Molecular-Beam Epitaxy of 3D Topological Insulator Thin Films and Devices on Si Substrates

Gregor Mussler

Topological insulators (TIs) are currently in the focus of interest due to their intriguing physical properties related to topologically protected surface states. The ability to grow thin films of these complex layered materials and even sophisticated devices is the key to explore their fundamental phenomena, giving insights into modern solid-state physics. However, complex materials composed of layers only weakly bonded via van der Waals (vdW) forces offer unmatched challenges for the deposition of thin epitaxial films. Herein, it is reported on the growth of (Bi,Sb)2(Se,Te)3 TI films on Si (111) substrates using molecular-beam epitaxy. Special issues are discussed, such as understanding the peculiar vdw growth mode, observing and annihilating crystal defects, reducing bulk carrier concentration, tuning the Fermi level to the Dirac point, and, finally, fabricating TI/superconductor devices fully in situ.

1. Introduction

In 2006, a new state of matter called the topological insulator (TI) has been theoretically predicted and experimentally observed in a number of materials, including the most prominent 3D TIs Bi2Te3[1–5], Bi2Se3[6–9], and Sb2Te3[10–12]. These materials are characterized by gapless surface states with a linear energy dispersion, similar to relativistic particles. Hence, carriers at the surface of TIs are expected to have unparalleled properties, such as extremely high mobility and spin-locked transport. Consequently, these features may lead to new applications in the field of spintronics and quantum computation,[13–15] as well as to study new phenomena of fundamental physics, for example, the detection of Majorana excitations.[16–18]

One important aspect for investigating properties of TIs as well as realizing devices is the deposition technique. Early works on TIs investigated bulk crystals fabricated by means of the Bridgman technique, mostly to carry out ARPES measurements to evidence the topologically protected surface states.[19–21]

However, for investigating thin films and realizing devices, a sophisticated growth technique, such as molecular-beam epitaxy (MBE), is mandatory. In this review, I report on the MBE growth and characterization of (Bi,Sb)2(Se,Te)3 TIs, and I will focus on special topics, such as studying the van der Waals (vdW) growth mode of the TI films on the Si substrates, detecting crystal defects and finding solutions how to remove them, scrutinizing the origin of the large bulk carrier concentration, tuning the Fermi level to the Dirac point by controlling the composition and/or the film thicknesses, and, finally, fabricating TI/superconductor Josephson junctions (TI/SC JJs) fully in situ.

2. Optimizing the MBE Growth Parameters

To study the MBE growth of 3D TIs, numerous samples were grown on Si(111) substrates under various conditions to cover the entire growth parameter space to find optimal conditions.

Figure 1b shows the X-ray data of the Bi2Te3 film thickness is determined. Figure 1a shows the 2θ/θ scans of Bi2Te3 samples grown in a Te/Bi partial pressure ratio of 20, whereas the substrate temperature, measured by the thermocouple above the substrate heater, was varied between 200 and 600 °C. For all X-ray diffraction (XRD) curves, signatures of the Bi2Te3 sample—grown under various growth conditions. Figure 1a shows the 2θ/θ scans of Bi2Te3 samples grown in a Te/Bi partial pressure ratio of 20, whereas the substrate temperature, measured by the thermocouple above the substrate heater, was varied between 200 and 600 °C. For all X-ray diffraction (XRD) curves, signatures of the Bi2Te3 sample—grown under various growth conditions. Figure 1a shows the 2θ/θ scans of Bi2Te3 samples grown in a Te/Bi partial pressure ratio of 20, whereas the substrate temperature, measured by the thermocouple above the substrate heater, was varied between 200 and 600 °C. For all X-ray diffraction (XRD) curves, signatures of the Bi2Te3 sample—grown under various growth conditions. Figure 1a shows the 2θ/θ scans of Bi2Te3 samples grown in a Te/Bi partial pressure ratio of 20, whereas the substrate temperature, measured by the thermocouple above the substrate heater, was varied between 200 and 600 °C. For all X-ray diffraction (XRD) curves, signatures of the Bi2Te3 sample—grown under various growth conditions.
Figure 1. a) XRD curves of Bi₂Te₃ samples grown at different substrate temperatures. XRR curves of Bi₂Te₃ sample grown at b) different Te effusion cell temperatures and c) different Bi cell temperatures. Figure adapted with permission.© 2011, Elsevier.

contrast to Figure 1c, where the XRR curves of three Bi₂Te₃ samples grown at constant Te flux but different Bi fluxes are plotted. Here, the periodicity of the oscillations becomes smaller for higher Bi fluxes, indicating thicker Bi₂Te₃ films and, thus, higher Bi₂Te₃ growth rate. Hence, the growth rate of Bi₂Te₃ is not influenced by the Te flux, but solely depends on the Bi flux, in a similar fashion as the growth of GaAs.²⁻³ It is worthwhile to mention that the highest number of oscillations in the XRR curve—and thus the smoothest Bi₂Te₃ surface and interface to the Si substrate—is obtained from the sample grown at the highest Te/Bi partial pressure of 40 (lowest XRR curve in Figure 1b). It is also important to note that the growth of Bi₂Te₃ epilayers for very low Te/Bi partial pressure ratios (ϕ = 1) results in the formation of pure Bi epilayers.³⁻⁴ The growth of Bi₂Te₃ epilayers at very high Te/Bi partial pressure ratios (ϕ > 40) causes Te crystallites in the Bi₂Te₃ films, as confirmed by XRD (data not shown).

Figure 2 shows the results of the structural characterization on Bi₂Te₃ sample grown at optimized conditions. In Figure 2a, the high-resolution transmission electron microscopy (TEM) image clearly indicates the quintuple layer (QL) structure, i.e., the layers of Te–Bi–Te–Bi–Te (see inset of Figure 2b), which constitute the Bi₂Te₃ crystal. An AFM image of the Bi₂Te₃ surface is shown in Figure 2b. The color-coded image solely shows atomically flat plateaus, and the surface is made of only three different QLs, whereas one QL covers more than 75% of the surface of the 3 x 3 µm² large area. A height analysis along the black dashed line of Figure 2b is plotted in Figure 2c. Plateaus with a step height of 1 nm (equivalent to 1 QL) are seen. The observed atomic steps strongly suggest that the growth takes place by a step-flow mechanism of extending 2D islands with single QL (the QL height amounts to 1.017 nm²). The steps have specific orientations, revealing the epitaxial relationship of the crystalline structure with respect to the Si(111) substrate, which is also confirmed by pole figure measurements, as will be discussed later.

3. Studying the TI/Si(111) Interface

Another important aspect of the MBE growth of 3D-TI (Bi, Sb)₂(Te, Se)₃ films on Si(111) substrate is the TI/Si interface. Prior to growth, the Si substrate is chemically cleaned by means of the RCA recipe to remove contaminants and the native oxide and to passivate the surface with hydrogen. The substrates were subsequently heated in situ to 600 °C for 20 min to desorb the hydrogen atoms from the surface. The MBE growth of (Bi, Sb)₂(Te, Se)₃ always starts by opening the shutters of the group VI elements (Se or Te) a few seconds prior to the group V elements (Bi or Sb) to saturate the dangling Si bonds with Te or Se. Figure 3 shows a high-resolution TEM image of a Bi₂Te₃ film grown on a Si(111) substrate. Two line scans across the interfaces, averaged parallel to the interface within the yellow and red areas, were accomplished. The line scans reveal the different intensity levels of Bi, Te, and Si (see Figure 3b). The maxima within the Bi₂Te₃ layer coincide with the ideal positions schematically shown as points at the bottom of the graph. Clearly, in both areas, a Te monolayer prevails at the Bi₂Te₃/Si interface, which is covalently bonded to the Si substrate, and, thus, prepares the Si surface for vdW epitaxy. The subsequently grown QLs are only weakly bonded via vdW forces, which offer the possibility to grow fully relaxed single-crystalline (Bi, Sb)₂(Te, Se)₃ layers without forming dislocations. This finding illustrates the main advantage of vdW epitaxy, i.e., the lattice mismatch between substrate and
epilayer (i.e., 14% in case of Bi$_2$Te$_3$ on Si(111)) is not crucial, and high-quality materials can be grown by means of the vdW epitaxy on highly mismatched substrates. This is also in stark contrast to conventional epitaxy based on covalent bonds, where the covalent bonds cause strained layers for very thin films, which results in a partial relaxation with a concurrent formation of dislocations. Hence, the vdW epitaxy of TI films can be realized on a large variety of substrates, and the MBE growth of fully relaxed TI films has been successfully accomplished on SrTiO$_3$, SiC, CdS, CdTe, Al$_2$O$_3$, GaAs, GaN, graphene, InP, and Si. However, most of these substrates have a hexagonal arrangement of the surface atoms, i.e., the (111) orientation for cubic substrates and (0001) orientation for hexagonal substrates, to match the hexagonal crystal structure of the TI material, which is mandatory to warrant single-crystal growth, as will be discussed in the next paragraph.

4. Suppressing Twin Domains in the TI Films

Despite the fact that the lattice mismatch between substrate and epilayer is not crucial in the vdW growth, it is important to harness substrates that are suitable for the growth of rhombohedral crystals, such as (Bi,Sb)$_2$(Te,Se)$_3$, i.e., substrates with an hexagonal arrangement of surface atoms. Bi$_2$Te$_3$ has a rhombohedral crystal structure with a threefold symmetry in the z-direction. Hence, Bi$_2$Te$_3$ may nucleate in two different in-plane orientations with respect to the Si(111) surface (which has a sixfold rotational symmetry). Figure 4a shows the topmost atoms of a Si(111) substrate as well as two Bi$_2$Te$_3$ crystals, i.e., crystals 1 and 2. The difference between these two crystals lies in the fact that they are in-plane rotated by 60°. A sketch of these two twin domains on top of each other is shown in Figure 4b. As a simple translational shift cannot congruently transform crystal 1 into crystal 2, there will be an out-of-plane boundary when these two crystals meet. Figure 4c shows a TEM image of a Si/Bi$_2$Te$_3$ interface with twin domains, indicated by two differently hatched patterns marked by the white lines. A change of the out-of-plane lattice constant is not noticed. The out-of-plane domain boundary only spreads over a region of a few atom distances and does not form any threading dislocations. All obtained TEM images implicate domain I to be on average only present within the first six QLs of the Bi$_2$Te$_3$ film. Above this limit, the film is dominated by domain II, encapsulating domain I completely. This observation may be deemed as a first hint that there is a preferential orientation of these domains, and a

![Figure 2.](image-url)
XRD measurements were carried out. Interestingly, three out of these six peaks are of high intensity, whereas the other three peaks are less intense. The presence of twin domains in Bi$_2$Te$_3$ is suppressed for lower growth rates, owing to the suppression of these twin domains.

Another striking feature that can be seen in the pole figure scan is that the (105) peaks of the Bi$_2$Te$_3$ layers are always collinear with the (220) and (311) peaks of the Si substrate, evidencing that, despite the weak vdW interaction between the Bi$_2$Te$_3$ film and the Si substrate, the Si substrate dictates the in-plane orientation of the Bi$_2$Te$_3$ film. This also explains the fact that the growth of (Bi,Sb)$_2$(Te,Se)$_3$ films works best on substrates with a hexagonal arrangement of surface atoms (either cubic substrates with a (111) orientation or hexagonal substrates in a (001) orientation), as discussed before. Interestingly, for twin-domain suppressed Bi$_2$Te$_3$ samples, the remaining three peaks are always collinear with the (311) reflection of the Si substrate. This phenomenon can possibly be explained by a vdW interaction with the Si surface atoms and the first Si atoms underneath the surface. Whereas the Si(111) surface has a sixfold rotational symmetry, the combination of Si surface and the first atomic layer underneath the Si surface has a threefold symmetry. Owing to the threefold symmetry, the two domains have different vdW potentials, and it is the threefold symmetry that dictates the formation of a single domain with a well-defined orientation on the Si substrates.

5. Bulk Carrier Reduction and Fermi-Level Tuning

Another very important aspect of TIs is the carrier concentration, as bulk carriers may be detrimental to investigate the intriguing properties of surface carriers. It turns out that binary TIs, such as Bi$_2$Te$_3$, Bi$_2$Te$_5$, or Bi$_2$Te$_7$, have a very high intrinsic bulk carrier concentration usually larger than $1 \times 10^{19}$ cm$^{-3}$. The large bulk carrier concentrations in the binary TI materials are attributed to naturally occurring defects. For example, Bi$_2$Te$_3$ is, in general, n-type doped, and Ab Initio calculations ascribe the n-type doping by Se vacancies, grown under Se-rich conditions, as well as Sb$_{Te}$ antisites grown under Bi-rich conditions.[$^{13}$] Bi$_2$Te$_3$ is also intrinsically n-type doped, and the first-principle density functional calculations explain the n-type doping by Te$_{Bi}$ antisites, induced by Te-rich growth conditions.[$^{14}$] The Sb$_2$Te$_3$ material is, in general, p-type doped due to Sb$_{Te}$ antisites.[$^{12}$]

To reduce the bulk carrier concentration in binary TIs is extremely difficult for MBE-grown samples, as the low substrate temperature of 350–400 °C does not allow to annihilate these point defects. However, as stated earlier, Bi$_2$Te$_3$ and Sb$_2$Te$_3$ are, in general, n-type doped and p-type doped, respectively. Hence, growing the (Bi,Sb)$_2$Te$_3$ alloy with the proper Sb concentration, intrinsic material is realized.$^{[15,16]}$ Figure 6 shows the results of Hall measurements and ARPES experiments on the ternary (Bi,Sb)$_2$Te$_3$ samples with different Sb contents. In Figure 6a, the Hall resistances of a set of seven (Bi,Sb)$_2$Te$_3$ samples with Sb content between 0 and 100% are plotted in a (001) orientation, and the slope of the Hall resistance is negative, indicating n-type bulk carriers. However, for large Sb content with $x_{Sb} = 49\%$ and higher, the slope of the Hall resistance is positive, owing to p-type doping. Hence, there is a
Figure 4. a) Sketch of how the two Bi₂Te₃ twin domains nucleate on the Si(111) surface. b) Side view of these two twin domains. c) TEM image of two twin domains with two differently crossed-hatched lines (white dashed lines). One domain (green color) is overgrown by the other domain. Figure adapted with permission.[42] Copyright 2015, American Chemical Society.

Figure 5. The 2D-pole figure scans of a Bi₂Te₃ sample grown at a) high growth rates and b) low growth rates. c) The 1D-pole figure scans of Bi₂Te₃ samples grown at different growth rates. Figure adapted with permission.[42] Copyright 2015, American Chemical Society.
transition from n-type to p-type doping in (Bi,Sb)$_2$Te$_3$ for a Sb content between 30% and 49%. Consequently, a second set of (Bi,Sb)$_2$Te$_3$ samples with Sb concentration between 30% and 49% was grown and analyzed by means of Hall measurements, and the lowest carrier concentration was determined for a Sb content of 43%. For this particular sample, a high-k dielectric and a top gate were deposited to carry out Hall measurements in dependence of the top gate. Figure 6b shows the Hall resistance of the nearly intrinsic (Bi,Sb)$_2$Te$_3$ sample with $x_{\text{Sb}} = 43\%$ for different top gate voltages. By applying a negative gate voltage, the intrinsic n-type doping turns into p-type doping. The turning point is reached at a moderate gate voltage of $U_G = -7.5\, \text{V}$.

A similar behavior—a transition from n-type to p-type doping—can also be seen in ARPES experiments. ARPES scans were performed on (Bi,Sb)$_2$Te$_3$ samples with Sb contents between 49% and −100%, as shown in Figure 6c. The presented spectra show a clear shift of the Fermi level with respect to the Dirac point. In the case of an Sb content of 49%, the ARPES spectra reveal an n-type character with the Fermi level close to the conduction band, which is similar to the well-known characteristics of pure Bi$_2$Te$_3$. This is followed by an intermediate state in the 82% and 93% Sb content samples where the Fermi level seems to be close to or smeared out around the Dirac point. Pure Sb$_2$Te$_3$ films exhibit a p-type character where only the lower Dirac cone is occupied. To investigate the stoichiometry around $x_{\text{Sb}} = 93\%$ in more detail, additional studies were performed on samples that were transferred in situ from the MBE to the ARPES system. The analysis reveals that for $x_{\text{Sb}} = 94\%$, the Fermi energy lies within the gap without truncating any bulk bands, shifted only by $\approx 5\, \text{meV}$ from the Dirac point of the surface states of pure Sb$_2$Te$_3$ (data not shown). This finding is in good agreement with the results shown in this work.

The discrepancy between the Hall measurements and the ARPES experiments, i.e., the intrinsic limit is reached for (Bi,Sb)$_2$Te$_3$ with $x_{\text{Sb}} = 43\%$ whereas the Fermi level resides at the Dirac point for (Bi,Sb)$_2$Te$_3$ with $x_{\text{Sb}} = 94\%$, is possibly related to the different nature of these two experiments. Whereas transport experiments probe the entire (Bi,Sb)$_2$Te$_3$ volume, ARPES scans are only sensitive to the Bi$_2$Te$_3$ surface. Hence, a bending of conduction and valence bands at the (Bi,Sb)$_2$Te$_3$ surface may

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**Figure 6.** a) Hall resistances of several (Bi,Sb)$_2$Te$_3$ samples with different $x_{\text{Sb}}$. For low (high) $x_{\text{Sb}}$, the samples are n-type (p-type) doped with a transition from n- to p-type between $x_{\text{Sb}} = 30\%$ and 49%. b) Hall resistances of a (Bi,Sb)$_2$Te$_3$ sample with $x_{\text{Sb}} = 43\%$, which is nearly intrinsic, in dependence of a top gate. A transition from n- to p-type occurs at $U_G = -7.5\, \text{V}$. c) ARPES data of (Bi,Sb)$_2$Te$_3$ samples with $x_{\text{Sb}} = 49\%$–100%. The Fermi level is closest to the Dirac point for $x_{\text{Sb}} = 93\%$. Figure adapted with permission.[47] Copyright 2016, IOP Publishing.
be the reason why the positions of the Fermi level in bulk material and the surface are different. Band bending effects also take place in Bi$_2$Te$_3$/Sb$_2$Te$_3$ heterostructures, and they will be quantitatively analyzed, as discussed in the next paragraph.

Besides the growth of the ternary (Bi$_2$Sb$_2$Te$_3$) alloy, another way to tune the Fermi level is the fabrication of n-Bi$_2$Te$_3$/p-Sb$_2$Te$_3$ heterojunctions. The underlying idea is—such as in conventional semiconductors—to create an n–p junction that results in the formation of a built-in electric field. This field drags free carriers away from this junction to make this junction intrinsic. For this study, a set of Bi$_2$Te$_3$/$\text{Sb}_2\text{Te}_3$ stacks have been grown and analyzed, whereas the intrinsic n-type doped Bi$_2$Te$_3$ thickness was kept constant ($d_{\text{Bi}_2\text{Te}_3} = 6 \text{ nm}$), and the intrinsic p-type doped Sb$_2$Te$_3$ layer thickness was varied between 6 and 25.5 nm.

**Figure 7** shows the TEM image and atom probe tomography (APT) results of such sample. Figure 7 displays the bright-field (Figure 7a) and dark-field images (Figure 7b) of a 15 nm Sb$_2$Te$_3$ film on 6 nm Bi$_2$Te$_3$. In the dark-field image, the Bi or Sb (Si) atoms appear bright (dark). As Sb and Te exhibit almost the same atomic number, no contrast difference is revealed between Te atoms and Sb (blue arrows), and no information about Bi/Sb interdiffusion processes is obtained.

To study Bi/Sb interdiffusion processes, APT experiments were carried out, which are shown in Figure 7c. Here, we see a 2D projection of the 3D ion map of the sample. The map shows the main isotopes of each element. The interdiffusion between the Sb$_2$Te$_3$ and Bi$_2$Te$_3$ films and the formation of ternary compounds are clearly visible. Figure 7d shows the quantified concentration profiles of the heterostructure. The Te content (green curve in Figure 7d) is constant over the whole heterostructure, indicating no change in the crystal phase and diffusion processes only between Sb and Bi lattice positions. The red and cyan lines in Figure 7d clearly show that the Bi (Sb) concentration does not abruptly stop at the Bi$_2$Te$_3$/Sb$_2$Te$_3$ interface, but there is a smooth transition into the Sb$_2$Te$_3$ (Bi$_2$Te$_3$) layer. Hence, the APT data shown in (b). A substantial Bi/Sb interdiffusion across the Bi$_2$Te$_3$/Sb$_2$Te$_3$ interface is seen. Figure adapted with permission.[49] Copyright 2016, American Chemical Society.

**Figure 7.** a) Bright-field and b) dark-field images of a 15 nm Sb$_2$Te$_3$ film on 6 nm Bi$_2$Te$_3$ buffer layer. c) APT image of the Bi$_2$Te$_3$/Sb$_2$Te$_3$ stack. d) Quantitative analysis of the APT data shown in (b). A substantial Bi/Sb interdiffusion across the Bi$_2$Te$_3$/Sb$_2$Te$_3$ interface is seen.

Quantitative analysis of the APT data shown in (b). A substantial Bi/Sb interdiffusion across the Bi$_2$Te$_3$/Sb$_2$Te$_3$ interface is seen. Figure adapted with permission.[49] Copyright 2016, American Chemical Society.
6. Selective Area Growth

As already discussed, it is mandatory to have an hexagonal arrangement of surface atoms for the growth of (Bi,Sb)$_2$(Te,Se)$_3$ by virtue of the vdW forces that dictate the in-plane orientation. In contrast, the growth of these materials on amorphous substrates, such as Si$_3$N$_4$ or SiO$_2$, does not yield single-crystalline films. Conversely, for optimized substrate temperatures, no
deposition of (Bi,Sb)$_2$(Te,Se)$_3$ takes place on amorphous substrates at all. Hence, there is a certain selectivity in the MBE growth of (Bi,Sb)$_2$(Te,Se)$_3$ films with respect to the type of substrate (amorphous or crystalline), and this selective area growth (SAG) can be harnessed to in situ grow lithographically defined TI structures. These lithographically defined structures can either be Si mesas etched out of Si(111) silicon-on-insulator (SOI) substrates,[51] or Si trenches processed out of a Si/Si$_3$N$_4$ wafer, as will be discussed here. SAG-grown TI nanowires (NWs) are shown in Figure 10. Figure 10a shows the processing steps of the substrate prior to growth. Here, a Si$_3$N$_4$ film is deposited on a Si(111) substrate, and trenches are lithographically defined into the Si$_3$N$_4$ and subsequently etched by reactive ion etching (RIE) down to the Si substrates. Such substrates are then transferred into the MBE system for the growth of the TI material. Figure 10b shows a scanning electron microscopy (SEM) image of Bi$_2$Te$_3$ NWs grown on a Si$_3$N$_4$/Si substrate by means of the SAG technique. Evidently, the growth of the TI material only takes place in the trenches to form these NWs. For this sample, the dimensions of the NWs are still quite large (180 nm x 3.2 μm), but smaller dimensions can be realized, and here, the only limiting factor is the spatial resolution of the lithographic system. It is also important to note that no further ex situ processing steps are needed to define the mesas, which will become important for TI/SC JJs, which are discussed in the next section.

7. TI-Based Devices

The last chapter of the review article addresses devices based on TIs. A possible avenue to harness TIs for device application is TI/SC JJs in the context of building quantum bits (qubits) based on Majorana modes. Most important for high-quality JJs is the transparency, i.e., the structural quality of the SC/TI interface, and the best transparencies are obtained for JJs that are completely realized in situ (without exposing the sample to air, so that the TI surface oxidizes prior to the deposition of the SC contacts). In this context, we developed a new process, i.e., a method to in situ deposit superconducting contacts—with a width of less than 100 nm—on top of a TI film. A sketch of this process is shown in Figure 11, and it is based on Si$_3$N$_4$ bridges.
fabricated out of Si/SiO2/Si3N4 substrates by means of e-beam lithography, RIE, and selective etching. The TI material is grown with substrate rotation, so that the TI films are also deposited underneath the bridges. The subsequent evaporation of the SC is done by means of an electron-beam evaporator. For the growth of the SC, the substrate rotation is turned off, so that these bridges serve as a shadow mask. As a result, a tiny stripe of the TI film ($d < 100$ nm) remains uncovered by the SC film, which constitutes the weak link of the JJ. In the last step, the entire sample is covered with stoichiometric Al2O3 by means of an electron-beam evaporator (substrate rotation is turned on again) to protect the sample surface from degradation. A false-colored SEM image of (Bi,Sb)2Te3/Nb JJ is shown in Figure 12a. The width of the weak link amounts to 100 nm. Figure 12b shows a cross-sectional SEM image, where all (Bi,Sb)2Te3, Nb, and Al2O3 layers are clearly resolved.

Conventional transport and Shapiro response measurements have been carried out. It turns out that the critical currents in these in situ grown TI/SC JJs have a factor 25 higher critical current compared with junctions of the same dimensions and materials, but fabricated conventionally. The critical current of these in situ grown TI/SC JJs lies in the $\mu$A range, and their TI–SC interface transparency is close to 1. Shapiro response measurements show a first missing odd Shapiro step at low frequencies, which indicates that—apart from Andreev bound states—transport across the JJs is also mediated by Majorana bound states. In networks of SC and confined TIs, localized Majorana excitations can be created at well-defined positions. These Majorana excitations are the key ingredient for qubits based on TI–SC hybrid devices. Further details are found in the study given by Schüffelgen et al.[18]

8. Conclusion

In this review article, I have shown that it is possible to grow single-crystalline TI films on Si(111) substrate by means of MBE, if the growth takes place in a group VI overpressure regime. There are only weak vdW bonds between substrate and epilayer, which allows to grow TI films dislocation-free on substrates with large lattice mismatches. Twin domains may occur in the TI films, which can be heavily suppressed by reducing the growth rate of the TIs. Binary TIs are generally n-type doped (Bi2Te3 and Bi2Se3) or p-type doped (Sb2Te3). Intrinsic TI material is realized by growing the ternary (Bi,Sb)2Te3 alloy or n-Bi2Te3/p-Sb2Te3 junctions. The selectivity of the MBE growth of TI films on pre-structured Si/Si3N4 substrates allows to in situ fabricate TI nanostructures of arbitrary shape. Using a newly developed process, TI–SC hybrid devices can be fabricated in situ that show signatures of Majorana excitations.

Acknowledgements

These results have been accomplished as a joint project with many collaborators. Special thanks go to G. Bihlmayer, S. Blügel, S. Borisova, N. Demarina, M. Eschbach, P. Gospodaric, D. Grützmacher, A. R. Jalil, J. Kämpf, J. Köllner, P. M. Koenraad, S. Kolling, J. Krumrain, M. Lanius, M. Luysberg, E. Mlynčák, M. Morgenstern, L. Plucinski, D. Rosenbach, Th. Schäpers, M. Schleenvoigt, T. Schmitt, C. M. Schneider, P. Schüffelgen, T. Stoica, and C. Weyrich. Financial support from the Deutsche Forschungsgemeinschaft (DFG)-funded priority program SPP 1666 is greatly acknowledged. Open access funding enabled and organized by Projekt DEAL.

Conflict of Interest

The author declares no conflict of interest.

Keywords

Josephson junction, molecular-beam epitaxy, topological insulators, van der Waals epitaxy

Received: January 3, 2020
Revised: April 22, 2020
Published online: May 14, 2020

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