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# Improved self-powered perovskite CH<sub>3</sub>NH<sub>3</sub>PbI<sub>3</sub>/SnO<sub>2</sub> heterojunction photodetectors achieved by interfacial engineering with a synergic effect

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Lead halide perovskite heterojunctions have been considered as important building blocks for fabricating high-performance photodetectors (PDs). However, the interfacial defects induced non-radiative recombination and interfacial energy-level misalignment induced ineffective carrier transport severely limit the performance of photodetection of resulting devices. Herein, interfacial engineering with a spin-coating procedure has been studied to improve the photodetection performance of CH<sub>3</sub>NH<sub>3</sub>PbI<sub>3</sub>/SnO<sub>2</sub> heterojunction PDs, which were fabricated by sputtering a SnO<sub>2</sub> thin film on ITO glass followed by spin-coating a CH<sub>3</sub>NH<sub>3</sub>PbI<sub>3</sub> thin film. It has shown that spin-coating of a SnO<sub>2</sub> layer on the sputtered SnO<sub>2</sub> thin films suppressed the surface oxygen vacancies of SnO<sub>2</sub> thin films and up-shifted their conduction band, which suppressed the interfacial non-radiative recombination and enhanced the carriers transport at the CH<sub>3</sub>NH<sub>3</sub>PbI<sub>3</sub>/SnO<sub>2</sub> interface, respectively. Accordingly, improved photodetection performance, such as the reduced dark current and increased photocurrent, has been observed in the CH<sub>3</sub>NH<sub>3</sub>PbI<sub>3</sub>/SnO<sub>2</sub> heterojunction PDs, where the responsivity and detectivity of 0.077 A W<sup>-1</sup> and 2.0 × 10<sup>11</sup> jones, respectively, at the zero bias have been demonstrated. These results show a simple way to suppress the interfacial non-radiative recombination and enhance the carrier transport at the interface to fabricate improved perovskite heterojunction PDs in the future.

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## Introduction

Photodetectors (PDs), which convert photons into electric charge, have attracted much attention recently owing to their important applications in various areas including digital imaging, optical communications, medical analysis, and environment monitoring.<sup>1–4</sup> In general, PDs can be classified into two categories, *i.e.*, photoconductors and photodiodes. The former ones possess a simple device structure but feature a high noise current and low photoconductive gains.<sup>5</sup> In contrast, the latter ones have usually been characterized with a low noise current and high photon-to-electric conversion efficiencies owing to the presence of an interfacial barrier and a strong built-in electric field.<sup>6,7</sup> Additionally, the photodiodes can operate without the requirement of a power supply. Heterojunctions, which are composed of two different semiconductors bound together, have been considered as important building blocks for photodiodes owing to simple fabrication process regarding free-of-controlled n- and p-type doping of selected semiconductors. In recent years, lead halide perovskites (LHPs) with a formula of ABX<sub>3</sub> (where A is the CH<sub>3</sub>NH<sub>3</sub><sup>+</sup> (MA<sup>+</sup>), and

HC(NH<sub>2</sub>)<sub>2</sub><sup>+</sup> (FA<sup>+</sup>); B is the Pb<sup>2+</sup>; X is the Cl<sup>-</sup>, Br<sup>-</sup>, and I<sup>-</sup>) have attracted much attention from the optoelectronics community due to their outstanding properties such as long charge carriers lifetime, long charge carriers diffusion length, high absorption coefficient, widely tunable energy bandgap, and low-temperature solution processability.<sup>8</sup> Benefiting from these advantages, LHP heterojunctions have been considered as promising candidates for achieving efficient photodetection.<sup>9–12</sup>

Generally, the working principles of PDs are similar to those of solar cells (SCs), where excess carriers are generated due to the valance-band-to-conduction-band transition upon light illumination, followed by the collection of excess carriers by the electrodes. As a wide bandgap semiconductor, tin dioxide (SnO<sub>2</sub>) features a high electron mobility, deep conduction band, and low-temperature processability, and has widely been employed as an electron transport layer to fabricate efficient SCs recently.<sup>13–15</sup> Meanwhile, various PDs have recently been demonstrated based on LHP/SnO<sub>2</sub> heterojunctions.<sup>16–18</sup> Similar to LHP SCs, the performance of LHP PDs is mainly related to the collection efficiencies of excess carriers, which are limited by the trap-assisted non-radiative recombination and ineffective carrier transport.<sup>19</sup> In heterojunction PDs, the former is mainly related to the bulk and interfacial defects, while the latter is owing to the energy-level misalignment at the heterojunction interface. Notably, various strategies have been explored to improve the crystalline quality of LHP thin films recently, which

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facilitated the rapid development of power conversion efficiency (PCE) of LHP SCs.<sup>20</sup> Typically, doping of elements has resulted in an increase in grain sizes and a suppression of intrinsic defects (such as halide and Pb vacancies) in LHP thin films and nanocrystals.<sup>21</sup> Besides, interfacial engineering with a thin dielectric layer and anchoring groups have been demonstrated but primarily to suppress the interfacial non-radiative recombination.<sup>22</sup> In contrast, the alignment of energy-level at the heterojunction interface has seldom been explored before. Practically, both interfacial passivation and energy-level alignment are fundamental for fabricating high-performance heterojunction PDs.<sup>23</sup> Herein, we showed the fabrication of improved solution-processed MAPbI<sub>3</sub>/SnO<sub>2</sub> heterojunction PDs by employing a stacked SnO<sub>2</sub> thin film prepared by sputtering and spin-coating methods in sequence. It showed that the subsequent spin-coating procedure behaved a synergic effect including a suppression of interfacial defects and alignment of the energy-level at the MAPbI<sub>3</sub>/SnO<sub>2</sub> interface. Accordingly, a decreased dark current and an increased photocurrent have been demonstrated simultaneously in the MAPbI<sub>3</sub>/SnO<sub>2</sub> heterojunction PDs. Finally, improved performance of photo-detection with the responsivity and detectivity as high as 0.077 A W<sup>-1</sup> and 2 × 10<sup>11</sup> Jones, respectively, at the zero bias have been evidenced.

## Experiments

MAPbI<sub>3</sub>/SnO<sub>2</sub> heterojunctions (Fig. 1b) were fabricated by spin-coating a MAPbI<sub>3</sub> thin film onto a stacked SnO<sub>2</sub> thin film, which was prepared by sputtering a SnO<sub>2</sub> thin film on cleaned ITO glasses then followed by spin-coating a thin SnO<sub>2</sub> layer. In detail, the ITO glasses were cleaned with acetone, ethanol and DI water in sequence. After that, a thin SnO<sub>2</sub> film was firstly grown using a sputtering method with a growth temperature of 500 °C and growth time of 10 min (labeled as SnO<sub>2</sub>-1). The RF power is 80 W. The Sn source was provided by a SnO<sub>2</sub> target (99.99%). Ar/O<sub>2</sub> with a flow of 55 sccm/5 sccm were used as working gas. Then, a thin SnO<sub>2</sub> layer was grown by spin-coating to obtain the stacked SnO<sub>2</sub> thin films (labeled as SnO<sub>2</sub>-2), in which the SnO<sub>2</sub> precursor solution (3 wt%) was prepared by diluting SnO<sub>2</sub> nanoparticles (~20 nm) into DI water. Finally, MAPbI<sub>3</sub> thin films were grown on SnO<sub>2</sub>-2 (labeled as MAPbI<sub>3</sub>-2) according to previous reports.<sup>24</sup> The perovskite precursor was prepared by dissolving PbI<sub>2</sub> and MAI into DMF/GBL/DMSO mixtures. The perovskite precursor was spin-coated with a two-step program (1000 rpm for 10 s and 5000 rpm for 20 s), where 400 μL toluene was dropped at 8 s of the second program. To reveal the roles of the spin-coated SnO<sub>2</sub> thin layer, MAPbI<sub>3</sub> thin films were directly prepared onto SnO<sub>2</sub>-1 (labeled as MAPbI<sub>3</sub>-1), as shown in Fig. 1a. The PDs were obtained by depositing a thin Au electrode with an area of 0.01 mm<sup>2</sup> on the MAPbI<sub>3</sub> thin films and an Ag paste on ITO.

The morphology and structure of SnO<sub>2</sub> and MAPbI<sub>3</sub> thin films were studied by field-emission scanning electron microscopy (SEM, JSM-7800F) and X-ray diffraction (XRD, Bruker D8 with Cu Kα radiation of 1.54 Å). The steady-state photoluminescence (PL) spectra of MAPbI<sub>3</sub> thin films were obtained

with the excitation of a cw 405 nm laser. The power of the incident laser was modulated by an attenuator. The transient PL spectra were excited with a ps-pulse laser at 450 nm, and the data were recorded by a ps time-correlated single photon counting technique. The absorption spectra were collected on an ultraviolet-visible spectrophotometer. The atomic electron binding energies of SnO<sub>2</sub> thin films were studied by X-ray photoelectron spectroscopy (XPS). The work function and energy band structure of SnO<sub>2</sub> were studied by ultraviolet photoelectron spectroscopy technique (UPS) using He I as the excitation source.

The performance of MAPbI<sub>3</sub>/SnO<sub>2</sub> heterojunction PDs was studied by a photoresponse system consisting of a Xe lamp, a monochromator, an electronic shutter, two probes, a microscope, a Keithley 2401 source-measure unit (SMU), and a semiconductor parameter analyzer (Keysight, B1500A). All experiments were carried out in the atmosphere with a humidity of ~55–60%.

## Results and discussions

Fig. 1c and d exhibit the SEM images of SnO<sub>2</sub>-1 and SnO<sub>2</sub>-2 thin films. It can be seen that the SnO<sub>2</sub>-1 thin films feature a smooth surface (Fig. 1c). This can be attributed to the high growth temperature (500 °C, as mentioned above), which facilitated the migration of species at the substrate surface. In comparison, a rough surface has been observed in the SnO<sub>2</sub>-2 thin films (Fig. 1d). According to previous report,<sup>25</sup> a rough surface of buried layer increases the contact area at the MAPbI<sub>3</sub>/SnO<sub>2</sub> interface, which improves the optical absorption in the MAPbI<sub>3</sub> thin films. XRD were carried out to investigate the crystalline structures of SnO<sub>2</sub>-1 and SnO<sub>2</sub>-2 thin films, and the results are illustrated in Fig. 1i. As depicted, both the SnO<sub>2</sub> thin films exhibit a similar crystalline structure. According to the XRD peaks which are located at 22.3°, 33.5°, 36.8°, and 52.5°, both the SnO<sub>2</sub> thin films were crystallized into tetragonal with a space group of *P4<sub>2</sub>/mnm*.

Fig. 1e and f exhibit the SEM images of MAPbI<sub>3</sub> thin films grown on SnO<sub>2</sub>-1 and SnO<sub>2</sub>-2. It shows that both the MAPbI<sub>3</sub> thin films have a dense structure, which is fundamental for the subsequent fabrication of heterojunction PDs. Notably, the grain sizes of MAPbI<sub>3</sub> thin films grown on SnO<sub>2</sub>-2 are around 200 nm, which is slightly larger than those grown on SnO<sub>2</sub>-1 (150 nm). In general, the MAPbI<sub>3</sub> thin films with a larger grain size possess a decreased density of boundary defects, which is favorable to improve the photodetection performance of resulting PDs.<sup>26,27</sup>

Fig. 1j shows the XRD patterns of MAPbI<sub>3</sub> thin films grown on SnO<sub>2</sub>-1 and SnO<sub>2</sub>-2. It shows that both the MAPbI<sub>3</sub> thin films exhibit a similar crystalline structure and quality. Accordingly, the spin-coated SnO<sub>2</sub> layer did not affect the growth of MAPbI<sub>3</sub> thin films. Further observations show that the contact angle of SnO<sub>2</sub>-2 is comparable to that of SnO<sub>2</sub>-1 (Fig. 1g and h), which is responsible for the similar crystalline quality of the two MAPbI<sub>3</sub> thin films.

As mentioned above, the SnO<sub>2</sub>-2 thin films were grown by sputtering and spin-coating SnO<sub>2</sub> layers in sequence. In the



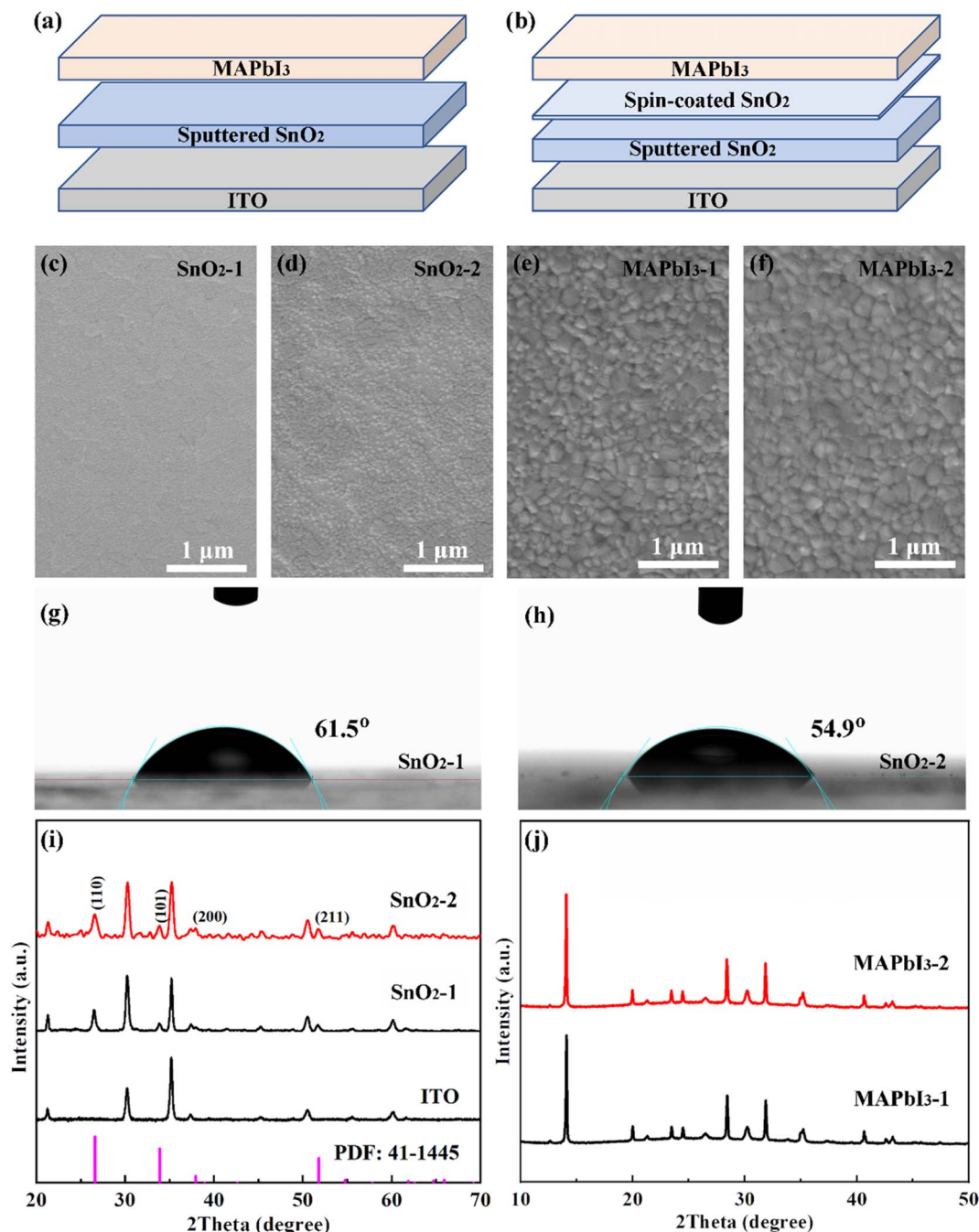


Fig. 1 Characteristics of SnO<sub>2</sub> and MAPbI<sub>3</sub> thin films. (a and b) Scheme of MAPbI<sub>3</sub>/SnO<sub>2</sub>-1 and MAPbI<sub>3</sub>/SnO<sub>2</sub>-2 heterojunctions. (c–f) SEM images of SnO<sub>2</sub> and MAPbI<sub>3</sub> thin films. (g and h) Water contact angles of SnO<sub>2</sub> thin films. (i and j) XRD patterns of SnO<sub>2</sub> and MAPbI<sub>3</sub> thin films.

latter case, the SnO<sub>2</sub> precursor was prepared by diluting the SnO<sub>2</sub> nanoparticles in DI water, therefore oxyhydroxyl groups (OH<sup>-</sup>) tended to be doped into the SnO<sub>2</sub> thin films.<sup>28</sup> To verify this, XPS measurements were performed, and the results are shown in Fig. 2. As depicted in Fig. 2a, only Sn and O signals were observed in both the SnO<sub>2</sub> thin films. This suggests that no impurities were incorporated during the preparation processes. Fig. 2b exhibits the O 1s core-level spectra of SnO<sub>2</sub>-1 and SnO<sub>2</sub>-2. The binding energies are observed at around 530.5 eV, 531.3 eV, and 532.5 eV, which are associated with Sn–O, oxygen vacancies

(V<sub>O</sub>), and OH<sup>-</sup>, respectively.<sup>29</sup> As reported, the content of various components can be qualitatively evaluated from their intensities of XPS peaks.<sup>29</sup> Accordingly, the SnO<sub>2</sub>-2 shows an increased content of OH<sup>-</sup> and a reduced content of V<sub>O</sub> compared to SnO<sub>2</sub>-1. The former is related to the oxyhydroxyl groups in the aqueous solution. The reasons for latter are unclear currently, but should be related to the SnO<sub>2</sub> nanoparticles used. Importantly, it has reported that surface treatment with oxyhydroxyl groups tuned the energy band of buried layers.<sup>30</sup> As shown later, the SnO<sub>2</sub>-2 thin films show an increased conduction band, which aligned the



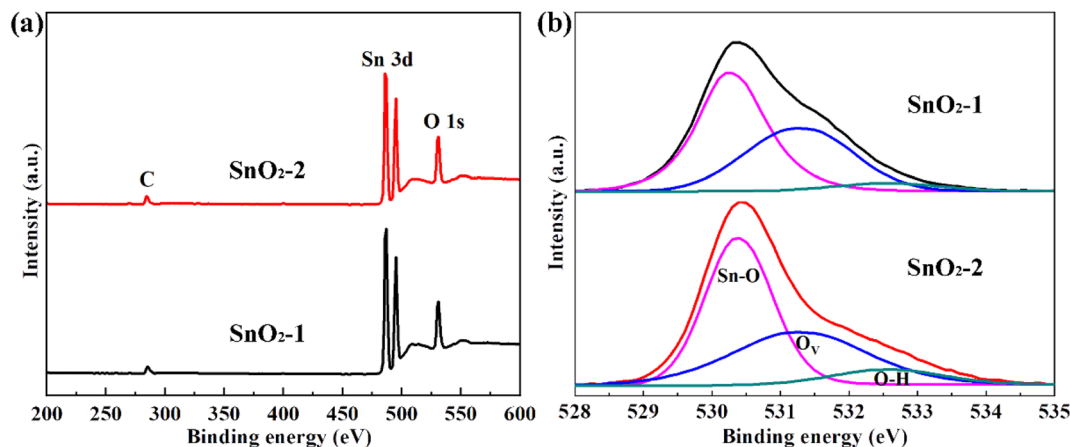


Fig. 2 XPS curves of SnO<sub>2</sub> thin films. (a) XPS survey spectra. (b) O 1s core-level spectra.

energy-level at the MAPbI<sub>3</sub>/SnO<sub>2</sub> interface. Meanwhile, the reduced V<sub>O</sub> suppressed the non-radiative recombination at the MAPbI<sub>3</sub>/SnO<sub>2</sub> interface.<sup>31</sup>

Fig. 3 shows the steady-state and time-resolved PL spectra of MAPbI<sub>3</sub> thin films grown on SnO<sub>2</sub>-1 and SnO<sub>2</sub>-2. It can be seen that the PL spectra captured from the front surface of MAPbI<sub>3</sub> thin films exhibit similar intensity and lifetime (Fig. 3a and c), showing that both MAPbI<sub>3</sub> thin films possess similar crystalline quality. This is in consistency with the XRD and SEM results shown above. Notably, the PL spectra captured from the back surface of MAPbI<sub>3</sub> thin films exhibit distinct behaviors. As can be seen in Fig. 3b and d, the MAPbI<sub>3</sub> thin films prepared on SnO<sub>2</sub>-2 exhibit a decreased PL intensity and lifetime. This suggests that the spin-coated SnO<sub>2</sub> layer enhanced the extraction of electrons at the MAPbI<sub>3</sub>/SnO<sub>2</sub> interface. Similar phenomena have been reported in graphdiyne decorated MAPbI<sub>3</sub>/SnO<sub>2</sub> and Mg<sup>2+</sup> decorated MAPbI<sub>3</sub>/NiO heterojunctions.<sup>32–34</sup> As shown below, the slight up-shift of the conduction band of the SnO<sub>2</sub>-2 thin films aligned the interfacial energy-level, which is responsible for the decreased PL intensity and lifetime in the back surface of MAPbI<sub>3</sub>/SnO<sub>2</sub>-2 heterojunctions.

Additionally, the interfacial defects at the MAPbI<sub>3</sub>/SnO<sub>2</sub> interfaces were further studied from the excitation power dependent integrated PL, and the results are shown in Fig. 3e and f. As depicted in Fig. 3e, the PL spectra of both MAPbI<sub>3</sub> thin films exhibit similar patterns with an increase in excitation power, showing that the recombination processes of excess carriers kept unchanged. Fig. 3f depicts the relationship between integrated PL intensity and corresponding excitation power. According to the trap-filling model, the PL intensity increases slowly with an increase in excitation power under the low excitation level, while it increases rapidly with the excitation power under the high excitation level.<sup>17,35,36</sup> As a result, the density of trap states ( $n_{\text{trap}}$ ) is proportional to the excitation power threshold of complete trap filling ( $P_{\text{th}}$ ), *i.e.*,  $n_{\text{trap}} \propto P_{\text{th}}$ .<sup>17,35,36</sup> As shown in Fig. 3f, the  $P_{\text{th}}$  for MAPbI<sub>3</sub> grown on SnO<sub>2</sub>-1 and SnO<sub>2</sub>-2 were observed at 9.2 μW and 10.9 μW, respectively. This suggests that around 10% trap states have been reduced at the MAPbI<sub>3</sub>/SnO<sub>2</sub>-2 interface compared to MAPbI<sub>3</sub>/SnO<sub>2</sub>-1

interface. It is speculated that the reduced trap states at the MAPbI<sub>3</sub>/SnO<sub>2</sub>-2 interface is associated with the decreased surface V<sub>O</sub> of the SnO<sub>2</sub>-2 thin films as verified from XPS results shown above.

The energy band structure of SnO<sub>2</sub> thin films was further studied by UPS and UV-vis absorption results. Fig. 4 exhibits the UPS spectra of SnO<sub>2</sub>-1 and SnO<sub>2</sub>-2 thin films. As reported before, the valence band maximum ( $E_{\text{V}}$ ) can be calculated from the following eqn (1):<sup>37</sup>

$$E_{\text{V}} = -[h\nu - (E_{\text{cutoff}} - E_{\text{V}}^{\text{F}})] \quad (1)$$

where  $h\nu$  is the photon energy of He I lamp (21.22 eV),  $E_{\text{cutoff}}$  is the cutoff energy of the secondary electrons, and  $E_{\text{V}}^{\text{F}}$  is the injection barrier (schematically shown in Fig. 4a). As shown in Fig. 4a and b, the  $E_{\text{cutoff}}$  are 16.31 eV and 16.20 eV, and the  $E_{\text{V}}^{\text{F}}$  are 3.41 eV and 3.17 eV for SnO<sub>2</sub>-1 and SnO<sub>2</sub>-2, respectively. This results in the  $E_{\text{V}}$  as  $-8.32$  eV and  $-8.19$  eV accordingly. In addition, according to the UV-vis absorption spectra shown in Fig. 4c and d, where the optical bandgap of SnO<sub>2</sub>-1 and SnO<sub>2</sub>-2 were obtained as around 4.1 eV, the corresponding conduction band minimums ( $E_{\text{C}}$ ) were calculated as  $-4.19$  eV and  $-4.02$  eV. This demonstrates that the spin-coating procedure up-shifted the conduction band (+0.13 eV) of the SnO<sub>2</sub> thin films. Similar phenomena have been demonstrated in graphdiyne and NH<sub>4</sub>S decorated SnO<sub>2</sub> thin films reported before.<sup>33,38,39</sup> Accordingly, the energy band diagrams of SnO<sub>2</sub>-1 and SnO<sub>2</sub>-2 thin films were obtained, as shown in Fig. 4a and b inset.

After depositing thin circular Au electrodes on MAPbI<sub>3</sub> thin films and Ag paste on ITO, MAPbI<sub>3</sub>/SnO<sub>2</sub> heterojunction PDs were fabricated. Fig. 5a illustrates the current-voltage ( $I$ - $V$ ) curves of MAPbI<sub>3</sub>/SnO<sub>2</sub>-1 and MAPbI<sub>3</sub>/SnO<sub>2</sub>-2 PDs captured in dark and under 540 nm illumination. As illustrated, the MAPbI<sub>3</sub>/SnO<sub>2</sub>-2 PDs shows a decreased dark current and an improved photocurrent compared to MAPbI<sub>3</sub>/SnO<sub>2</sub>-1 PDs. According to previous reports,<sup>17,40</sup> the suppressed V<sub>O</sub> at the surface of SnO<sub>2</sub>-2 thin films is responsible for the reduced dark current and improved photocurrent. Besides, the up-shift of conduction band of SnO<sub>2</sub>-2 thin films further increases the photocurrent.<sup>41</sup> Fig. 5b shows the current-time curves of both



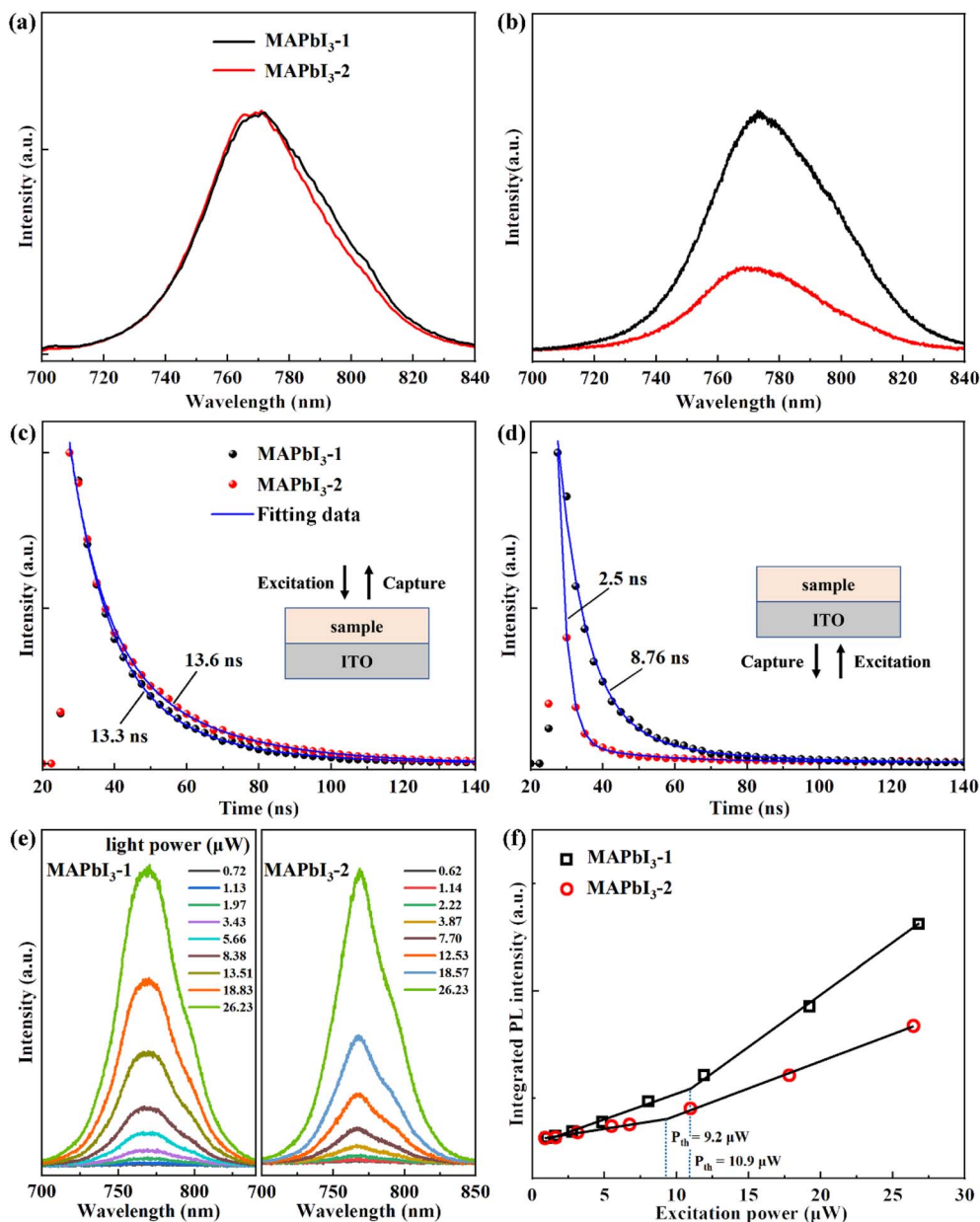


Fig. 3 Optical properties of MAPbI<sub>3</sub> thin films prepared on SnO<sub>2</sub>-1 and SnO<sub>2</sub>-2. (a and b) Steady-state and (c and d) transient PL spectra. (e and f) Excitation power dependent PL spectra. The PL data in a and c were captured from the front surface of MAPbI<sub>3</sub>, as shown in Fig. 1c inset. The PL data in b, d, e and f were captured from the back surface of MAPbI<sub>3</sub>, as shown in Fig. 1d inset.

PDs captured at zero bias with 540 nm illumination switched on and off. As shown, the photocurrent of both PDs increases linearly with an increase in light power, in which the slope of 0.06 A W<sup>-1</sup> and 0.02 A W<sup>-1</sup> for MAPbI<sub>3</sub>/SnO<sub>2</sub>-2 and MAPbI<sub>3</sub>/SnO<sub>2</sub>-1 PDs were derived (Fig. 5b inset). Generally, the increased slope of light power dependent photocurrent in MAPbI<sub>3</sub>/SnO<sub>2</sub>-2 PDs suggests the improved photodetection performance of them.

The photodetection performance of MAPbI<sub>3</sub>/SnO<sub>2</sub> PDs was further evaluated from the responsivity and detectivity parameters. According to previous reports, the responsivity ( $R$ ) and detectivity ( $D^*$ ) were calculated using the following eqn (2) and (3):<sup>17</sup>

$$R = \frac{I_p - I_D}{P \times S} \quad (2)$$

$$D^* = \frac{R}{(2eI_D/S)^{0.5}} \quad (3)$$

where  $I_p$  and  $I_D$  are the photocurrent and dark current,  $S$  is the device area,  $P$  is the light power density (in W cm<sup>-2</sup>), and  $e$  is the electron charge. Fig. 5c and d exhibit the responsivity and detectivity of MAPbI<sub>3</sub>/SnO<sub>2</sub>-1 and MAPbI<sub>3</sub>/SnO<sub>2</sub>-2 PDs at zero bias. As shown, both the PDs shows a good photodetection performance in the UV to vis spectral range (350 nm–750 nm). Notably, the MAPbI<sub>3</sub>/SnO<sub>2</sub>-2 PDs exhibits an improved responsivity and detectivity (0.077 A W<sup>-1</sup> and  $2 \times 10^{11}$  jones), which



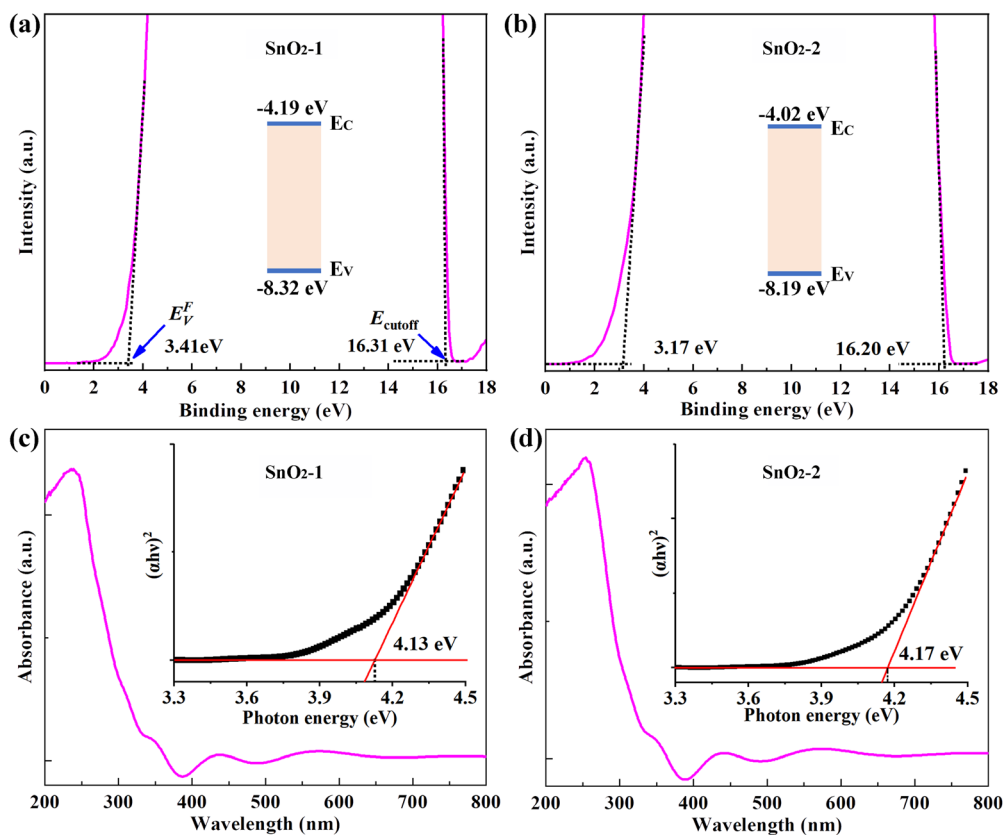


Fig. 4 (a and b) UPS and (c and d) absorption spectra of (a and c) SnO<sub>2</sub>-1 and (b and d) SnO<sub>2</sub>-2 thin films. The inset in (a and b) shows the energy band diagrams and in (c and d) shows the plot of  $(\alpha h\nu)^2$  as a function of photon energy ( $h\nu$ ).

are 2.5 times and 2 times higher than the MAPbI<sub>3</sub>/SnO<sub>2</sub>-1 PDs (0.03 A W<sup>-1</sup> and  $1 \times 10^{11}$  jones). As discussed above, the improved photodetection performance is attributed to the modification of the buried SnO<sub>2</sub> thin films in terms of reduced surface V<sub>O</sub> and up-shift of conduction band. Meanwhile, a weak peak has appeared at around 850 nm in the responsivity and detectivity, which can be attributed to the defect-related absorption in the MAPbI<sub>3</sub> thin films.<sup>42</sup>

Finally, the mechanisms of improved photodetection performance in MAPbI<sub>3</sub>/SnO<sub>2</sub>-2 PDs are discussed. Fig. 6 illustrates the photodetection processes of MAPbI<sub>3</sub>/SnO<sub>2</sub>-1 and MAPbI<sub>3</sub>/SnO<sub>2</sub>-2 PDs. Initially, electron-hole pairs were generated within the MAPbI<sub>3</sub> thin films upon illumination. Then, the electrons were extracted by the SnO<sub>2</sub> thin films owing to the interfacial electrical field and the holes moved toward the Au electrode. As a result, remarkable photocurrent has been observed in the MAPbI<sub>3</sub>/SnO<sub>2</sub>-1 and MAPbI<sub>3</sub>/SnO<sub>2</sub>-2 PDs. However, remarkable non-radiative recombination occurred at the MAPbI<sub>3</sub>/SnO<sub>2</sub>-1 interface (Fig. 6a) due to the high density of V<sub>O</sub> at the surface of SnO<sub>2</sub>-1 thin films. This resulted in an increased dark current and decreased photocurrent. In comparison, the non-radiative recombination was suppressed at the MAPbI<sub>3</sub>/SnO<sub>2</sub>-2 interface (Fig. 6b) owing to the suppressed V<sub>O</sub> at the surface of SnO<sub>2</sub>-2. Accordingly, improved photocurrent and decreased dark current have been observed

simultaneously in the MAPbI<sub>3</sub>/SnO<sub>2</sub>-2 PDs. Besides, the up-shift of conduction band of SnO<sub>2</sub>-2 facilitated the electron transport at the MAPbI<sub>3</sub>/SnO<sub>2</sub> interface (Fig. 6b), which further improved the photocurrent. Collectively, the spin-coating procedure behaved a synergic effect including reduced surface V<sub>O</sub> and increased conduction band of the SnO<sub>2</sub>, which accounted for the improved performance of photodetection of MAPbI<sub>3</sub>/SnO<sub>2</sub>-2 PDs.

In summary, performance-improved solution-processed MAPbI<sub>3</sub>/SnO<sub>2</sub> heterojunction PDs have been fabricated by employing a stacked SnO<sub>2</sub> thin film prepared by sputtering and spin-coating methods in sequence. It has shown that the spin-coating procedure have little impacts on the growth of MAPbI<sub>3</sub> thin films, while it behaved a synergic effect such as reduced the surface V<sub>O</sub> and up-shifted the conduction band of SnO<sub>2</sub> thin films. The former suppressed the interfacial non-radiative recombination, while the latter aligned the energy-level at the MAPbI<sub>3</sub>/SnO<sub>2</sub> interface and further improved the carrier transport. On these basis, the MAPbI<sub>3</sub>/SnO<sub>2</sub> heterojunction PDs with a stacked SnO<sub>2</sub> thin film exhibited an improved photodetection performance in terms of reduced dark current and increased photocurrent, where an improved responsivity and detectivity of 0.077 A W<sup>-1</sup> and  $2.0 \times 10^{11}$  jones, respectively, at the zero bias have been achieved. The results provided in this work provide a simple strategy for suppressing



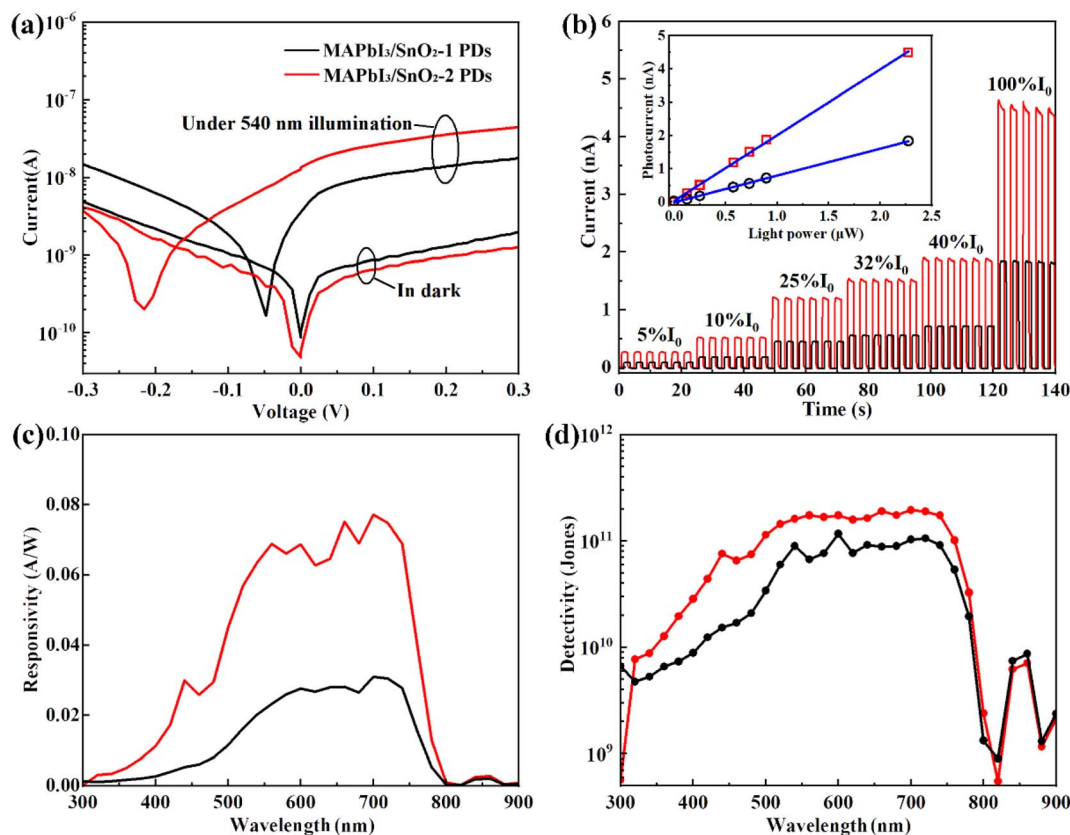


Fig. 5 Photodetection performance of MAPbI<sub>3</sub>/SnO<sub>2</sub> PDs. (a) *I*–*V*. (b) Current–time curves. (c) Responsivity. (d) Detectivity. The bias in (b–d) is 0 V. The light was illuminated at the back surface of MAPbI<sub>3</sub> thin films.

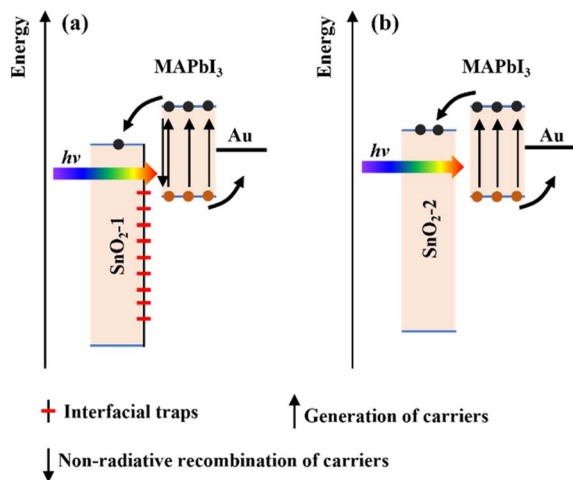


Fig. 6 Photodetection processes of (a) MAPbI<sub>3</sub>/SnO<sub>2</sub>-1 and (b) MAPbI<sub>3</sub>/SnO<sub>2</sub>-2 PDs.

the interfacial defects and aligning the energy-level simultaneously for improving the performance of perovskite hetero-junction PDs in the future.

## Data availability

The data that support the findings of this study are available from the corresponding authors upon reasonable request.

## Conflicts of interest

The authors have declared that no conflict of interest exists.

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