Electron-transport properties of degenerate ZnSnN$_2$ doped with oxygen

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Abstract

In this study, analysis of the electron mobility in ZnSnN$_2$ epilayers that were unintentionally doped with oxygen (ZnSnN$_{2-x}$O$_x$) was performed to elucidate the reason for the low mobilities of $\sim$ 20 cm$^2$ V$^{-1}$ s$^{-1}$. While roughly 30% of the incorporated oxygen donated electrons, the rest existed as neutral impurities. Seebeck-effect measurements revealed that scattering by neutral impurities governed the electron transport. The theoretical mobility calculated taking into account the scattering by neutral impurities and ionized impurities reproduced the experimental Hall mobility. We concluded that the low electron mobility is attributed to the presence of the neutral oxygen impurities in high concentration.

Keywords: ZnSnN$_2$, Mobility, Impurity scattering, Hall effect, Seebeck effect

Introduction

Zn–IV–N$_2$ (IV = Si, Ge, Sn) compounds, which are derived from wurtzite-type group III nitride by replacing the group III elements with an equal number of Zn and group IV elements, can be regarded as a pseudo group III nitride [1, 2]. It has been demonstrated that similar to the well-known In$_x$Ga$_{1-x}$N system, the bandgap ($E_g$) of alloyed ZnSn$_{1-x}$Ge$_x$N$_2$ was tunable from 3.1 to 2.0 eV by varying the Sn content ($x$) from 0 to 1 [3]. This makes Zn–IV–N$_2$ an intriguing semiconductor system, and studies on ZnSn$_{1-x}$Ge$_x$N$_2$ as a counterpart to the In$_x$Ga$_{1-x}$N system have been carried out in recent years [3, 4].

Zinc tin nitride (ZnSnN$_2$) is the least studied Zn–IV–N$_2$ compound. For example, the synthesis of ZnSnN$_2$ thin films [5, 6] and powders [7, 8] has been reported in only the last 5 years. Since theoretical studies showed that the $E_g$ value of 1.0–2.0 eV for ZnSnN$_2$ is appropriate for a photovoltaic absorber [1, 2, 9–11], experimental investigations have been carried out on this material [12–15]. Indeed, ZnSnN$_2$ has a large optical absorption coefficient ($\sim$ 10$^3$ cm$^{-1}$) in most of the solar spectrum [13]. In addition, p-Si/n-ZnSnN$_2$ and p-SnO/n-ZnSnN$_2$ p–n junctions for photovoltaic application were successfully fabricated recently [15, 16].

ZnSnN$_2$ has two phases that depend on the ordering of the cation sublattice: the ordered phase, which is derived by alternately replacing the cation sublattice with Zn and Sn in the wurtzite structure; the disordered-phase, in which Zn and Sn randomly occupy the cation sublattice. Theoretical studies showed that $E_g$ of ZnSnN$_2$ is dependent on the cation-sublattice ordering: $E_g = 2$ eV for the ordered phase and $E_g = 1$ eV for the disordered phase [10, 17]. Optical studies on ordered- and disordered-ZnSnN$_2$ thin films obtained $E_g$ values close to the above-mentioned theoretical values [10, 12, 14, 17, 18]. Owing to intensive study from theoretical and experimental points of view, the intrinsic $E_g$ value of ZnSnN$_2$ has gradually become clear.

However, even though control of the transport properties is also important for photovoltaic applications, such properties of ZnSnN$_2$ are not well understood. ZnSnN$_2$ thin films usually show degenerate $n$-type conductivity owing to unintentional oxygen doping [12, 18]. The oxygen impurities occupy the nitrogen sites and behave as...
electron donors [11]. As a result, ZnSnN₂ films have electron densities of the order of 10¹⁹–10²¹ cm⁻³ [9, 10, 12, 14, 17, 18]. The electron mobility of ZnSnN₂ thin films are usually 10 cm² V⁻¹ s⁻¹ or lower [9, 10, 12, 14, 17, 18] even though the films are single crystalline. To establish ZnSnN₂ as a photovoltaic absorber, the major challenges include suppression of the conduction electron density and enhancement of the electron mobility.

For the suppression of the conduction electron density, Floretti et al. proposed a method in which hydrogen doping takes place during film growth and subsequent annealing [19]. They successfully obtained nondegenerate Zn₁₋ₓSnₓN₂ films with a carrier density of 10¹⁶ cm⁻³. On the other hand, no method for improving the electron mobility has been reported so far, and the low electron mobility remains one of the open questions of the electron-transport properties of ZnSnN₂. In the case of heavily-doped GaN epilayers with an electron density of ~10²⁰ cm⁻³ [20], an electron mobility above 100 cm² V⁻¹ s⁻¹ has been obtained. In comparison, the electron mobility of ZnSnN₂ appears to be unusually low, but an understanding of the limiting factor of the electron mobility in ZnSnN₂ is lacking.

In this study, detailed analysis of the electron-transport properties of epitaxially grown ZnSnN₂ thin films was carried out to elucidate the reason for the low electron mobility. Herein, we present the analysis results and discuss in detail the limiting factor of the mobility.

**Results and discussion**

Cation-disordered ZnSnN₂ films were grown on the (111) plane of yttria-stabilized zirconia (YSZ) single-crystalline substrates with a temperature of 300 °C; the films were grown by reactive radio-frequency magnetron sputtering using a Zn–Sn alloy target with Zn concentration of 50 at.%. In general, high growth temperature is preferable for the growth of thin films composed of larger crystalline grains with higher crystallinity. In the case of ZnSnN₂, however, the highest growth temperature is limited to ~300 °C, because the decomposition temperature of this compound was estimated to be ~350 °C [8]. Indeed, we previously confirmed that no film was grown on YSZ (111) at 350 °C [18]. Accordingly, 300 °C is the highest temperature for the growth of ZnSnN₂ thin films.

Out-of-plane and in-plane X-ray diffraction (XRD) measurements confirmed that the ZnSnN₂ films were epitaxially grown on YSZ(111) with the epitaxial relationship of (0001)|[ZnSnN₂] || (111)[Sn] || YSZ (Additional file 1: Fig. S1). The Zn/(Zn + Sn) atomic ratio in the films was determined by X-ray photoelectron spectrometry (XPS) to be in the range of 0.52–0.55. That is, slightly Zn-rich off-stoichiometric films were obtained in this study. The XPS measurements also confirmed that all the films were unintentionally doped with oxygen, forming ZnSnN₂₋ₓOₓ. The oxygen content x could be controlled by varying the nitrogen partial pressure (PN₂) during the growth [18]. When PN₂ was decreased from 2.0 to 1.2 Pa, x in ZnSnN₂₋ₓOₓ increased from 0.056 to 0.079 (the x values were semi-quantitatively determined by using the integrated intensity of the O 1s core spectra). The XPS spectra and elemental composition data are given in Additional files 2 and 3, respectively. Impurity substitution usually expand or shrink a unit cell, leading to a small shift of XRD peak. In the case of the ZnSnN₂₋ₓOₓ films with x<0.08, however, significant peak shift by the incorporation of oxygen was not observed (Additional file 4: Fig. S3a), indicating that the lattice constant did not change (Additional file 4: Fig. S3b). The difference of the ionic radii of N³⁻ (146 pm) and O²⁻ (138 pm) is so small that the oxygen substitution hardly caused the change of the lattice constant.

Hall-effect measurements were carried out to determine the carrier density (nₑ) and Hall mobility (μₑ) in the films. As shown in Fig. 1a, nₑ increased with increasing x, indicating that the unintentionally incorporated oxygen (Oₓ) served as an electron donor. Temperature (T)-independent behavior of nₑ can be seen in Fig. 1b, which suggests that the ZnSnN₂₋ₓOₓ films were degenerate semiconductors. The curve in Fig. 1a represents nₑ calculated assuming that 30% of O(N) was ionized (the

![Figure 1](image-url)
ionization rate of \( \eta = 30\% \)). The curve reasonably approximated the experimental \( n_e \) values, indicating that ON was highly compensated at room temperature (RT). The reason for such high compensation is probably related to the Zn-rich composition of the films. The Zn excess introduces ZnSn acceptor-like defects in ZnSnN\(_2\) [19]. Recent theoretical calculation suggests that electrically neutral ZnSn\(_{2}\)-2ON complexes forms when ZnSn and ON coexist in ZnSnN\(_2\) [21]. The formation of the ZnSn\(_{2}\)-2ON complexes implies the compensation of the ON donors. Hence, many ZnSn\(_{2}\)-2ON complexes are likely to present in the epilayers.

Figure 1c shows the \( x \) dependence of \( \mu_H \) at RT. The unintentional oxygen doping caused the decrease in \( \mu_H \), suggesting that ON acted as an impurity scattering center. Indeed, the \( \mu_H \) values were independent of \( T \) (Fig. 1c), representing typical behavior of mobility limited by ionized- or neutral-impurity scattering in degenerate semiconductors. Furthermore, the \( T \)-independent behavior of \( \mu_H \) clearly indicates that the contribution of \( T \)-dependent phonon scattering to electron transport was negligible.

Recently, Hamilton et al. reported that grain-boundary scattering is one of mobility-limiting factors in polycrystalline films in ZnSnN\(_{2-x}\)O\(_x\) [19]. Even in epitaxial films, grain-boundary scattering sometimes cannot be ignored, because epitaxial films frequently have a biaxially-oriented grain structure [22–24]. As we previously reported, ZnSnN\(_2\) epitaxial films sputtered on YSZ (111) have a compact grain structure with lateral grain diameters of ~30 nm [18]. In other words, sputtered ZnSnN\(_{2-x}\)O\(_x\) epilayers are not single-crystalline films but rather biaxially-oriented polycrystalline films. Accordingly, grain-boundary scattering should be considered, when we analyze the electron transport properties of the sputtered ZnSnN\(_{2-x}\)O\(_x\) epilayers.

The optically derived mobility \( \mu_{\text{opt}} \) using the Drude model corresponded to the intra-grain mobility without the grain boundary contribution, whereas DC-measured \( \mu_H \) included the contribution of the grain-boundary scattering. To obtain the \( \mu_{\text{opt}} \) values, we measured the infrared reflectance spectra of the ZnSnN\(_{2-x}\)O\(_x\) films at near-normal incidence (~5°) in the wavelength range of 1–5 μm, and then fitting analysis was carried out using the Drude model. In our previous study, we showed that the frequency \( (\omega) \)-dependent dielectric function of the ZnSnN\(_{2-x}\)O\(_x\) \( (\varepsilon(\omega)) \) can be modeled by combining the Drude function \( (\varepsilon_D(\omega)) \) with double Tauc–Lorentz (TL) functions \( (\varepsilon_{\text{TL1}}(\omega) \text{ and } \varepsilon_{\text{TL2}}(\omega)) \): i.e., \( \varepsilon(\omega) = \varepsilon_{\text{TL1}}(\omega) + \varepsilon_{\text{TL2}}(\omega) + \varepsilon_D(\omega) \). The Drude function is given by

\[
\varepsilon_D(\omega) = -\frac{\omega_p^2}{\omega^2 - i\Gamma_D\omega},
\]

where \( \omega_p \) is the plasma frequency, \( \Gamma_D \) is the scattering rate, \( \varepsilon_0 \) is the static dielectric constants of free space, and \( m^* \) is the effective mass. The TL function is given in Ref. [25]. The theoretical reflectance spectra calculated through the Fresnel formula combined with the modeled dielectric function were fitted to the experimental spectra \( (\varepsilon_\text{ex} \text{ and } \Gamma_D) \) were used as the fitting parameters. In the fitting analysis, the parameters in the TL functions were fixed for simplicity at constant values that were the same as reported values in the literature [13]. We obtained good fits, as shown in Fig. 2a–d. The best-fit parameters are listed in Additional file 7: Table S2 (Additional file 7).

Using the best-fit parameters, the \( \mu_{\text{opt}} \) values were calculated using the relationship

\[
\mu_{\text{opt}} = \frac{1}{en_e} \frac{\varepsilon_0 \omega_p^2}{\Gamma_D}.
\]

The values of \( \mu_{\text{opt}} \) and \( \mu_H \) were almost identical, as shown in Fig. 1c, indicating that the contribution of grain-boundary scattering to the electron transport in the ZnSnN\(_{2-x}\)O\(_x\) epilayers was negligible. In other words, \( \mu_H \) corresponded to the intra-grain mobility, implying that the electron transport properties of the ZnSnN\(_{2-x}\)O\(_x\) epilayers were close to those of single crystalline films.

Here, we compare \( \mu_H \) values of the ZnSnN\(_{2-x}\)O\(_x\) epilayers with those of InN epilayers. Figure 3 shows \( n_e \) versus \( \mu_H \) for the ZnSnN\(_{2-x}\)O\(_x\) and InN epilayers [26, 27]. The \( \mu_H \) values of the present ZnSnN\(_{2-x}\)O\(_x\) films are in comparable level with those of the degenerate InN epilayers reported in the 1990s. Nevertheless, the \( \mu_H \) values in the present study are substantially lower than those of the
InN layers fabricated in 2010s. We believe that there is still room for the improvement of electron mobility, if the mobility-limiting factors are fully understood.

To obtain insight into the intra-grain electron transport in ZnSnN$_{2-x}$O$_x$, the $\mu_H$ values were compared with theoretical models. We first considered ionized-impurity scattering, which generally governs the electron transport in heavily-doped degenerate semiconductors. The theoretical mobilities limited by the ionized-impurity scattering ($\mu_I$) were calculated by using the Brooks–Herring model [28]. We used an effective mass of $0.37m_0$ [18] and a static dielectric constant ($\varepsilon_s$) of 11 for the calculation. The calculated $\mu_I$ is plotted as a function of $n_e$ in Fig. 4a. The $\mu_I$ curve gives the upper limit of the electron mobility in ZnSnN$_{2-x}$O$_x$; namely, the electron mobility of ZnSnN$_{2-x}$O$_x$ potentially reaches $\sim 100$ cm$^2$ V$^{-1}$ s$^{-1}$. Nevertheless, the experimental $\mu_H$ values had a maximum of 20 cm$^2$ V$^{-1}$ s$^{-1}$, implying that an additional scattering source must be taken into account.

Thermopower measurements were carried out for further investigation. For highly degenerate $n$-type semiconductors with a parabolic conduction band, Seebeck coefficient ($S$) is given by [28]

$$S = \frac{-8\pi^2}{3} \left( r + \frac{3}{2} \right) \frac{k_B^2 m^* T}{e^2 h^2} n_e^{-\frac{2}{3}},$$

where $r$ represents the scattering constant (e.g., $r=3/2$ for ionized-impurity scattering, and $r=0$ for neutral-impurity scattering), $k_B$ is the Boltzmann constant, and $h$ is the Planck constant. A plot of $S$ versus $n_e^{-2/3}$ allows us to gain insight into the electron-transport mechanisms from the viewpoint of $r$ values [29]. As shown in Fig. 4b, the experimental data lie close to the theoretical curve for $r=0$ (solid line), whereas the experimental $S$ values are located away from the theoretical $S$ curve (dashed line) for the ionized-impurity scattering ($r=3/2$). These trends suggest that neutral-impurity scattering was the dominant scattering mechanism. The temperature-independent behavior of $\mu_H$ is consistent with this interpretation because neutral-impurity scattering does not depend on temperature [30, 31]. Since $\sim 70\%$ of O$_N$ was not ionized (Fig. 1a), it is reasonable to suppose that the compensated oxygen that is likely to form the Zn$_{10}$–2O$_N$ complex acted as the neutral-impurity scattering centers. In fact, Hamilton et al. recently pointed out the possibility that the Zn$_{10}$–2O$_N$ complexes behave as the neutral-impurity scattering centers [21].

We further analyzed the $n_e$–$\mu_H$ data by taking account of the neutral-impurity scattering. We hypothesized that all the compensated oxygen served as neutral-impurity scattering centers, namely, the density of the neutral impurity ($n_N$) was defined as $n_N = (n_e/\eta)(1-\eta)$. The mobility limited by the neutral-impurity scattering ($\mu_N$) was calculated on the basis of the model developed by Mayer et al. [31]:

![Fig. 3](image1.png)

**Fig. 3** Comparison of Hall mobility ($\mu_H$) as a function of electron density ($n_e$) for ZnSnN$_{2-x}$O$_x$ and degenerate InN epilayers [26, 27]. All the data are included in Additional file 8

![Fig. 4](image2.png)

**Fig. 4** a Dependence of Hall mobility on carried density. The circles represent data for the same films shown in Figs. 1 and 2. The triangles represent data for films grown under conditions similar to those employed for the grown of films shown in Figs. 1 and 2. The lines labeled $\mu_I$ and $\mu_N$ are theoretical curves of mobility limited by ionized- and neutral-impurity scatterings, respectively. The curve labeled $\mu_{total}$ represents the total mobility calculated from $\mu_{total} = (1/\mu_I + 1/\mu_N)^{-1}$. b Seebeck coefficient as a function of (carrier density)$^{-2/3}$; theoretical lines for $r=3/2$ (dashed line) and 0 (solid line) are also included. All the data are included in Additional file 9.
the experimental data was found to be located near the growth of an additional series of ZnSnN$_2$ epitaxial films with very high concentration of neutral oxygen impurities. From the Drude dielectric function (5)

$$\mu_N = \frac{e^3 m^*}{4\pi \varepsilon_0 \hbar^2} \frac{1}{n^2}$$

where $w = (k_\text{F}\alpha_B)^2$,

$$A(w) = \frac{35.2 (1 + e^{-50w}) (1 + 80.6w + 23.7w^2)}{w^2} \left[ \frac{1}{w} \ln(1 + w) - \frac{1}{2w} \ln \left( \frac{1 + \frac{1}{8w} - \frac{1}{6w} w^2}{1 + w} \right) \right],$$

$k_\text{F}$ is the Fermi wave number, and $\alpha_B$ is the effective Bohr radius ($\alpha_B = 1.5$ nm for ZnSnN$_2$). Assuming that the ionization ratio of O$_\text{N}$ was $\eta = 30\%$, the total mobility was calculated using Matthiessen’s rule, $\mu_{\text{Total}} = (1/\mu_N + 1/\mu_\text{Total})^{-1}$. The calculated $n_e - \mu_{\text{Total}}$ curve agrees well with the experimental $n_e - \mu_\text{Total}$ data, as shown in Fig. 4a. Furthermore, we grew an additional series of ZnSnN$_2$-xO$_x$ epitaxial films under conditions similar to those mentioned above, and their $n_e - \mu_{\text{Total}}$ data are also plotted in Fig. 4a (triangles). Again, the experimental data was found to be located near the theoretical $n_e - \mu_{\text{Total}}$ curve. These findings led us to the conclusion that the unintentionally incorporated oxygen had an ionization rate as low as approximately $\sim 30\%$ and the unionized oxygen acted as neutral-impurity scattering centers that dominated electron transport in the ZnSnN$_2$ epitaxial films. The low $\mu_{\text{Total}}$ values in the ZnSnN$_2$-xO$_x$ epitaxial films resulted from the high concentration of neutral oxygen impurities. Frequent reported low electron mobilities in ZnSnN$_2$ epitaxial films can be explained by the same reason. Recent study showed that $\mu_{\text{Total}}$ of ZnSnN$_2$-xO$_x$ films with very high oxygen contents ($x > 0.4$) was low as $<5 \text{ cm}^2 \text{ V}^{-1} \text{ s}^{-1}$ [21]. That is, the more the oxygen content increases, the lower $\mu_{\text{Total}}$ becomes. This supports the idea that electron scattering by the oxygen-related defects governs the electron-transport properties of ZnSnN$_2$-xO$_x$.

Conclusions

In summary, we analyzed the electron-transport properties of unintentionally oxygen-doped ZnSnN$_2$ epitaxial layers. We confirmed that the incorporated oxygen impurities behaved as electron donors with the low ionization rate of 30%. The Hall- and Seebeck-effect measurements revealed that the compensated oxygen impurities, which are likely to form the electrically neutral Zn$_{30}$-2O$_N$ complexes, behaved as neutral-impurity scattering centers and further governed electron transport in the ZnSnN$_2$-xO$_x$. The low ionization rate led to the high concentration of neutral-impurity scattering centers. Therefore, we conclude that the low electron mobilities reported even in ZnSnN$_2$ single-crystalline films are attributed to the high concentration of neutral oxygen impurities. Suppression of the oxygen concentration in ZnSnN$_2$ is crucial not only to obtain nondegenerate ZnSnN$_2$, but also to achieve high mobility.

Methods

Thin-film growth

Reactive radio-frequency (RF) magnetron sputtering was employed to grow ZnSnN$_2$ epitaxial films on YSZ(111) single-crystalline substrates at the growth temperature of 300 °C. The base pressure of $\sim 2 \times 10^{-4}$ Pa was established prior to the film-growth. A disk of Zn$_{0.5}$Sn$_{0.5}$ alloy was used as a target (diameter of 1 cm and purity of 3 N). An RF power of 70 W was applied to the target. A mixture of Ar and N$_2$ gas with various N$_2$/(N$_2$+Ar) ratios was introduced into the chamber through two independent mass flow controllers with a total flow rate of 5 sccm. The total pressure in the growth chamber ($P$) was held at 2.0 Pa during film growth. The nitrogen partial pressure was defined as $P_{\text{N}_2} = f(N_2) \times P$. The growth time was adjusted to obtain films with thicknesses of 100–300 nm. Prior to the film growth, a 5 min-long sputter-etching with pure Ar gas was performed, followed by a 5 min-long pre-sputtering under the same condition with the film growth. After the film growth, the as-grown films were immediately stored in another vacuum chamber (the pressure was about 1 Pa) for subsequent measurements.

Characterization

A Rigaku ATX-G X-ray diffractometer with Cu Kα radiation was employed to perform out-of-plane (θ/ω) and in-plane (2θ/φ) scans to confirm the epitaxial growth. The compositions (Zn/(Zn+Sn) and x) of the ZnSnN$_2$ films were examined by XPS (PHI Versa Probe), using monochromated Al Kα (hv = 1486.6 eV) radiation. The XPS measurements were performed on 3-min Ar$^+$-sputter-etched surface of the films. The relative sensitivity factor (RSF) approach was exploited to determine the compositions. It was confirmed that the compositions determined by the RSF method were consistent with those determined by Rutherford backscattering spectrometry [29, 32]. Hence, we believe that the compositions in this study are sufficiently reliable. The details of the RSF approach were already described elsewhere [18]. Electrical properties were determined by Hall-effect measurements in the van der Pauw configuration (Toyo Corp. Resitest 8200). Optical transmittance and reflectance were collected between 0.3 and 5.0 μm using a UV–Vis–NIR spectrophotometer (Shimadzu UV-3150) and FTIR spectrometer (Shimadzu IRAffinity-1).

Fitting analysis of reflectance spectra

As seen from Fig. 2a–d, free-electron reflection is clearly seen in the IR region. The well-known Drude dielectric
model (Eqs. 1, 2) was employed to describe the optical response by the free electrons. In addition to the Drude model, the Tauc–Lorentz (TL) dispersion model was considered to describe the optical response across the whole spectral region. The explicit expression of the TL model is given in literature [25]. Deng et al. previously demonstrated that the dielectric response of ZnSnN₂ in the UV to visible region can be reproduced by double TL functions [13]. Hence, we modelled the dielectric function of degenerate ZnSnN₂−xOₓ as a sum of double TL and Drude functions. Theoretical reflectance spectra calculated via the Fresnel formulas combined with the dielectric function above were fitted to the experimental spectra. In the fitting analysis, the parameters in the TL functions were fixed for simplicity at constant values that were the same as the values reported by Deng et al. [13]. Thus, ωp and ΓD in the Drude function were used as the fitting parameters.

Additional files

Additional file 1. Typical out-of-plane and in-plane XRD patterns for ZnSnN₂ epitaxial films grown on YSZ (111).

Additional file 2. Oxygen 1s core spectra of ZnSnN₂ unintentionally doped with oxygen.

Additional file 3. Zn/(Zn + Sn) atomic ratio and x in ZnSnN₂−xOₓ.

Additional file 4. Out-of-plane XRD patterns for ZnSnN₂−xOₓ films with different x values and x-dependence of c-axis length.

Additional file 5. Electron density and mobility as functions of temperature and oxygen impurity concentration, x.

Additional file 6. Experimental reflectance and the best-fit curves using the TL and Drude combination model.

Additional file 7. Electron density and Drude parameters (plasma frequency, ωp, and broadening factor, ΓD in the Drude function) extracted from the best-fit spectra shown in Fig. 2a–e in the main text.

Additional file 8. "n" vs. "µ" data for ZnSnN₂−xOₓ and InN epilayers. The data for InN were taken from Refs. [26, 27].

Additional file 9. Theoretical mobility (τₚ vs. µ) and τₚ vs. S data shown in Fig. 4.

Abbreviations

µᵣ: Hall mobility; nₑ: electron density; Oₓ: impurity oxygen on nitrogen site in ZnSnN₂; x: oxygen content in ZnSnN₂−xOₓ; XPS: X-ray photoelectron spectroscopy; XRD: X-ray diffraction; ZnSn⁺: zinc atom on tin site in ZnSnN₂.

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Author contributions

NY and FK designed the experiments. XC performed fabrication and measurements. NY carried out the theoretical analysis. TT, YN and NY were the PIs. XC and NY wrote the manuscript. All authors discussed the results and reviewed the manuscript. All authors read and approved the final manuscript.

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Availability of data and materials

All data generated or analyzed during this study are included in Additional files 1–9.

Competing interests

The authors declare that they have no competing interests.

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