Tungsten-doped barium stannate as a transparent conducting film

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Near-infrared (NIR) transparent conducting films have recently garnered significant attention due to their relevance in various emerging devices. However, La-doped BaSnO3 (BSO) with high carrier mobility is hampered by the challenge of carrier reflection in the NIR region. In this study, we develop an approach to reconcile the conflict between conductivity and NIR transparency by incorporating W dopants into the Sn sites of BSO thin films. Our findings reveal that a 6% W-doped BSO thin film exhibits a remarkable carrier mobility of 26.6 cm2 V−1 s−1 (at a carrier concentration of 1.11 × 1020 cm−3) and achieves exceptional NIR transmittance of ~70% (at a wavelength of 2000 nm). This performance surpasses that of BSO films doped with other ions on the Sn sites. Furthermore, it is evidenced that the enhanced optoelectrical performance arises from the relatively lower effective mass of W ions, in contrast to Ta and Sb ions, when doped into BSO thin films. These outcomes provide valuable insights for the advancement of NIR transparent conducting films through the strategic incorporation of d-block ions into the Sn sites of stannate materials.

Introduction

Transparent conductive films (TCFs) have emerged as essential components in a wide array of modern technologies, ranging from touchscreens and displays to photovoltaic devices and smart windows.1 In recent years, the demand for TCFs with high transmittance in the near-infrared (NIR) range, coupled with excellent electrical conductivity, has gained substantial attention.2,3 This demand is driven by the growing recognition of the NIR region’s significance in numerous emerging fields, such as biomedical imaging, communication systems, and night-vision devices.4 Achieving high transmittance in this wavelength range is crucial for maximizing the efficiency of light-based processes, while maintaining excellent electrical conductivity ensures optimal device functionality.5

The current design strategy for state-of-the-art TCFs involves degenerately doping wide-bandgap semiconductors, such as Sn-doped indium oxide (ITO) and F-doped tin oxide (FTO), Al-doped zinc oxide (AZO), and cadmium oxide (CdO). The increase in electron concentration (n) leads to higher electrical conductivity (σ) according to the formula σ = neμ, where e represents the elementary charge and μ stands for carrier mobility.6 This strategy maintains transparency in the visible light range for TCFs. However, the transmittance shows a reduced tendency due to the inherent properties of these materials, including plasma resonances and free carrier absorption. These properties result in undesirable absorption and scattering of NIR light. Consequently, their performance in applications requiring efficient NIR transmission becomes compromised. Nevertheless, the development of NIR transparent electrodes is crucial for improving the efficiency of certain optoelectronic devices as well as opening up applications in other emerging areas, like NIR switching devices, NIR photodetectors, sensors and modulators for telecommunication.3,7

A high μ is a key parameter for ensuring high NIR transmittance and conductivity.8 In recent years, perovskite BaSnO3 (BSO) has exhibited a wide optical band gap along with a high μ, which has garnered significant attention for its transparent conducting performance.9,10 The promising indium-free and thermally stable TCFs have motivated researchers to extensively explore their optoelectronic properties. However, most studies have focused on rare-earth element doping at the Ba sites, leading to a significant decline in transmittance in the NIR region.11,12 Doping at the Sn site may be a judicious choice, as lower optical absorption coefficients have been observed in Sn site doping in BSO ceramics.13,14 Besides, an increased optical band gap was observed in Sb-doped BSO nanocrystals when compared to the La-doped BSO, which is beneficial for...
broadening a transparency window across the entire spectrum for BSO.\textsuperscript{15}

Studies on BSO thin films with dopants, typically p-block dopant Sb, at Sn sites have not been extensively investigated due to their inferior electrical transport properties when compared to dopants at the Ba sites.\textsuperscript{16–18} Kim et al.\textsuperscript{19} discovered that doping directly within the direct conduction pathway at Sn sites reduces the electrical properties of BSO single crystals due to increased lattice disorder and electron effective mass ($m^*$).

\hspace{0.5cm} Mizoguchi et al.\textsuperscript{14} reported the presence of mixed-valent antimony (Sb$^{5+}$/Sb$^{3+}$) in Sb-doped BSO ceramics. A significant decrease in $\mu$ to 6.18 cm$^2$ V$^{-1}$ s$^{-1}$, almost seven times lower, in BSO thin films was observed by Kim et al.,\textsuperscript{16} when transitioning the dopant from La at Ba sites to Sb at Sn sites. The low $\mu$ of Sb-doped BSO thin films has also been confirmed by Liu et al.,\textsuperscript{17} (1.75 cm$^2$ V$^{-1}$ s$^{-1}$) and Wei et al.,\textsuperscript{18} (3.05 cm$^2$ V$^{-1}$ s$^{-1}$). As a result, the low $\mu$ of Sb-doped BSO thin films remains one of the primary challenges in the field of NIR applications.

As a rule of thumb, the d states of d-block dopants can weaken the hybridization between the dopants and host cations, which is likely to reduce the $m^*$ and consequently enhance $\mu$.\textsuperscript{20} For instance, the Mo ion behaves as a highly charged donor in a +3 cation-based In$_2$O$_3$ system, demonstrating its efficacy as a dopant for achieving high $\mu$.\textsuperscript{21} Similarly, tungsten (W) with higher energy d orbitals has been shown to have limited hybridization with Sn 5s orbitals, resulting in a low $m^*$ and consequently high $\mu$ in W-doped SnO$_2$ TCFs.\textsuperscript{8} Furthermore, these d-block ion doped thin films have demonstrated favourable NIR transparency.\textsuperscript{22} According to the Drude model, achieving high $\mu$ and excellent NIR transparency in TCFs hinges on carefully moderating the carrier density. Ideally, d-block ions such as W$^{5+}$ and Mo$^{6+}$ offer two electrons when substituting Sn$^{4+}$ in the BSO system, whereas Ta$^{5+}$ contributes only one electron due to their oxidation states. Considering the higher d-state doping levels of W, coupled with the similar ionic radius of W ions and Sn ions, doping with tungsten is proposed as an effective approach to addressing the challenges of achieving high NIR transparency and improving mobility in BSO thin films when doping at the Sn sites. Furthermore, the effects of W doping on optoelectronic properties have never been investigated to date.

In this study, we report BSO thin films doped with W at the Sn sites to create visible-to-NIR TCFs. All films are deposited using a non-vacuum solution method, as in our previous works. We systematically investigate the effects of W doping content on microstructures, electrical transport, and optical transparency based on first-principles calculations. The optimized room-temperature $\mu$ of 26.6 cm$^2$ V$^{-1}$ s$^{-1}$ (at $n = 1.1 \times 10^{20}$ cm$^{-3}$) surpasses that of all other BSO thin films doped with various ions in Sn-sites. This improvement in $\mu$ can be attributed to the reduction in $m^*$ in the W-doped series compared to others such as Ta and Sb. Moreover, we also find that the NIR transparency reaches approximately 70% at a wavelength of 2000 nm while maintaining a low room-temperature resistivity of 2.1 m$\Omega$ cm.

**Experimental**

Density functional theory (DFT) was utilized in combination with the projector-augmented wave method implemented in VASP for first-principles calculations. The local density approximation (LDA) was used to describe the exchange–correlation energy. A plane wave cutoff energy of 500 eV is chosen. The Brillouin zone is sampled based on the Monkhorst–Pack scheme. We used $2 \times 2 \times 2$ grid sampling for the primitive cell and supercell of BaSnO$_3$, respectively. The lattice parameters and atomic coordinates are optimized until the forces on each atom are less than 0.01 eV Å$^{-1}$. The unit cell of BaSnO$_3$ consists of one Ba, one Sn, and three O atoms. To obtain the geometric structure of the 6% W-doped system BaSnO$_3$, we initially established a $4 \times 2 \times 2$ supercell of BaSnO$_3$ as a parent cell. Then, one impurity element W was substituted for one Sn atom, resulting in the doped geometric structure. The doping concentration of this supercell is 0.0625, which closely approximates the experimental impurity concentration of 0.06.

Epitaxial W-doped BaSnO$_3$ (BSWO) thin films were deposited onto (001)-oriented SrTiO$_3$ (STO) substrates. Stoichiometric barium acetate Ba(CH$_3$COO)$_2$ (99.99%), tin chloride hydrate SnCl$_2$·5H$_2$O (99.99%), and tungsten chloride WCl$_6$ (VI, 99.9%) were dissolved in a mixed solvent of ethylene glycol and glacial acetic acid. The precursor solutions (cation concentration of 0.4 mol L$^{-1}$) were stirred for 10 minutes at 70 °C and then for 6 hours at room temperature. The spin-coating procedure was performed on a spin-coater at a spin rate of 7000 rpm for 30 seconds under an ambient temperature of around 50°C. The wet thin films underwent pyrolysis at 350°C for 10 minutes and were then crystallized at 1000°C for 15 minutes. These steps were repeated multiple times to increase the film thickness to about 370 nm. The W-doping content in the prepared BSO thin film varied from 0 to 12%. The same procedure was applied to LaAlO$_3$ (001) substrates to prepare W, Ta (tantalum ethoxide, TaCl$_5$·H$_2$O$_5$), and Sb (antimony acetate, Sb(CH$_3$COO)$_3$)-doped BSO thin films for detecting the optical band gap.

The crystal structure and orientation of the prepared thin films were identified using high-resolution X-ray diffraction (HR-XRD). The epitaxial growth relationship for the derived BSWO thin film on the STO substrate was determined through X-ray φ-scans. Rocking curves (out-of-plane measurements) for the BSWO thin films were conducted using 2θ/θ scans. Changes or variations in crystal orientations and in-plane lattice constants were examined through reciprocal space mappings (RSMs, in-plane measurements). The valence states of the elements were investigated using X-ray photoelectron spectroscopy (XPS) measurements with an Al-Kz anticathode within an ultrahigh vacuum chamber. The binding energy data were calibrated to the C 1s peak at 284.8 eV. Surface morphology was characterized using an atomic force microscope (AFM). Optical transmittance spectra (6000–250 nm) were measured using a UV-Vis-NIR spectrometer. Electronic transport properties were determined through the standard four-probe method and van der Pauw geometry using a physical property measurement system (PPMS).
Results and discussion

The geometric and band structures of BaSnO$_3$ and 6% W-doped BaSnO$_3$ were calculated. The optimized lattice constant of BaSnO$_3$ (~4.11 Å) is consistent with the experimental value (4.117 Å) and also agrees with the data from other theoretical calculations, demonstrating the reliability of our computational method. Moreover, the band structure of the W doping system indicates that it transforms into heavily doped semiconductors in Fig. 1(b), and W doping provides electron states right inside the conduction band of BaSnO$_3$. Here, the band gap value (~1.00 eV) obtained using the LDA functional is smaller than the experimental values [3.10–3.60], which can be explained by the underestimated band gap of semiconductors due to the inherent limitations of the approximation method. This theoretical result validates that doping W elements in the BaSnO$_3$ system is an effective method to obtain elevated conductivity.

Following the theoretical results, a series of W-doped BaSnO$_3$ thin films were prepared. Fig. 2(a) illustrates the c-axis lattice constant in relation to the W doping content. The BSWO thin films exhibit a decreasing trend in lattice constant as the W doping content increases. Due to the constraint of solid solubility, the lattice constant reaches saturation when W doping exceeds 6%. The rocking curves of the (200)-peak for all prepared thin films with varying W-doping contents are depicted in Fig. 2(b). The low full width at half maximum (FWHM) values, ranging from 0.36° to 0.13°, signify the favourable quality of the prepared thin films. This value demonstrates a declining trend with increasing W-doping contents. This reduction can be attributed to the potential decrease in crystallization temperature associated with higher doping content, similar to observations in Sb-doped BSO. Furthermore, the decrease in lattice parameters that induce optimized lattice mismatch between BSWO and STO (3.91 Å) is another factor influencing the variation in rocking FWHM.

Fig. 2(c) presents the X-ray φ-scans of the (110)-plane for the parent and 6% W-doped BSO thin films and the substrate, shedding light on the epitaxial growth mode of the prepared BSWO thin films. A set of four distinct peaks with 90° separation is clearly visible for the prepared thin film, indicating that the epitaxial relationship follows BSO(001)[110]||STO(001)[110]. The epitaxial growth mode of the BSO film on the STO substrate is also determined through TEM, as depicted in Fig. 2(d). The result of selected-area electron diffraction (SAED) confirms the conclusion drawn from the φ-scans. Moreover, the high crystalline quality of the prepared thin films is also confirmed by the X-ray RSM, where the asymmetric (103) diffraction spot of BSO and the (103) diffraction spot of STO are displayed in Fig. 2(e) and (f). The red line indicates that the prepared thin film is fully relaxed and exhibits incoherent epitaxial growth. These observations suggest that high quality can be certainly achieved using an easy non-vacuum process.

Fig. 2(g) and (h) display the surface morphology of the typical matrix and 6% W-doped BSO thin films. It is evident that both films exhibit uniform and dense surface morphologies with no discernible microcrack. The prepared thin films demonstrate a low surface root-mean-square (RMS) roughness of 7–9 nm, which aligns with the results of our previously optimized solution process for depositing BSO thin films.

To elucidate the chemical compositions and bonding states, XPS measurements of Ba, Sn, W, and O elements of three typical thin films were detected as presented in Fig. 3. Ba 3d$_{3/2}$ and Ba 3d$_{5/2}$ spin–orbit splitting components are located at about 794.3 eV and 779.1 eV, respectively, indicating the valence state of Ba ions is +2 within Ba–O bonds in the prepared thin films. Sn 3d consists of doublet peaks of Sn 3d$_{5/2}$ and Sn 3d$_{3/2}$ and the peak differences between these are 8.4 eV for all the prepared thin films, which is related to the dominated Sn$^{4+}$ state. As shown in Fig. 3(c) and (d), the W core-level XPS spectra of 6% W:BSO and 12% W:BSO samples are characterized by W$^{6+}$ species at 35.6 ± 0.1 eV and W$^{5+}$ species at 33.0 ± 0.1 eV. The decreased lattice constant after doping W element can be further pinpointed by the smaller W$^{5+}$ (0.62 Å) and W$^{6+}$ (0.60 Å) as dopants for Sn$^{4+}$ (0.69 Å) and they prominently offer electron carriers in the BSWO system. With an increase in W-doping content, the concentration of W$^{6+}$ species rises. Thus, it is reasonable to assume that WO$_x$ contributes to the saturated lattice constant after the W doping content reaches 6% (Fig. 2(a)). The insulating WO$_x$ phase could potentially lead to a degradation in electrical transport properties.

The O 1s spectra in Fig. 3(d) could be fitted by three peaks at 529.6 ± 0.1, 531.3 ± 0.1, and 532.4 ± 0.1 eV using Gaussian distribution, which is assigned to the lattice oxygen (O$_{\text{lattice}}$), chemisorbed water (O–H), and adsorbed oxygen (O$_{\text{ads}}$), respectively. Based on the fitting results, the area ratio of the O$_{\text{lattice}}$ peak is 41.7%, 22.3%, and 20.8% for the matrix, 6% W, and 12% W-doped BSO thin films. This indicates that oxygen vacancy increases with increasing W doping content, since the lattice oxygen decreased. As reported by Yoon et al., the oxygen vacancy-assisted relaxation process that reduces the density of threading dislocations is beneficial for increasing electron mobility in BSO epitaxial thin films. Herein, the declining trend for FWHM with increasing W-doping contents in Fig. 2(b) may also be due to the enhanced oxygen vacancy. Additionally, the O 1s spectra for the 6% Ta-doped BSO thin films is also given in Fig. 3(d). The Ta dopant induces a similar oxygen vacancy (24.6% for O$_{\text{lattice}}$) as compared with W.
Fig. 4(a) shows the temperature dependent resistivity of the matrix and W-doped BSO thin films from 300 K to 2 K. As shown, the 2% W:BSO thin film displays a semiconducting behaviour with a negative temperature coefficient of resistivity in the whole measured temperature range. A characteristic of metal-to-insulator transition appears around 230 K when the W doping content increases to 4%. This transition further decreases to $\sim$ 100 K for the thin film of the 6% W-doped BSO sample, indicating that more free carriers contribute to the conducting behaviour in the BSWO thin film. Corresponding to the saturated W doping content (Fig. 2(a)), the resistivity shows an increasing trend for the 8% W- and further 12%-doped BSO samples. These results suggest that the W doping limit for the BSO at Sn-sites is around 6% consistent with the saturate lattice constant in Fig. 2(a), and the optimized room-temperature resistivity is 2.11 m$\Omega$ cm.

Temperature dependent resistivity behaviours of the 2% W- and 6% W-doped BSO thin films are analysed (Fig. 4(b) and (c)), to further explore the conduction mechanism. As shown, the $\rho$–$T$ curve can be fitted with a fluctuation inducted tunnelling pattern: $\rho = A \times \exp(-T/T_0 + 2T_1)$ (Fig. 4(b)), where $A$ is a constant, and $T_0$ and $T_1$ are related to the properties of the insulating barriers. The 6% W doped thin film exhibits a positive temperature coefficient of resistivity in the range of...
300–100 K, which can be fitted with the Bloch–Grüneisen mode: \[ \rho^{-1} = \rho_p^{-1} + (\rho_o + \rho_{ph})^{-1}, \]
where \( \rho_p = \rho_1 \coth(\theta_k/2T)[1 + (2/3)\sinh^2(\theta_k/2T)]^{-1} \), where \( \theta_k \) is the Einstein temperature, \( \rho_o \) is the parallel resistivity, \( \rho_1 \) is the residual resistivity and \( \rho_{ph} \) is the scattering by ionized dopants, as also observed in previous studies. As plotted in the inset of Fig. 4(c), the room-temperature resistivity at low temperature can be fitted by the three-dimensional variable-range-hopping (3D-VRH) model: \[ \rho = \rho_0 \exp(T_s/T)^{1/4} \]
where \( T_s \) depends on the density of localized states at the Fermi level and the falloff rate of the wave functions associated with these states, suggesting the presence of gap states in the vicinity of the Fermi level. The linear relationship between \( \ln(\rho) \) and \(-T^{1/4}\) demonstrates the dominant mechanism at low temperatures is weak localization. With further increased W-doping content, the resistivity shows an increment and ultimately displays a semiconducting behaviour (such as 12% W-doped BSO), which stems from the enhanced lattice disorder and impurity scattering.

Hall measurements were carried out for the BSWO thin film, and the calculated \( n \) and \( \mu \) are displayed in Fig. 4(d) and (e). As shown, the \( n \) shows a similar temperature independent tendency for each BSWO thin film with decreasing temperature, suggesting that all the samples are in a degenerately doped semiconducting regime.\(^{10,12}\) The slight decrement of \( n \) in the low temperature range is largely attributed to the emerging free-out ionized dopants, as also observed in previous studies. Moreover, the \( n \) with the level of \( \sim 10^{20} \text{ cm}^{-3} \) shows as expected an increasing trend with respect to the increasing of W-doping content, indicating that free carriers from W ions certainly contribute to the conducting path for the BSWO thin films.

It is shown in Fig. 4(e) that \( \mu \) increases systematically with decreasing temperature to 50 K and then shows a subtle decrease at a lower temperature, in accordance with the electrical conduction mechanism in La doping at Ba sites in the BSO system. Additionally, the \( \mu \) increases with the increasing W-doping content up to 6% and shows a decline for further W doping, which can be elucidated by the Brooks–Herring theory.\(^{31}\) The initially improved \( \mu \) may be caused by the screened impurities scattering due to increased \( n \). However, the scattering by ionized dopants turns out to be dominated by higher \( n \) after the W-doping content reaches 6%, which hampers the increment of \( \mu \). It is important to emphasize here that the room-temperature \( \mu \) occupies the top level in various doping ions at Sn sites for BSO thin films, which is compared in Fig. 4(f).\(^{16–18,32}\)

Optical transmittance, as presented in Fig. 5(a), was detected ranging from 250 to 6000 nm for the prepared BSWO thin films. In comparison, the transmittance of our previous (Ba,La)SnO\(_3\) (BLSO),\(^{33}\) Ba(Sn,Sb)O\(_3\) (BSSO),\(^{18}\) and Sn-doped In\(_2\)O\(_3\) (ITO)\(^{34}\) thin films are also introduced. The overall transmittance shows a depressing tendency with the increment of W-doping content owing to the enhanced carrier reflection. Higher transmittance over a broad range of wavelengths in NIR regions for BSWO thin film is maintained as compared to BLSO and ITO thin films. According to the screened plasma frequency\(^6\) \( \omega_p = (e/\sqrt{\varepsilon_0 \varepsilon_r}) \sqrt{n/e^2} \) (where \( e \) is the elemental charge, \( \varepsilon_0 \) is the permittivity of vacuum, and \( \varepsilon_r \) is the relative permittivity of the material), a lower carrier concentration \( (n) \) results in a lower \( \omega_p \), which is beneficial for obtaining higher NIR transparency. Plasmon excitations that obey the Drude model in BSWO thin films display a much lower absorbance in the NIR regions as compared to BLSO, BSSO, and ITO thin films due to a lower
carrier concentration of BSWO.\textsuperscript{14} This gives an explanation to the higher NIR transparency in BSWO thin films.

Fig. 5(b) depicts the average optical transmittance of thin films in NIR portions and transmittance in the wavelength of 2000 nm. The average transmittance over the NIR region from 780 to 2500 nm is calculated using:  
\[
T_{\text{average}} = \frac{\int_{780}^{2500} T(\lambda) d(\lambda)}{\int_{780}^{2500} d(\lambda)}.  
\]
Notably, the transmittance of BLSO and ITO thin films show a striking decline in NIR portions and reduce to 41.51\% and 15.27\% in 2000 nm, respectively, while BSWO thin films maintain a flat tendency and pleasurable transmittance to 2000 nm (~70\% for 6\% W-doped BSO).

To elucidate the improved $\mu$ as well as NIR transmittance of the W-doped BSO thin films, in-depth analysis needs to be conducted. Here, the direct bandgap of doped BSO thin films subjected to different dopants (4\% W, 6\% W, 4\% Ta, 6\% Ta, 6\% Sb, and 9\% Sb) was extrapolated from an optical transmittance spectrum. According to the Tauc’s relationship:\textsuperscript{19}  
\[
(zh)^2 = A(h\nu - E_g)  
\]
where $z$ is the absorption coefficient, $h$ is Planck’s constant, $\nu$ is the incident photon frequency, and $A$ is an ordinary constant, the optical direct bandgap ($E_g$) can be estimated by the $z^2$ vs. photon energy $h\nu$ curves. As shown in Fig. 5(c), the $E_g$ values of the BSO thin films for bare, W-, Ta-, and Sb-doped BSO thin films are 3.46, 3.71, 3.65, and 3.73 eV, respectively. The difference in optical direct gap ($\Delta E$) for doped BSO thin films as compared to bare BSO can extract the effective mass ($m^*$) experimentally in accordance with the Burstein–Moss shift:\textsuperscript{19}  
\[
\Delta E = h^2 (3n^2/8m^*)^{1/3}  
\]
Fig. 5(d) gives the resultant $\Delta E$ vs. $n^{2/3}$ curves of the prepared thin films. Fitting from the linear slope of $\Delta E$ vs. $n^{2/3}$, the $m^*/m_0$ for the W-, Ta-, and Sb-doped BSO thin films is estimated to be ~0.37, 0.39, and 0.73, respectively. It is noted here that d-block dopants (W, Ta) doped BSO thin films exhibit similar $m^*$ and show an obviously lower $m^*$ as compared with p-block dopants (Sb). Considering the above, it is plausible to elucidate the superior carrier mobility of W-doped BSO thin films by the lower $m^*$ as compared with Sb-doped BSO. This explanation applies equally to Ta-doped BSO thin films, which is also the d-block ions.

**Conclusions**

In summary, conducting W-doped BaSnO\textsubscript{3} has been proved by the results of both theoretical calculation and experiment. Epitaxial BSWO thin films with promising mobility and Vis-NIR transparency were deposited using a solution deposition method. The improved electrical transport properties were achieved in a film of 6\% W-doped BSO with a room-temperature resistivity of 2.11 m$\Omega$ cm and mobility of 26.61 cm$^2$ V$^{-1}$ s$^{-1}$. Moreover, all BSWO thin films exhibit improvement in transmittance as compared with other Sn site doping studies in both visible and NIR regions. Plasmon excitations that obey the Drude model in BSWO thin film with...
a lower carrier concentration display a much lower absorbance in the NIR regions. It is noteworthy that d-block dopant (W and Ta) doped BSO thin films behave similarly with a smaller effective mass compared with p-block dopants (Sb) doped. The low effective mass of BSWO eventually improves the electrical transport properties. This work suggests that the d-block dopant at Sn sites is an effective route to obtain stannate films with excellent opto-electronic performance, and the W-doped BSO is a potential candidate for NIR transparent conducting films.

Author contributions

Yuandi He: conceptualization; methodology; original draft; review and editing. Yao-long Kang: software; original draft. Renhui Wei and Peng-Lai Gong: review and editing; data curation. Ling Hu and Wenhai Song: formal analysis. Xuebin Zhu: review and editing; validation. Yuting Sun: project administration.

Conflicts of interest

There are no conflicts to declare.

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References