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Carrier density control of Sb-doped rutile-type SnO₂ thin films and fabrication of a vertical Schottky barrier diode

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We report on the control of carrier density in r-SnO₂ thin films grown on isostructural r-TiO₂ substrates by doping with Sb aiming for power-electronics applications. The carrier density was tuned within a range of 3×10^{16} – 2×10^{19} cm⁻³. Two types of donors with different activation energies, attributed to Sb at Sn sites and oxygen vacancies, are present in the thin films. Both activation energies decrease as the concentration of Sb increases. A vertical Schottky barrier diode employing a Sb:r-SnO₂/Nb:r-TiO₂ exhibits a clear rectifying property with a rectification ratio of 10³ at ± 1 V. © 2024 The Author(s). Published on behalf of The Japan Society of Applied Physics by IOP Publishing Ltd

Rutile-type (r-) oxides, such as r-GeO₂, r-SnO₂, and r-TiO₂, have been reevaluated as a new class of wide bandgap semiconductors due to their superior electrical properties for power-electronics applications, including wide bandgap and large breakdown field as well as ability to modulate the bandgap.^{1,2)} Especially, r-GeO₂ was theoretically predicted to possess ambipolar dopability and high electron/hole mobilities, along with an ultra-wide bandgap of 4.7 eV.²⁾ In addition, a bulk single crystal of r-GeO₂ is available through the flux growth method,^{3,4)} which would enable device-quality homoepitaxial growth in the future. These excellent material properties make r-GeO₂ a promising candidate for next-generation power semiconductors, leading to the accelerated research on preparation of r-GeO₂ thin films.^{5–9)}

Also, the potential of r-GeO₂ has disclosed the significance of r-SnO₂, which has the second-largest bandgap of 3.6 eV among the rutile-type oxide semiconductors;¹⁰⁾ r-SnO₂ has been accepted as an end member of r-Ge_xSn_{1–x}O₂ alloys. Recent reports indicate that r-Ge_xSn_{1–x}O₂ alloy films can be epitaxially grown on r-TiO₂ substrates throughout the entire composition range.^{1,11–13)} This allows for the tuning of bandgap between 3.8 and 4.4 eV.^{1,11,13)} The bandgap modularity suggests that r-Ge_xSn_{1–x}O₂ alloys can be utilized to produce complicated heterojunction devices such as modulation-doped field-effect transistors.¹⁾ Recently, it was also reported that r-Ge_xSn_{1–x}O₂ films could be grown coherently on r-TiO₂ substrates without misoriented domains and TEM-detectable threading dislocations in the films.¹⁴⁾ The quality of the coherent films is high enough to fabricate Schottky barrier diodes (SBDs) with high rectification ratios, demonstrating the potential of r-Ge_xSn_{1–x}O₂ alloys as practical power semiconductors.¹⁴⁾

At present, however, power-electronic applications of r-GeO₂, r-SnO₂, and r-Ge_xSn_{1–x}O₂ alloy have been limited due to the difficulty in doping control. It is well known that several n-type oxides, i.e. In₂O₃, ZnO as well as SnO₂, are unintentionally doped (usually, the carrier density $\geq 10^{18}$ cm⁻³) by intrinsic defects and/or unexpected extrinsic impurities.^{15,16)} Owing to their features of unintentional

doping, these n-type oxides have been primarily studied for their use as transparent electrodes in solar cells, flat panel displays, etc.¹⁷⁾ In contrast, for practical power-electronics applications, it is necessary to reduce the unintentional carrier density, because the drift layers ought to have carrier density controlled within a range of 10^{15} – 10^{17} cm⁻³ for standard power devices with high breakdown voltages. Previously, White et al. reported the fabrication of an unintentionally doped r-SnO₂ film with a carrier density of 2.7×10^{17} cm⁻³, while maintaining doping control within a range of $\geq 10^{18}$ cm⁻³.¹⁸⁾ The recent growing interest in rutile-type oxide semiconductors as power device materials highlights the importance of establishing their doping controllability, particularly for simple binary compounds such as r-SnO₂ and r-GeO₂ rather than complex ternary alloys. In this study, we report on the control of carrier density in r-SnO₂ thin films on isostructural r-TiO₂ (001) substrates by doping with Sb, a commonly used n-type dopant for r-SnO₂.^{18–21)} The carrier density of the films was tuned within a range of 10^{16} – 10^{19} cm⁻³ according to the amount of Sb dopant. Additionally, we fabricated vertical SBDs employing a Sb:r-SnO₂/Nb:r-TiO₂, which exhibited clear rectification properties.

Unintentionally doped and Sb-doped r-SnO₂ thin films (Sb:r-SnO₂) were grown on r-TiO₂ (001) substrates by a mist CVD technique. A SnCl₂·2H₂O which was dissolved in an aqueous solution with 10% hydrochloric acid was used as a precursor solution. An SbCl₃ was used as an Sb precursor. The molar ratio of Sb to Sn (Sb/Sn) in the precursor solutions was varied as 10⁻⁴, 10⁻³, 10⁻², and 10⁻¹ mol% by adding SbCl₃ solution which included 10% hydrochloric acid. The precursor solution was atomized using an ultrasonic transducer at a frequency of 2.4 MHz. By using O₂ gas through carrier and dilution lines at flow rates of 3.0 and 0.5 l min⁻¹, respectively, the precursor mist was supplied to a reactor consisting of a quartz tube heated at growth temperatures. The structural properties of the fabricated films were explored by X-ray diffraction (XRD) measurements using Cu K_{α1} radiation. The electrical properties of the films including carrier density and mobility were examined by Hall effect



measurements with a DC magnetic field of 0.43 T. The thicknesses of the films were measured using spectroscopic ellipsometry. The electrodes used for the Hall effect measurements (Ti/Au with van der Pauw configuration) and the SBDs (Pt and Ti/Au for anodes and cathodes, respectively) were deposited using electron beam evaporation. Before the deposition of Pt anode, the surface of Sb:r-SnO₂ film was treated with O₃ for 1 h.

First, to determine the optimized growth temperature for Sb:r-SnO₂, we evaluated the structural and electrical properties of undoped r-SnO₂ thin films grown at three different growth temperatures (700 °C, 750 °C, and 800 °C). The thicknesses of the undoped films were ~ 1000 nm. Figure 1(a) shows the XRD 2θ - ω scan profiles of undoped r-SnO₂ thin films grown at 700 °C, 750 °C, and 800 °C. The 002 diffraction peaks ascribable to r-SnO₂ and r-TiO₂ are observed for all the films. The 26.5° peaks observed in the samples grown at 700 °C and 800 °C are attributable to forbidden reflections of the 001 diffraction. For the samples grown at 800 °C, there exist additional diffraction peaks that do not originate from the (001) plane. The results indicate that (001)-oriented r-SnO₂ thin films were grown at 700 °C and 750 °C without any domain having different orientations, but at 800 °C with (011)-, (121)-, (220)- and (330)-oriented domains. XRD 301 diffraction ϕ scans were performed for the films and substrates (not shown here). For all the samples, the 301 diffraction peaks of the films appear at the same angles as the substrates at 90° intervals, which reflects a fourfold in-plane rotational symmetry of the (001)-oriented rutile-type structure. This suggests a tetragonal-on-tetragonal epitaxial relationship between the films and substrate. Carrier density and mobility of the undoped r-SnO₂ thin films measured using Hall effect are shown in Fig. 1(b). The r-SnO₂ thin film grown at the growth temperature of 700 °C exhibits the lowest carrier density of $7 \times 10^{16} \text{ cm}^{-3}$, which is much lower than those of MBE-grown and mist-CVD-grown

undoped r-SnO₂ films on *r*-plane sapphire substrates reported thus far.^{18,22)} Two reasons are possible for the lower carrier density of the film grown at 700 °C. Firstly, using r-TiO₂ substrates with smaller lattice mismatches compared to sapphire might result in fewer intrinsic defects, including interstitial Sn and oxygen vacancy. Secondly, the mist-CVD method can create an oxygen-rich atmosphere during the growth of r-SnO₂, leading to fewer oxygen vacancies. The increase in carrier density with an increase in the growth temperature is attributable to the fact that the growth condition is more reducing at higher growth temperatures. Since the reduction reaction is endothermic, Sn⁴⁺ is more easily reduced to Sn²⁺ at higher temperatures, resulting in an increase in oxygen vacancies. In contrast to the carrier density, the mobility does not manifest a clear temperature dependence. Therefore, the growth temperature of Sb:r-SnO₂ was set to 700 °C.

Sb:r-SnO₂ films with different Sb/Sn ratios were grown on r-TiO₂ (001) substrates at 700 °C. The thicknesses of the Sb-doped films were ~ 1500 nm. Figure 2 shows the carrier density and mobility of the films as a function of Sb/Sn ratio in the source solutions. The carrier density increases as the Sb/Sn ratio increases, indicating that the carrier density can be controlled by the amount of Sb in the source solution; the carrier density was tuned from 3.3×10^{16} to $1.9 \times 10^{19} \text{ cm}^{-3}$ by varying the Sb/Sn ratio from 10^{-4} to $10^{-1} \text{ mol\%}$. In particular, it is worth noting that we succeeded in controlling the carrier density in a range of $< 10^{17} \text{ cm}^{-3}$, which is favorable for drift layers of power-electronics devices. The carrier density of the Sb-doped r-SnO₂ film at the Sb/Sn ratio of $10^{-4} \text{ mol\%}$ is lower than $7 \times 10^{16} \text{ cm}^{-3}$, which is the lowest carrier density limit of undoped r-SnO₂ films. Farahani et al. reported that in lattice-mismatched systems, thicker r-SnO₂ films have lower carrier density due to less influence from dislocations with donor-like properties.²³⁾ It is speculated that the Sb-doped r-SnO₂ film was less affected by

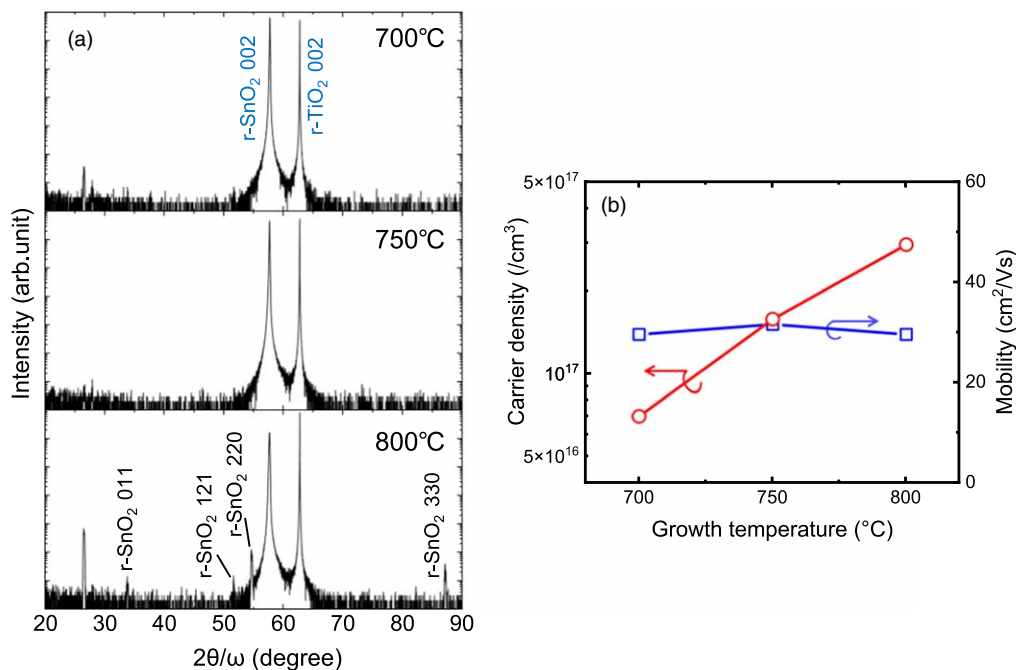


Fig. 1. (a) XRD 2θ - ω scan profiles of undoped r-SnO₂ thin films grown at 700 °C, 750 °C, and 800 °C. A single phase is grown at 700 °C and 750 °C. (b) Carrier density and mobility of the thin films as a function of growth temperature. Red circle and blue square represent carrier density and mobility, respectively.

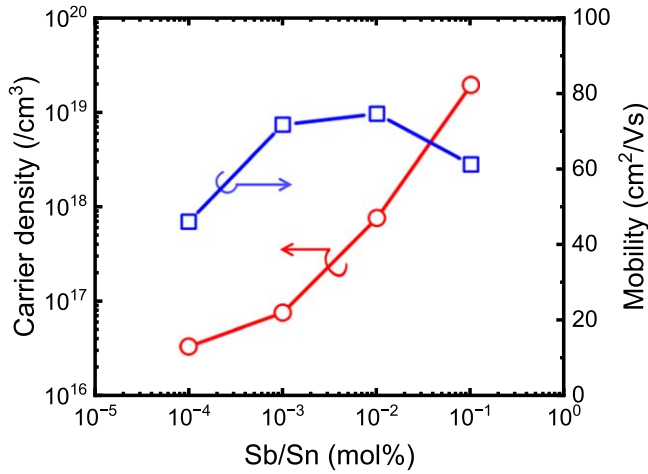


Fig. 2. Carrier density and mobility of Sb:r-SnO₂ films as a function of Sb/Sn ratio.

dislocations due to its larger thickness than that of the undoped film, resulting in its lower carrier density.

To clarify the states of donors and carrier transport mechanisms, temperature-dependent Hall effect measurements were performed for the Sb:r-SnO₂ films. The data for the Sb:r-SnO₂ films grown at Sb/Sn ratios of 10⁻⁴ and 10⁻³ mol% at temperatures below 240 and 200 K, respectively, are not shown because those data are scattered. Figure 3(a) illustrates carrier densities as a function of temperature for all the fabricated films. Figure 3(b) represents magnified data of the Sb:r-SnO₂ films grown at Sb/Sn ratios of 10⁻⁴ and 10⁻³ mol%. The activation energy of the donors in each film was estimated by

the linear fitting of the data assuming the Arrhenius equation. Only one type of activation energy is observed for the films grown at Sb/Sn ratios of 10⁻² and 10⁻¹ mol%. In contrast, two types of activation energies are found for the films grown at Sb/Sn ratios of 10⁻⁴ and 10⁻³ mol% as depicted in Fig. 3(b), indicating that there are two types of donors in the Sb:r-SnO₂ films. The activation energies observed in the films at Sb/Sn ratios of 10⁻¹ and 10⁻² mol% and those with smaller values in the films at Sb/Sn ratios of 10⁻³ and 10⁻⁴ mol%, which decreases from 51.7 to 0.14 meV as the Sb/Sn ratio increases, are considered to originate from Sb at Sn sites (Sb_{Sn}).¹⁹ It is worth noting that the electronic structure of the film grown at 10⁻¹ mol% is almost degenerate, as evidenced by its nearly zero activation energy, i.e. 0.14 meV. On the other hand, the activation energies with larger values in the films at Sb/Sn ratios of 10⁻³ and 10⁻⁴ mol%, i.e. 111.4 and 198.1 meV, respectively, are considered to originate from oxygen vacancies.^{19,24} As the ratio of Sb to Sn increases, the activation energy of oxygen vacancies decreases. The decrease in the activation energy of each donor (Sb_{Sn} and oxygen vacancies) is caused by two factors: the screening effect of the Coulomb potential of dopants by charged impurities and free carriers, and spatial fluctuations at the conduction band-edge due to the potential of charged impurities.²⁵ Such behavior of the activation energy has been also reported elsewhere.^{26,27} This suggests that Sb is incorporated accordingly as the Sb/Sn ratio increases. Additionally, the number of oxygen vacancies increases. This increase in the number of oxygen vacancies is, in part, attributed to the decrease in crystallinity resulting from Sb doping. Figure 3(c) shows the temperature dependence of mobility for all the films. The mobility of the films grown at

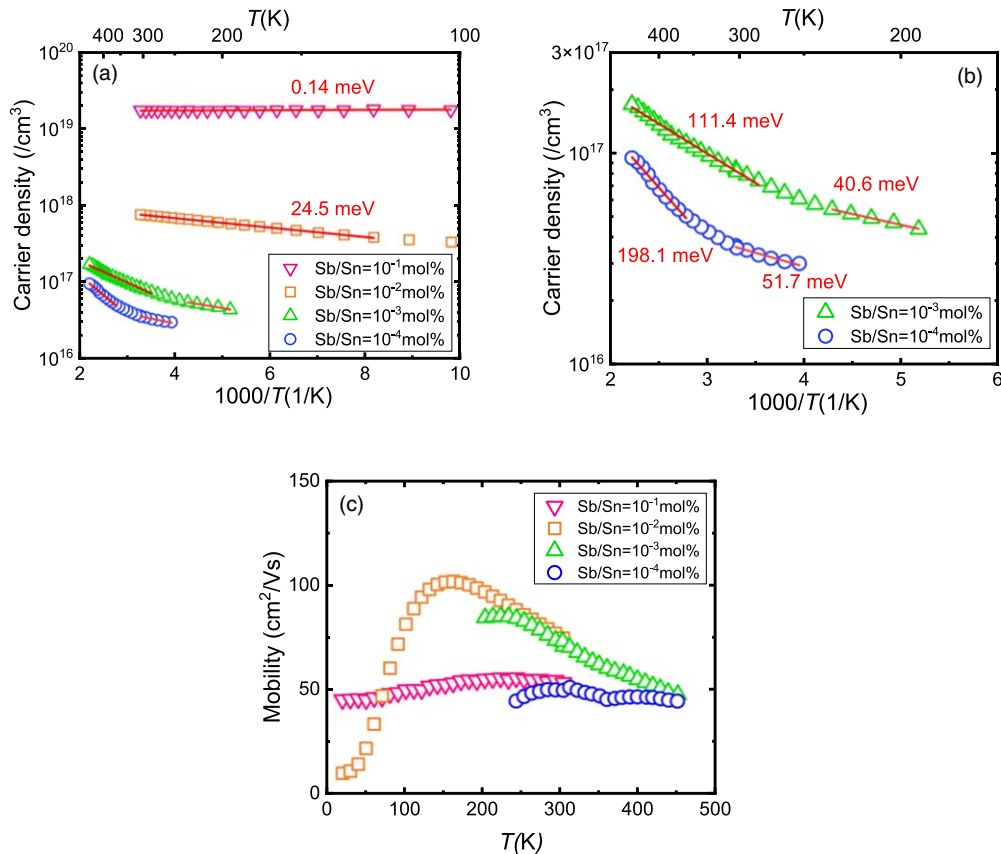


Fig. 3. (a) Temperature dependence of carrier density for Sb:r-SnO₂ films. (b) Enlarged figure for the samples with Sb/Sn ratio = 10⁻³ and 10⁻⁴ mol%. Solid lines represent linear fit to the data (Arrhenius equation). (c) Temperature dependence of mobility for Sb:r-SnO₂ films.

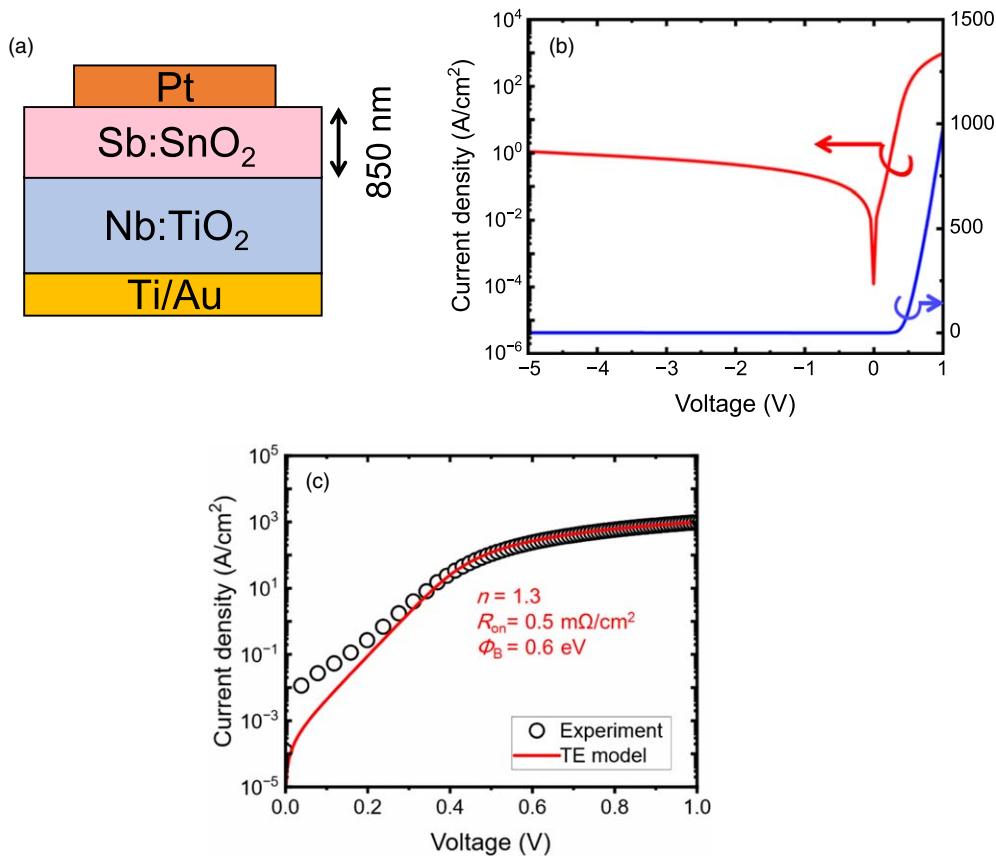


Fig. 4. (a) A schematic illustration of the structure of Sb:r-SnO₂ vertical SBD. (b) *J-V* characteristics of the SBD. (c) Forward *J-V* characteristics of the SBD experimentally obtained (open circles) along with the calculated values based on the TE model (red line).

Sb/Sn ratios of 10^{-3} and 10^{-4} mol% decreases as the temperature decreases at low temperatures. This suggests that the ionized impurity or dislocation scattering is the dominant limiting factor of the mobility at low temperatures.²⁸⁾ Despite their lower carrier density, their mobility is lower than that of the films at Sb/Sn ratios of 10^{-2} mol% at all temperatures. Therefore, the dislocation scattering is considered to be the dominant limiting factor of the mobility at low temperatures in the films grown at Sb/Sn ratios of 10^{-3} and 10^{-4} mol%. In contrast, the mobility of the film grown at Sb/Sn ratios of 10^{-1} mol% exhibits minimal temperature dependence at low temperatures, confirming that the electronic structure of the film is almost degenerate, as mentioned above. Although the film should contain numerous dislocations as previously reported for the r-SnO₂/r-TiO₂ systems,^{29,30)} the dislocation scattering becomes less effective for degenerate semiconductors by the screening effect.^{31,32)} Therefore, the ionized impurity scattering is considered to be the dominant limiting factor of mobility at low temperatures in the film grown at Sb/Sn ratios of 10^{-1} mol%. For the film grown at Sb/Sn ratios of 10^{-2} mol%, both the ionized impurity and dislocation scattering are considered to be effective. Since the mobilities of the films grown at Sb/Sn ratios of 10^{-2} , 10^{-3} , and 10^{-4} mol% decrease with increasing temperature at >160, 250, and 300 K, respectively, the polar optical phonon scattering is considered to be effective at higher temperatures.²⁸⁾

Finally, vertical SBDs were fabricated by using the Sb-doped r-SnO₂ thin film grown at Sb/Sn ratio of 10^{-4} mol% on a highly conductive Nb-doped r-TiO₂ (001) substrate. A schematic illustration of the SBDs structure is shown in Fig. 4(a). Figure 4(b) displays the current density–voltage

(*J-V*) curve of the SBDs, indicating a clear rectifying property with a rectification ratio of 10^3 at ± 1 V. As presented in Fig. 4(c), the forward *J-V* characteristic follows the thermionic emission (TE) model,³³⁾ revealing that the on-resistance is as low as $0.5 \text{ m}\Omega/\text{cm}^2$, however, the ideality factor is 1.3, which is far from unity. Moreover, the high reverse leakage current of over 1 A cm^{-2} was observed at -5 V. These high ideality factor and reverse leakage current are primarily due to the presence of additional current components such as tunneling current and/or defect-induced current caused by the charge accumulation at the surface^{34–36)} and threading dislocations. There have been some attempts to improve Schottky contacts of oxide semiconductors including oxygen plasma treatment,³⁵⁾ and oxidized metal contact,^{37–39)} which should be our future works.

In conclusion, our study sought to establish the Sb doping in r-SnO₂. The undoped r-SnO₂ thin film grown on isostructural r-TiO₂ (001) substrates at a growth temperature of 700 °C exhibits a carrier density of $7 \times 10^{16} \text{ cm}^{-3}$. The carrier density in the fabricated Sb-doped r-SnO₂ thin films can be tuned within a range of 3.3×10^{16} to $1.9 \times 10^{19} \text{ cm}^{-3}$ by varying the Sb/Sn ratio in the source solutions from 10^{-4} to 10^{-1} mol%. There are two types of activation energies for the films grown at Sb/Sn ratios of 10^{-4} and 10^{-3} mol%. They are considered to be Sb at Sn sites and oxygen vacancies. The activation energy of both types of donors decreases as the Sb/Sn ratio increases. The mobility of the film with the lowest carrier density of $3.3 \times 10^{16} \text{ cm}^{-3}$ is limited by the dislocation scattering. Additionally, vertical SBDs employing

an Sb:r-SnO₂/Nb:r-TiO₂ were fabricated. The vertical SBDs exhibit a clear rectifying property with a rectification ratio of 10³ at ±1 V. We believe that our research should contribute to exploring the doping control methods of rutile-type oxide semiconductors including r-GeO₂, r-SnO₂, and r-Ge_xSn_{1-x}O₂ alloy for future device applications.

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- 1) H. Takane, Y. Ota, T. Wakamatsu, T. Araki, K. Tanaka, and K. Kaneko, *Phys. Rev. Mater.* **6**, 084604 (2022).
- 2) S. Chae et al., *Appl. Phys. Lett.* **118**, 260501 (2021).
- 3) C. A. Niedermeier, K. Ide, T. Katase, H. Hosono, and T. Kamiya, *J. Phys. Chem. C* **124**, 25721 (2020).
- 4) S. Chae et al., *J. Vac. Sci. Technol. A* **40**, 050401 (2022).
- 5) S. Chae, H. Paik, N. M. Vu, E. Kioupakis, and J. T. Heron, *Appl. Phys. Lett.* **117**, 072105 (2020).
- 6) H. Takane and K. Kaneko, *Appl. Phys. Lett.* **119**, 062104 (2021).
- 7) G. Deng, K. Saito, T. Tanaka, M. Arita, and Q. Guo, *Appl. Phys. Lett.* **119**, 182101 (2021).
- 8) G. Deng, Y. Huang, Z. Chen, K. Saito, T. Tanaka, M. Arita, and Q. Guo, *Mater. Lett.* **326**, 132945 (2022).
- 9) G. Deng, Y. Huang, Z. Chen, K. Saito, T. Tanaka, M. Arita, and Q. Guo, *J. Lumin.* **267**, 120353 (2024).
- 10) K. Reimann and M. Steube, *Solid State Commun.* **105**, 649 (1998).
- 11) Y. Nagashima, M. Fukumoto, M. Tsuchii, Y. Sugisawa, D. Sekiba, T. Hasegawa, and Y. Hirose, *Chem. Mater.* **34**, 10842 (2022).
- 12) F. Liu, T. K. Truttmann, D. Lee, B. E. Matthews, I. Laraib, A. Janotti, S. R. Spurgeon, S. A. Chambers, and B. Jalan, *Commun. Mater.* **3**, 69 (2022).
- 13) H. Takane, I. Takeya, H. Izumi, T. Wakamatsu, Y. Isobe, K. Kaneko, and K. Tanaka, *J. Appl. Phys.* **134**, 165706 (2023).
- 14) H. Takane, T. Oshima, T. Harada, K. Kaneko, and K. Tanaka, *Appl. Phys. Express* **17**, 011008 (2024).
- 15) P. Ágoston, K. Albe, R. M. Nieminen, and M. J. Puska, *Phys. Rev. Lett.* **103**, 245501 (2009).
- 16) P. D. C. King et al., *Phys. Rev. B* **80**, 081201 (2009).
- 17) T. Minami, *Semicond. Sci. Technol.* **20**, S35 (2005).
- 18) M. E. White, O. Bierwagen, M. Y. Tsai, and J. S. Speck, *J. Appl. Phys.* **106**, 093704 (2009).
- 19) C. G. Fonstad and R. H. Rediker, *J. Appl. Phys.* **42**, 2911 (1971).
- 20) R. F. Martinez-Gazoni, M. W. Allen, and R. J. Reeves, *Phys. Rev. B* **98**, 155308 (2018).
- 21) J. I. Scott, R. F. Martinez-Gazoni, M. W. Allen, and R. J. Reeves, *J. Appl. Phys.* **126**, 135702 (2019).
- 22) T. Okuno, T. Oshima, S. D. Lee, and S. Fujita, *Phys. Status Solidi C* **8**, 540 (2011).
- 23) S. K. V. Farahani, T. D. Veal, A. M. Sanchez, O. Bierwagen, M. E. White, S. Gorfman, P. A. Thomas, J. S. Speck, and C. F. McConville, *Phys. Rev. B* **86**, 245315 (2012).
- 24) J. A. Marley and R. C. Dockerty, *Phys. Rev.* **140**, A304 (1965).
- 25) B. I. Shklovskii and A. L. Efros, *Electronic Properties of Doped Semiconductors* (Springer, Berlin, 1984).
- 26) K. Irmscher, Z. Galazka, M. Pietsch, R. Uecker, and R. Fornari, *J. Appl. Phys.* **110**, 063720 (2011).
- 27) K. Akaiwa, K. Kaneko, K. Ichino, and S. Fujita, *Jpn. J. Appl. Phys.* **55**, 1202BA (2016).
- 28) M. Lundstrom, *Fundamentals of Carrier Transport* (Cambridge University Press, Cambridge, 2000).
- 29) T. Zin Win, K. Inaba, S. Kobayashi, Y. Kanetake, and Y. Nakamura, *Trans. Mater. Res. Soc. Jpn.* **45**, 173 (2020).
- 30) H. Takane, T. Oshima, K. Tanaka, and K. Kaneko, *Appl. Phys. Express* **16**, 045503 (2023).
- 31) H. M. Ng, D. Doppalapudi, T. D. Moustakas, N. G. Weimann, and L. F. Eastman, *Appl. Phys. Lett.* **73**, 821 (1998).
- 32) H. Takane, H. Izumi, H. Hojo, T. Wakamatsu, K. Tanaka, and K. Kaneko, *J. Mater. Res.* **38**, 2645 (2023).
- 33) S. M. Sze, *Physics of Semiconductor Devices* (Wiley, Hoboken, 1981) 2nd ed.
- 34) P. D. C. King and T. D. Veal, *J. Phys.: Condens. Matter* **23**, 334214 (2011).
- 35) T. Nagata, O. Bierwagen, M. E. White, M. Y. Tsai, and J. S. Speck, *J. Appl. Phys.* **107**, 033707 (2010).
- 36) S. K. V. Farahani, T. D. Veal, J. J. Mudd, D. O. Scanlon, G. W. Watson, O. Bierwagen, M. E. White, J. S. Speck, and C. F. McConville, *Phys. Rev. B* **90**, 155413 (2014).
- 37) H. Von Wenckstern, D. Splith, F. Schmidt, M. Grundmann, O. Bierwagen, and J. S. Speck, *APL Mater.* **2**, 046104 (2014).
- 38) A. M. Hyland, R. A. Makin, S. M. Durbin, and M. W. Allen, *J. Appl. Phys.* **121**, 024501 (2017).
- 39) C. Hou, K. R. York, R. A. Makin, S. M. Durbin, R. M. Gazoni, R. J. Reeves, and M. W. Allen, *Appl. Phys. Lett.* **117**, 203502 (2020).