(111) wafer having resistivity of about 20 $\Omega$cm. To prepare clean Si(111) surface, flash annealing at 950°C for 10 min was performed after degassing at 600°C for 12 hours. Finally, flash annealing at 1250°C and cooling in the chamber took place. A clean (7 $\times$ 7) surface was then confirmed through RHEED system. Knudsen cells were incorporated to evaporate elements of high purity In, Ga and Sb. To monitor the substrate temperature throughout the deposition process, an infrared pyrometer was used. Initially Si(111)-$\sqrt{3}\times\sqrt{3}$-Ga surface phase was obtained through deposition of 0.33 ML (which took around 20 seconds) Ga on the cleaned Si(111)-7 $\times$ 7 reconstructed surface at the substrate temperature of 560°C. Substrate temperature was operated through electric heating for all the epitaxial layers except the HQ GaSb buffer layer. To grow HQ GaSb buffer layer, substrate temperature was operated under heater heating for low temperature growth. Ga and Sb cell temperatures were set to deposit 1 molecular-layer (MoLa)/min for the buffer layer. Shutters of Ga and Sb were switched on when the substrate temperature reached 260°C under heater heating to start deposition. To grow a twinless HQ GaSb buffer layer, the flux ratio of one for Sb/Ga was determined through trial and error basis. The Ga and Sb cell temperatures were fixed at around 760°C and 320°C respectively. During the initial growth of HQ GaSb buffer layer, RHEED images have showed asymmetric patterns resembling single crystal growth. The GaSb buffer layer was grown for 30 minutes and the thickness was around 18nm in each case. The substrate temperature was increased from 260–420°C within the first 10 minutes. Also to get sharp streaks and maintain the flux ratio, the Sb cell temperature was increased by 20°C at an increase of 5°C/5min within first 18 $\sim$ 20 minutes. The substrate temperature was immediately reduced to room temperature after the HQ GaSb deposition. The asymmetric RHEED pattern, ensures the growth of high quality GaSb thin film without island pattern showing bright streaks [28]. Electrical heating for 1 min at 1.5 Ampere was performed to anneal the grown GaSb surface. For the intermediate InGaSb layer deposition, substrate temperature was set to 380 and increased up to 430°C at an increasing rate of 1°C/min, where the flux ratio of GaSb remained constant. The flux ratio was estimated through RHEED observation for the time of surface reconstruction from 7 $\times$ 7 to $\sqrt{3}\times\sqrt{3}$ for Ga and In. To prepare all the samples, the range of cell temperature was varied for In–746–754°C and Ga–825–860°C. The flux rate for Sb was fixed at temperature of 475°C. In the case of the sample S8, the 40 nm InSb thin film was directly deposited on the Si(111)-7 $\times$ 7 with the symmetric RHEED pattern. The detail of the prepared major samples (S1–S4) and supporting samples (S5–S8) are listed in Table I. The thickness of the InSb thin film layer for S1–S4 was 40 nm. This may have caused that the $\varphi$-scan was not possible for the very thin InSb layer. Sample S5 was prepared with 800 nm-thick InSb layer to observe the $\varphi$-scan pattern for the InSb layer and also to know the surface morphology of a relatively thicker layer. To be sure of the S1–S4 growth nature, sample S6 was prepared with 400 nm In$_{0.46}$ Ga$_{0.54}$ Sb layer. As in this report we are reporting about the single crystalline nature of the InSb channel layer with fewer defect densities, a sample of 90 nm GaSb thin film–S7 along with 400 nm In$_{0.46}$ Ga$_{0.54}$ Sb thin film–S6 and 800 nm InSb thin film–S5 were prepared to check the crystalline nature of GaSb, In$_x$Ga$_{1-x}$Sb and InSb thin films. 40 nm-thick InSb sample–S8 was directly grown on Si(111)-7 $\times$ 7 with In–Sb bi-layer to compare the surface defects of all the InSb epitaxial layers on Si(111)-7 $\times$ 7, HQ GaSb-Si(111)-$\sqrt{3}\times\sqrt{3}$-Ga and on In$_x$Ga$_{1-x}$Sb-HQ.
GaSb-Si(111)-\(\sqrt{3} \times \sqrt{3}\)-Ga. Though \(\phi\)-scan was not possible for the very thin 40 nm InSb film of S8, our experiences of growing InSb with the In, Sb bi-layer suggests that the InSb film of S8 is 30° rotated film on Si(111).

The composition of In to Ga for S1–S4 was 0.9:0.1 to 0.75:0.25 at a step of 0.05 to compare the advancement of the previously investigated epitaxial surface [1]. The samples were primarily observed through the RHEED pattern operating at 15 kV. To operate the XRD \(\phi\)-scanning, initially the 2\(\theta\) value (with reference to Si (111) substrate) is detected for the target thin film with XRD 2\(\theta/\omega\) scan. Then to match the angular setting for (111) peaks of each target films, the \(\chi\) value is set to -69°, 2\(\theta\) value was set to the obtained value, \(\omega\) as the half of 2\(\theta\) value and \(\phi\) at 0°. Then the in-plane orientation was examined through XRD \(\phi\)-scan. This process ensures that only the (111) peaks of the desired thin film appear.

III. RESULTS AND DISCUSSIONS

To analyze the growth nature and surface morphology of the grown samples, a series of analysis was performed through RHEED, SEM, FESEM and XRD. The XRD characterization was performed using Cu K\textsubscript{α1} radiation.

Figure 2 shows the RHEED images of all four samples (S1–S4) after the 40 nm InSb thin film deposition. It was observed that the streaks are clear bright 2x streak pattern for InSb epitaxial layer which is coherent to the clear bright streaks and grows with similar properties as the intermediate layer of InGaSb grows on HQ GaSb buffer layer observed in Fig. 2(e-h) of the reference [1]. These indicate the growth nature of InSb film on InGaSb intermediate layer.

To analyze the surface morphology, the samples S1–S4 were examined with SEM. Figure 3 is summarized with the surface images where it is clearly visible that the defect densities reduces as the gallium contents are increased providing smoother surfaces. It is inferred that the InSb films’ growth is coherent to the film surfaces of InGaSb grown on GaSb/Si(111)-\(\sqrt{3} \times \sqrt{3}\)-Ga [1]. It is found through SEM images for the sample S3 with the Gallium ratio 0.20, In\textsubscript{x}Ga\textsubscript{1-x}Sb grow the best epitaxial surface than the other compositions. The same was reported earlier for In\textsubscript{x}Ga\textsubscript{1-x}Sb thin film growth on HQ GaSb thin film [1].

The main objective is to have better epitaxial growth of InSb channel layer of a smooth surface, fewer defects with twinless crystal. Hence, SEM images of the samples’ surfaces (S1–S4 for figure 3 a–d) were analyzed. To compare the growth of the InSb, complementary samples S5, S6 and S8 was analyzed with FESEM. Figure 4a shows that the InSb thin film directly grown on Si(111)-7 \times 7 reconstructed surface with In, Sb bi-layer provides a surface with defects though the large lattice mismatch of about 19.3% between InSb and Si was reduced to 3.3% [27]. Figure 4b shows that the In\textsubscript{0.76}Ga\textsubscript{0.24}Sb thin film has a better surface when grown on HQ GaSb buffer layer with fewer defects. Figure 4c shows a smoother surface where 800 nm InSb thin film layer was grown on the In\textsubscript{0.90}Ga\textsubscript{0.10}Sb intermediate layer on the HQ GaSb buffer layer. It was observed that the defects in the surfaces of the InSb were reduced, and smoothness increased when grown on In-GaSb intermediate layer. It was also noticeable that the smoothness and reduction of defects for the InSb thin film is coherent to the base surface of InGaSb film which is dependent on the Gallium content shown in Fig. 3. This observation is very similar to the surface grown for InGaSb on Si(111) shown in Fig. 3(a–d) [1]. The InSb thin film requires better growth condition to comply with the twin eliminated crystal quality as reported by our group for InSb growth on Si(111) [25–28]. It also gives the evidence that an increase in Gallium contents for In\textsubscript{x}Ga\textsubscript{1-x}Sb is helpful in growing single crystalline epitaxial layer proving that the HQ GaSb buffer layer is effective for all the successive epitaxial layers in getting single crystal growth.

XRD 2\(\theta/\omega\) scan has been performed to characterize the grown crystals. According to the specification shown in Table I, the thickness of the HQ GaSb is 18nm and InSb is 40 nm. These may have caused that the XRD 2\(\theta/\omega\) scan was identifiable only for In\textsubscript{x}Ga\textsubscript{1-x}Sb thin films due to the obtained value, \(\omega\) as the half of 2\(\theta\) value and \(\phi\) at 0°. Then the in-plane orientation was examined through XRD \(\phi\)-scan. This process ensures that only the (111) peaks of the desired thin film appear.

FIG. 2. RHEED pattern evaluation for InSb epitaxial layers showing well ordered 2x bright streaks for the samples a) S1, b) S2, c) S3, and d) S4.

FIG. 3. SEM images of the InSb epitaxial layers showing a reduction in defects as Ga contents are increased from 10% to 25%. Here, the images are for the samples a) S1, b) S2, c) S3 and d) S4.
TABLE I. Detail of prepared samples.

| Sample | Thickness (nm) | Deposition Time (min) | Composition | InGaSb intermediate layer Thickness (nm) | Deposition Time (min) | InSb Channel layer Thickness (nm) | Deposition Time (min) |
|--------|----------------|-----------------------|-------------|------------------------------------------|-----------------------|-----------------------------------|-----------------------|
| S1     | 18             | 30                    | In$_{0.90}$Ga$_{0.10}$Sb | 460                      | 120                   | 40                               | 10                    |
| S2     | 18             | 30                    | In$_{0.85}$Ga$_{0.15}$Sb | 460                      | 120                   | 40                               | 10                    |
| S3     | 18             | 30                    | In$_{0.80}$Ga$_{0.20}$Sb | 500                      | 120                   | 40                               | 10                    |
| S4     | 18             | 30                    | In$_{0.75}$Ga$_{0.25}$Sb | 490                      | 120                   | 40                               | 10                    |
| S5     | 18             | 30                    | In$_{0.90}$Ga$_{0.10}$Sb | 400                      | 120                   | 800                              | 210                   |
| S6     | 18             | 30                    | In$_{0.75}$Ga$_{0.25}$Sb | 400                      | 120                   | 40                               | 10                    |
| S7     | 90             | 100                   | –            | –                                       | –                     | –                                | –                     |
| S8     | –              | –                     | –            | –                                       | –                     | 40                               | 100                   |

FIG. 4. FESEM images of a)InSb 40 nm (S8) thin film on Si(111)-7 $\times$ 7, b)InGaSb 400 nm (S6) thin film on HQ GaSb/Si(111)-$\sqrt{3} \times \sqrt{3}$-Ga, c)InSb 800 nm (S5) thin film on InGaSb-HQGaSb-Si(111)-$\sqrt{3} \times \sqrt{3}$-Ga.

to more thickness ($\geq$ 400 nm). As shown in Fig. 5(a–d) it is clearly observed that the In$_x$Ga$_{1-x}$Sb peaks shift towards right indicating increase in diffraction angle to the increase in Gallium contents. Figure 6 provides the quantitative measures through line diagram. Though the peaks are not sharp, the films may be improved by increasing the thickness of GaSb buffer layer. Si(111) peaks are visible at around the standard angle of 28.4419°. The $2\theta/\omega$ patterns’ range is plotted for 20°–32°, showing InGaSb(111)-related peaks and the sharp peaks for Si(111). InSb layer peaks also seem to present in highly concentrated shoulder peaks (around 23.7°) to the left of InGaSb peaks.

Figure 7 summarizes the XRD $2\theta/\omega$ patterns of the supporting samples S5–S7, diffracted from (111) peaks of a 90 nm GaSb ($2\theta/\omega=25.3372^\circ$) thin film (sample S7 shown in Fig. 7a), 400 nm InGaSb ($2\theta/\omega=24.0980^\circ$) thin film (sample S6 shown in Fig. 7b) and an 800 nm InSb ($2\theta/\omega=23.7805^\circ$) thin film (sample S5 shown in Fig. 7c).

The FWHM of the samples S1–S4 were calculated (ranging 0.1526–0.1647° or 549.36–592.92 arcsec) and compared with the supporting samples S5, S6 and S7 (0.1124, 0.1257 and 0.1315° or 404.64, 452.52, 473.40 arcsec respectively). These resemble that the $2\theta/\omega$ scans diffracted from (111) peaks of the GaSb (for S7), InGaSb (for S6) and InSb (for S5) thin films show that the sharp peaks indicating good crystal growth. Also the diffraction angle $2\theta/\omega$ decreases as the GaSb is replaced with InGaSb and InSb shown in Fig. 7a–c respectively proving that the epitaxial layers were grown in a desired manner.

The in-plane orientation of the supporting samples (S5–S7) was measured through XRD $\phi$-scan patterns. Figure 8

http://www.sssj.org/ejssnt (J-Stage: http://www.jstage.jst.go.jp/browse/ejssnt/) 23
TABLE II. Lattice mismatch of the InGaSb intermediate layer with adjacent layers.

| CS Material        | Lattice Constant in Ångstrom (Å) | Lattice Mismatch with GaSb (%) | Lattice Mismatch with InSb (%) |
|--------------------|----------------------------------|-------------------------------|------------------------------|
| Si                 | 5.431                            |                               |                              |
| InSb               | 6.479                            |                               |                              |
| GaSb               | 6.096                            |                               |                              |
| S1=In_{0.90}Ga_{0.10}Sb | 6.442                   | 5.67                          | 0.58                         |
| S2=In_{0.85}Ga_{0.15}Sb | 6.422                   | 5.35                          | 0.88                         |
| S3=In_{0.80}Ga_{0.20}Sb | 6.403                   | 5.04                          | 1.19                         |
| S4=In_{0.75}Ga_{0.25}Sb | 6.383                   | 4.72                          | 1.49                         |

FIG. 6. The XRD 2θ/ω-scan peaks shift to a higher angle to the increase in Ga contents for the In_{x}Ga_{1−x}Sb thin films.  

The in-plane orientation of the grown samples (S1–S4) for InSb epitaxial layers on In_{x}Ga_{1−x}Sb intermediate layers. The ϕ-scan patterns of (111) peaks' reflection of In_{x}Ga_{1−x}Sb films instead of InSb, as only (111) peaks of InGaSb could be detected due to its higher thickness compared to 18 nm GaSb and 40 nm InSb thin films. The solid circles represent the direction of the peak position of Si(111) substrates. Due to the angular constraint of the machine, XRD ϕ-scan peaks at scan angle around 180° could not be shown. The vertical axis being log scale, the peak intensity around ±60° is very small. The scan pattern shows the full twin elimination (for Fig. 8a) or twin eliminating tendency for the epitaxy grown for the InSb thin film which is shown and described as proof in Fig. 9 with supporting samples S5–S7. The absence of twin peaks around ±60° for Fig. 8a represent that the low temperature growth of HQ GaSb buffer layer was effective in eliminating twins and the regulation of the growth condition was as expected. This can be an indication for the possibility of twin elimination for Fig. 8(b–d) with further precise control over parameters’ fluctuations.

It has already been justified that the HQ GaSb buffer layer is effective in growing In_{x}Ga_{1−x}Sb intermediate layers as the crystals show full or partial twin elimination tendency to grow single crystal shown in Fig. 6(a–d) of...
the Si(111)-φ-scan pattern for a 90 nm HQ GaSb thin film (S7) on complementary samples (S5–S7). Figure 9a shows the above assumption, films representing the InSb thin film quality. To advocate intermediate layers. In this report also, we have found thin films with twin eliminating crystalline nature for InGaSb films representing the InSb thin film quality. To advocate the above assumption, φ-scan was performed for some complementary samples (S5–S7). Figure 9a shows the φ-scan pattern for a 90 nm HQ GaSb thin film (S7) on the Si(111)-√3 × √3-Ga surface phase. In this case, the peaks diffracted from GaSb(111) peaks along the Si(111) peaks represent a single crystalline film for GaSb thin film without rotation. Solid circles in the figure indicate peak position of Si(111) peaks. The φ-scan patterns of the epitaxial films diffracted from InGaSb(111) peaks of 400 nm grown on HQ GaSb(111)-√3 × √3-Ga (S6) and InSb of 800 nm grown on InGaSb-HQ GaSb-Si(111)-√3 × √3-Ga (S5) are shown in Fig. 9b & c respectively. Here, the single crystalline nature is also observed which proves that the HQ GaSb buffer layer is really effective in growing further twinless epitaxial layers.

Table II has been prepared to show the lattice mismatch between adjacent layers of the InGaSb intermediate layer, precisely with InSb and GaSb layers. Due to the deployment of the InGaSb intermediate layer, the lattice mismatch with the HQ GaSb buffer layer is only around 5.67–4.72% which provides a very good growth base for the crystals to grow. The lattice mismatch for the InGaSb and InSb layers are as low as 0.59–1.49% for the samples (S1–S4). The small amount of lattice mismatch indicates the possibility of the HQ InSb channel layer growth on the InGaSb intermediate layer with an effective thin film. After analyzing the images and patterns through RHEED, SEM and XRD, we can conclude that the growth nature of InSb channel layer on the InGaSb/GaSb/Si(111)-√3 × √3-Ga surface shows a coherent growth nature in compare to InGaSb epitaxial layers’ growth on GaSb/Si(111)-√3 × √3-Ga. We have seen the same bright RHEED patterns in both the cases of InSb on InGaSb/GaSb/Si(111)-√3 × √3-Ga and InGaSb on GaSb/Si(111)-√3 × √3-Ga surface phase where 2× streaks give an impression of coherent growth nature. Through SEM, it was found that the coherent growth nature shows increase in Gallium contents reduce defects to produce a smoother surface. As in our previous report, we have observed the best composition for the InGa1-xSbx lies near Gallium composition of 20% [1], the same has been observed in this case for the final epitaxial layer of 40 nm InSb layer on the InGa1-xSbx intermediate layer for the composition of InGaSb at the Gallium ratio of 0.2 for S3 leaving it to further research.

IV. CONCLUSIONS

The growth nature of InSb channel layer on InGaSb intermediate layer was observed and studied. The base template used was InGa1-xSbx grown on HQ GaSb/Si(111)-√3 × √3-Ga surface phase as our group have already reported the effectiveness of the said base layer(s). To grow the InSb channel layer the previously proven effective layers were grown with the same specifications and conditions. All the prepared major four samples were grown with 40 nm-thick InSb layer on InGa1-xSbx intermediate layer where the Ga ratio was varied from 0.10 to 0.25 with a step 0.05. Comparing the grown surfaces of InSb and the intermediate base layer of InGaSb, it is found that the RHEED patterns advocate the similar growth nature and the streaks were asymmetric during HQ GaSb buffer layer. Consequently for InGaSb, well-ordered 2× patterns were observed throughout all the experiments which stood still for InSb growth as well. Observing all the SEM images, we stand by the fact that the grown surfaces get smoother reducing defects to increase in Gallium contents for both the InGa1-xSbx intermediate layer and other epitaxial layer(s) on it. From the XRD analysis, it was found that the 2θ peaks (diffracted from InGaSb(111) peaks) shift towards the right (increasing) as Gallium contents are increased. XRD φ-scan diffracted from InGaSb(111) peaks shows the single crystalline or twin eliminating nature proving the effectiveness of HQ GaSb buffer layer. Also, some complementary samples were analyzed to support and prove the initial assumptions. The conclusions were drawn observing the 40 nm InSb channel layer growth on the InGa1-xSbx layer grown on HQ GaSb/Si(111)-√3 × √3-Ga surface base.

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