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## Surface passivation of organometal halide perovskites by atomic layer deposition: an investigation of the mechanism of efficient inverted planar solar cells<sup>†</sup>

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Interface passivation plays a pivotal role in achieving high-efficiency organic metal halide perovskite solar cells (PSCs). It has been recently revealed that atomic layer deposition (ALD) of wide-band gap oxides shows great potential to effectively passivate defects at the interface, and ALD is also of great technological promise for industrial upscaling. However, the conflicting observations of ALD passivation are often reported in the literature, even with very similar ALD conditions. To unveil the involved crucial mechanism, this work carefully investigates the evolution of a representative  $\text{MAPbI}_3$  perovskite surface during the ALD of  $\text{Al}_2\text{O}_3$ , by employing the technique of *in situ* X-ray photoelectron spectroscopy. The ALD at 125 °C was found to cause significant degradation of the perovskite; lowering the deposition temperature can largely minimize the degradation, and 75 °C was found to be the best ALD temperature. Following this conclusion, inverted planar perovskite solar cells were prepared in ambient conditions with ALD  $\text{Al}_2\text{O}_3$  interlayers. Indeed, cells with the interlayer deposited at 75 °C exhibited a significantly enhanced power conversion efficiency from 18.8% (champion 19.2%) to 20.0% (champion 20.4%). Photoluminescence measurements further evidence that the ALD layer can effectively passivate defect states at the perovskite surface. Considering the great representativeness and broad applicability of  $\text{MAPbI}_3$  and ALD  $\text{Al}_2\text{O}_3$ , the mechanism and strategy reported herein should be of significant value for the perovskite interface engineering in general.

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## Introduction

Organic-inorganic halide perovskite solar cells (PSCs) are highly promising emerging photovoltaic cells, and they have shown rapid improvement in certified power conversion efficiency (PCE) from 3.8% to 25.5% over the past decade.<sup>1,2</sup> While the halide perovskites appear to be resilient to bulk point defects (such as vacancies, interstitials, and antisites), their optoelectronic properties are fairly susceptible to the quality of perovskite (PVK) surfaces and interfaces.<sup>3</sup> Given current synthesis techniques,

perovskite surfaces are rarely stoichiometric and usually defective.<sup>3</sup> Defects at perovskite surfaces could induce significant nonradiative charge recombination and deteriorate the solar cell performance.<sup>3–7</sup> Particularly, in state-of-the-art solar cells, the nonradiative recombination at interfaces between the perovskite and charge transport layers is predominant compared to those in the bulk or at grain boundaries.<sup>4</sup> Therefore, the passivation of interface defects is pivotal to achieve high-efficiency PSCs. To this end, numerous passivation approaches have been reported, for instance, by adding an interlayer of small molecules, polymers, and inorganic compounds or incorporating wide-band gap 2D perovskites.<sup>8–11</sup> Despite their promising improvements on solar cell efficiency, concerns still remain about the upscalability of these approaches.

Recently, atomic layer deposition (ALD) of wide-band gap oxides (e.g.  $\text{Al}_2\text{O}_3$ ) has emerged as a promising strategy to effectively passivate perovskite surfaces.<sup>12</sup> ALD is a scalable vapor-phase thin-film deposition technique, which relies on sequential alternate self-limiting surface reactions, and it is renowned for producing highly uniform, conformal thin films on non-flat surfaces with atomic-precise film thickness control and

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remarkable process reproducibility.<sup>13,14</sup> Currently, ALD has been widely used in the Si photovoltaic industry for the production of passivated emitter and rear cells (PERC).<sup>15</sup> Analogously, the passivation of perovskites by ALD is of great technological promise for future industrial upscaling. However, despite a number of positive reports showing promising improvements in the efficiency and stability of PSCs, direct growth of ALD on the top of perovskites has often encountered significant incompatibility issues.<sup>12,16,17</sup> During an ALD process, perovskites are subjected to an often elevated temperature and the exposures of ALD precursors could potentially change the structure, composition, and therefore the optoelectronic properties of perovskites. Notably, these effects are largely dependent on the ALD surface reaction chemistries and the properties of perovskites (such as composition and surface condition).<sup>12,16,17</sup> Hence, understanding the fundamental mechanisms of the interactions between ALD precursors and perovskites becomes the key to reliably employing this ALD technology for PSCs. Toward this goal, a growing number of reports have recently appeared to investigate the related ALD mechanisms, but often conflicting observations were published in the literature.<sup>12</sup> For instance, taking the most studied system of ALD  $\text{Al}_2\text{O}_3$  on methylammonium lead iodide ( $\text{MAPbI}_3$ ) as an example, under quite similar deposition conditions, both increase and decrease in the solar cell efficiency have been reported.<sup>18–26</sup> Notably, a recent study<sup>27</sup> using *in situ* quartz crystal microbalance (QCM) and quadrupole mass spectrometry (QMS) showed that the differences in the ALD temperature and precursor partial pressure could result in drastically different outcome interfaces, which were suggested to account for the discrepancies observed across different laboratories. Their conclusion, together with the well-known sensitivity of the perovskite surfaces and subtleties in preparation, highlights two very important issues for this type of studies. First, the mechanism study should be accompanied with the high-reliability and high-performance PSCs; otherwise, the conclusions do not reflect the *de facto* limiting factors. Second, the mechanism study should be conducted *in situ* or *in vacuo*, in order to avoid possible ambient interference when exposed to air.

To find out the critical mechanism underlying ALD passivation, we herein focused on the PSCs with high benchmarking PCE (18.8% in average) and employed the technique of *in situ* X-ray photoelectron spectroscopy (XPS) to investigate the evolution of the perovskite surface during the ALD process. The representative perovskite of  $\text{MAPbI}_3$  and the ALD material of  $\text{Al}_2\text{O}_3$  were chosen for this study. Following the careful study, we found that quite a few experimental observations were rather different from previous reports, which highlight the importance of carrying out *in situ* investigation on high-efficiency PSCs. Therefore, the results and mechanism reported herein should be of significant value for the interface engineering of high-performance PSCs in general.

## Experimental

### Materials

Fluorine-doped tin oxide (FTO, 1.5 cm × 1.5 cm) was purchased from OPV TECH LTD. Methylammonium iodide (MAI) was

purchased from Dyesol Ltd. Lead iodide ( $\text{PbI}_2$ , extra dry, 99.999%) and nickel(II) acetylacetone (95%) were purchased from Alfa Aesar Ltd. *N,N*-Dimethylformamide (DMF, anhydrous, 99.8%), dimethyl sulfoxide (DMSO, anhydrous, 99.8%) and chlorobenzene (CB, extra dry, 99.5%) were purchased from Acros Ltd. 2,2,2-Trifluoroethanol (TFE, 99.5%) was purchased from Energy Chemical Ltd. [6,6]-Phenyl-C61-butyric acid methyl ester ( $\text{PC}_{61}\text{BM}$ ) was purchased from NANO-C Tech Ltd. Anisole (anhydrous, 99.7%) was purchased from Aladdin Ltd. Diethanol amine (DEA, 99%) was purchased from Sigma-Aldrich Ltd.

### Device fabrication

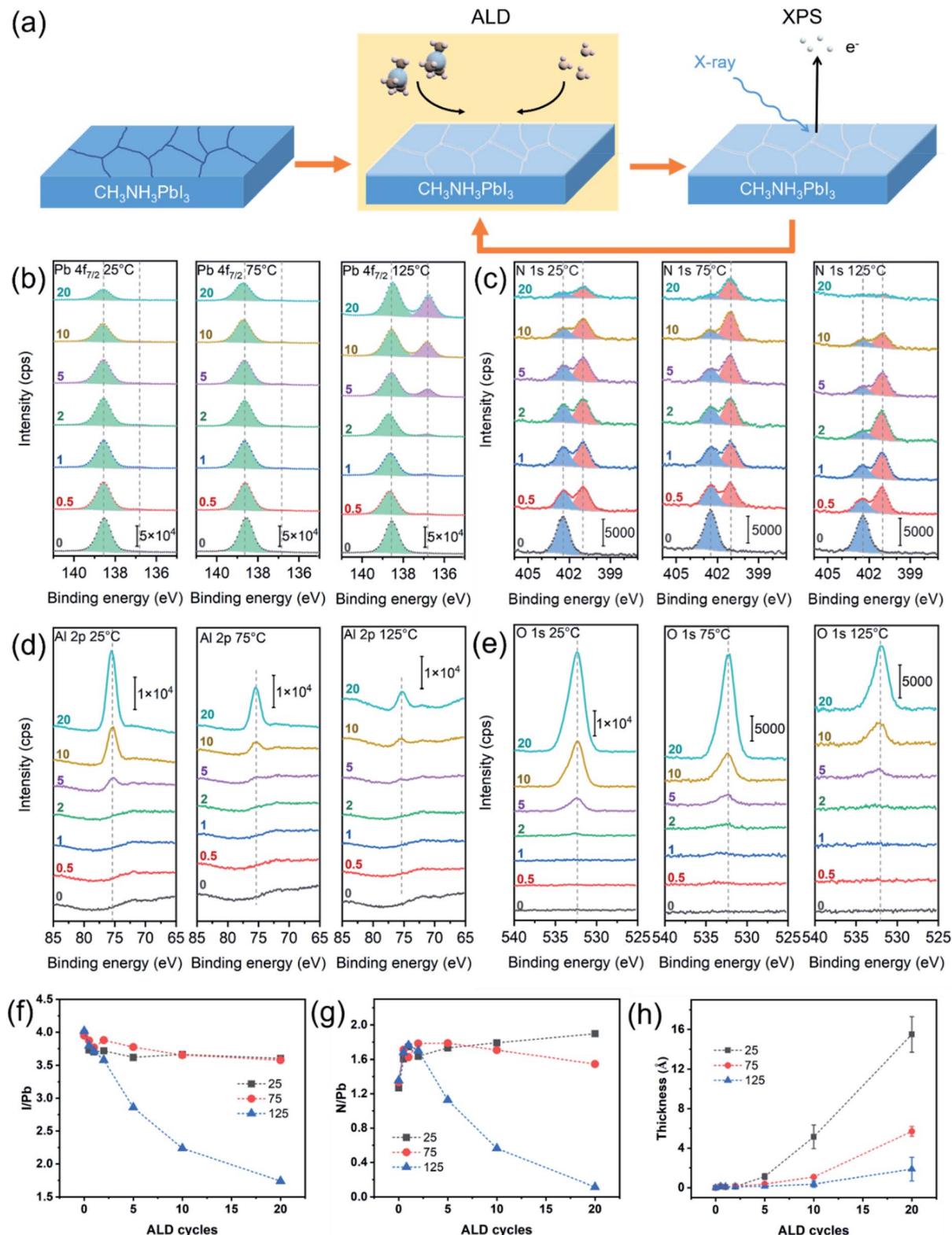
**PSC fabrication.** FTO substrates were ultrasonically cleaned with deionized water, ethanol, isopropanol, acetone and ethanol sequentially, and dried in a circulation oven at 80 °C. A NiO film was fabricated according to our previous publication.<sup>8</sup> The sol-gel solution was prepared by dissolving nickel(II) acetylacetone and DEA in anhydrous ethanol. A NiO layer was fabricated using a home-made ultrasonic spray.

The perovskite precursor solution was prepared according to our previous publications.<sup>28,29</sup> First, 461 mg of  $\text{PbI}_2$  and 159 mg of MAI were dissolved in a mixture of 434  $\mu\text{L}$  DMSO and 124  $\mu\text{L}$  DMF. After stirring on a 60 °C hotplate overnight, the transparent yellowish solution was filtered using a filter head of 0.22  $\mu\text{m}$ . A perovskite film was prepared using the as-prepared solution in a fume hood under the humidity between 20 and 60% RH. Then, 35  $\mu\text{L}$  of the as-prepared precursor solution was drop-casted onto the substrate at room temperature. The spin coating process consisted of four steps, which were 1000 rpm for 45 s, 6000 rpm for 40 s, 0 rpm for 10 s, and 6000 rpm for 10 s sequentially. During the last 5 s of the second step and the whole process of the fourth step, 80  $\mu\text{L}$  anisole was dropped onto the substrate, respectively. After the first dripping, the film was still transparent and it could maintain the state for a long time in ambient air. After the second dripping, the film turned brown in color, and the as-prepared film was converted into a compact perovskite film at room temperature or 98 °C for 7 min in an ambient atmosphere. On top of the perovskite film, atomic layer deposition (ALD) of  $\text{Al}_2\text{O}_3$  was carried out under different conditions.

The  $\text{PC}_{61}\text{BM}$  solution (20 mg  $\text{mL}^{-1}$  in CB) was fabricated and stirred at 60 °C for 2 h. First, 35  $\mu\text{L}$  of  $\text{PC}_{61}\text{BM}$  solution was drop-casted onto the substrate and spun at 3000 rpm for 30 s. The interface material PPDIN6 was synthesized according to our previous publication.<sup>11</sup> The PPDIN6 solution (0.5 mg  $\text{mL}^{-1}$  in TFE) was fabricated and stirred overnight at room temperature. Then, 40  $\mu\text{L}$  of PPDIN6 solution was drop-casted onto the substrate and spun at 3000 rpm for 30 s. Finally, 150 nm of Ag electrode was evaporated on top.

**ALD of  $\text{Al}_2\text{O}_3$ .** The deposition process was conducted using a home-built tubular ALD reactor, with trimethylaluminum (TMA) and water as the aluminum and oxygen precursors, respectively. Both precursors were kept at room temperature in separate containers. To control the precursor dosage, the precursor vapor was delivered first into a 5 mL gas trap and then into a deposition chamber. Each ALD cycle consisted of one pulse of TMA, 40 s of  $\text{N}_2$  purge, one pulse of water, and 100 s of





**Fig. 1** (a) Schematic of the *in situ* experimental procedure. High-resolution XPS core-level spectra of (b)  $\text{Pb} 4f_{7/2}$ , (c)  $\text{N} 1s$ , (d)  $\text{Al} 2p$ , and (e)  $\text{O} 1s$  taken from 0 cycles (i.e., initial surface) through 20 cycles of ALD of  $\text{Al}_2\text{O}_3$  on  $\text{MAPbI}_3$ . The deposition temperatures were 25 °C, 75 °C and 125 °C. Atomic ratios of (f)  $\text{I}/\text{Pb}$  and (g)  $\text{N}/\text{Pb}$  and (h) the equivalent thickness of the deposited  $\text{Al}_2\text{O}_3$  with respect to the number of the ALD cycles.

$N_2$  purge in sequence. During  $N_2$  purging, the pressure of the deposition chamber was 0.5 Torr.

## Characterization

**In situ X-ray photoelectron spectroscopy.** The *in situ* X-ray photoelectron spectroscopy measurements were performed using a modified Escalab 250Xi (Thermo Scientific) XPS system equipped with an ALD chamber. Details about this system can be found in our previous publications.<sup>30–32</sup> The system contained an ALD chamber for deposition, an intermediate chamber for sample transfer without vacuum break, and an analysis chamber for XPS measurements. The analysis chamber was equipped with a monochromatic Al  $K\alpha$  X-ray source (1486 eV). The pass energy was 20 eV for all high-resolution spectra. The binding energy was referenced to Au 4f<sub>7/2</sub> (84.0 eV). For each measurement, the spectrum was taken on a new spot to avoid possible radiation damage from X-ray. Standard sensitive factors provided in Thermo Advantage (v5.945) were used to calculate the atomic ratios, and the peak areal intensities were used for the calculation. The equivalent thickness of  $Al_2O_3$  was calculated based on the XPS signal attenuation,<sup>33</sup> by  $d = \lambda \sin \theta \ln \left( \frac{I_{Al\ 2p}}{\beta I_{I\ 3d}} + 1 \right)$ , where  $\lambda$  (=2.3 nm (ref. 34)) is the effective attenuation length of the photoelectrons in the  $Al_2O_3$  layer,  $\theta$  (=90°) is the photoelectron take off angle,  $I_{Al\ 2p}$  and  $I_{I\ 3d}$  are the measured intensities of the Al 2p and I 3d signals from the substrate, respectively, and  $\beta$  ( $= I_{Al\ 2p}^\infty / I_{I\ 3d}^\infty$ ) is the ratio of the Al 2p and I 3d intensities for thick  $Al_2O_3$  and perovskite substrates.

**Photoluminescence spectra.** Steady-state PL spectra were recorded using a Horiba Fluorolog-3 spectrophotometer with a 450 W xenon lamp as the continuous light source. Time-resolved PL spectra were recorded using a PicoFluo TRPL spectrophotometer (Pico-1000). The wavelengths of the excitation light and detection light were 405 nm and 767 nm, respectively. The obtained TRPL curves were fitted by a bi-exponential decay function,  $y = A_1 e^{-\frac{t}{\tau_1}} + A_2 e^{-\frac{t}{\tau_2}} + y_0$ ,  $\tau_{ave} = \frac{A_1 \tau_1^2 + A_2 \tau_2^2}{A_1 \tau_1 + A_2 \tau_2}$ , where  $A_1$  and  $A_2$  are the relative amplitudes and  $\tau_1$ ,  $\tau_2$ , and  $\tau_{ave}$  are the fast, slow, and average decay lifetimes, respectively.

**Other characterizations.** The morphologies of the samples were characterized using a field-emission scanning electron microscope (SEM, JEOL 7100F). The X-ray diffraction (XRD) patterns were recorded using a diffractometer (D8 Discover, Bruker AXS, Germany) with a Cu  $K\alpha$  radiation. The current density and voltage ( $J-V$ ) of the devices were measured using a source meter (Keithley, 2400) under 1 Sun illumination (AM 1.5G, 100 mW cm<sup>-2</sup>) with a solar simulator (Newport, Oriel Sol3A). The measurement area was confined by a mask of 0.08 cm<sup>2</sup>. The incident photon-to-current conversion efficiency (IPCE) was measured using a photoelectric detection test system (Zolix Instrument).

## Results and discussion

*In situ* XPS was employed to monitor the evolution of the MAPbI<sub>3</sub> perovskite surface upon the ALD of  $Al_2O_3$  from

trimethylaluminum (TMA) and  $H_2O$  (Fig. 1a). Three representative ALD temperatures of 25 °C, 75 °C and 125 °C were selected to investigate the temperature effect. As a few recent reports<sup>27,35</sup> suggested that the MAPbI<sub>3</sub> perovskite could be etched by excessively exposing to the TMA and  $H_2O$  vapors, we chose to use relatively small exposures of 0.025 and 0.051 Torr s for the TMA and  $H_2O$  vapors, respectively, in each ALD cycle. These exposures were sufficient to saturate the normal ALD growth of  $Al_2O_3$  on the flat  $SiO_2$  surface.<sup>36</sup>

Fig. 1b–e shows the acquired *in situ* XPS spectra for the core-level emissions of Pb 4f<sub>7/2</sub>, N 1s, Al 2p, and O 1s. All the XPS spectra were taken for 0 cycle (*i.e.*, initial surface) through 20 cycles of  $Al_2O_3$  deposited on the perovskite surface. Half-integer and full-integer numbers were used to denote the spectra taken after the TMA and  $H_2O$  half-cycles, respectively. Fig. 1b compares the evolutions of the Pb 4f<sub>7/2</sub> spectra upon the ALD at 25, 75 and 125 °C. All the spectra for the pristine perovskite surfaces (0 cycle) show a similar symmetric peak at a binding energy (BE) of 138.6 eV, which corresponds to  $Pb^{2+}-I$  in perovskites.<sup>37</sup> However, as the ALD proceeded, the Pb 4f<sub>7/2</sub> spectra behaved differently for the depositions at different temperatures. At a low deposition temperature of 25 °C or 75 °C, the intensity of the  $Pb^{2+}-I$  peak continually diminished as the deposition proceeded, probably owing to the increasing attenuation of the photoelectrons by the deposited  $Al_2O_3$  layer on top. Comparatively, the signal diminished faster for 25 °C than 75 °C, suggesting a faster per-cycle growth of  $Al_2O_3$  at 25 °C. As for the ALD at 125 °C, the Pb 4f<sub>7/2</sub> spectrum continually developed an additional peak at a lower BE of 136.8 eV, and this BE corresponds to metallic lead ( $Pb^0$ ),<sup>37</sup> which suggests that the  $Pb^{2+}$  in perovskite was reduced upon the ALD at 125 °C. Considering that the conduction band minimum of MAPbI<sub>3</sub> has mostly nonbonding Pb *p* character,<sup>38</sup> the reduction of  $Pb^{2+}$  to  $Pb^0$  can significantly alter the optoelectronic properties of MAPbI<sub>3</sub>. Perhaps because it was only present on the surface and of a small amount compared to the bulk, the metallic lead was not observed by XRD or IR in previous ALD studies,<sup>18,35,39</sup> and thus its adverse effect on the solar cell performance was likely overlooked. In addition, it is worth noting that no  $Pb-OH$  species (at 138.2 eV (ref. 20)) was detected on the pristine perovskite surfaces or after the ALD, and this observation is in stark contrast to a previous study on relatively low-PCE (13.6%) MAPbI<sub>3</sub> surface,<sup>20</sup> which therefore signifies a different passivation mechanism for high-PCE MAPbI<sub>3</sub> surfaces.

Fig. 1c compares the evolutions of the N 1s spectra upon the ALD at 25, 75 and 125 °C. All the spectra of the pristine perovskite surfaces show a single peak at 402.4 eV for N in  $CH_3NH_3^+$ ,<sup>37</sup> but after the TMA dose, a pronounced new peak emerged at a lower BE of 401.0 eV. The appearance of this peak is suggestive of the deprotonation of  $CH_3NH_3^+$  by TMA to afford  $CH_3NH_2$ ,<sup>35,40</sup> which has a lower N 1s BE.<sup>41,42</sup> Although  $CH_3NH_2$  itself is a gas at room temperature (b.p. –6.8 °C), for the ALD at  $\leq 75$  °C, most of the afforded  $CH_3NH_2$  seemed to remain in the perovskite or be trapped at the perovskite/ $Al_2O_3$  interface. In contrast, the ALD at 125 °C resulted in a significant reduction of the overall N 1s signal (*e.g.* 20 cycles), which indicates a significant loss of  $CH_3NH_3^+$  at 125 °C.



Fig. 1d and e displays the acquired Al 2p and O 1s spectra upon the ALD at 25, 75 and 125 °C. All the Al 2p spectra show a similar trend, where no Al 2p signal was observed for the first 2 cycles, indicating that the growth of ALD  $\text{Al}_2\text{O}_3$  on the perovskite was delayed. After 5 ALD cycles, the spectra showed observable Al 2p peaks, and the peak intensities later increased as the ALD proceeded. The peak BE was found to be at 75.5 eV, which agrees with  $\text{Al}_2\text{O}_3$  (Al-OH).<sup>21,43</sup> The O 1s spectra showed correlated results that the O signals were barely observable after 2 cycles and became prominent (at 532.7 eV, Al-OH<sup>21,43</sup>) after 5 cycles, and these results corroborate that the ALD  $\text{Al}_2\text{O}_3$  growth was delayed on the perovskite. It is worth noting that the O 1s spectra for the initial  $\text{MAPbI}_3$  surfaces do not contain any detectable O signals, which is in stark contrast to previous photoelectron spectroscopy studies.<sup>20,21</sup> This discrepancy signifies different ALD growth behaviors, which could therefore lead to different passivation mechanisms for the perovskite solar cells. Nevertheless, the intensities of the Al 2p and O 1s peaks are correlated with the amount of the  $\text{Al}_2\text{O}_3$  grown on the perovskite surface, and apparently the growth was slower at a higher temperature. In addition, I 3d<sub>5/2</sub> spectra were also taken. As shown in Fig. S1,† aside from a continual signal intensity decrease with the ALD cycle number, all the I 3d<sub>5/2</sub> spectra show only a single peak at 619.4 eV for  $\text{Pb}^{2+}$ -I,<sup>37</sup> and thus no iodate-like species (at 623.8 eV (ref. 44)) was formed.

Based on the above-mentioned *in situ* XPS results, the change in the perovskite surface composition was traced. Fig. 1f and g plot the atomic ratios of I/Pb and N/Pb extracted from the XPS spectra. Because the mean free path of the photoelectrons is about 2–3 nm,<sup>34</sup> these ratios reflect only the compositional information near the perovskite surface. The I/Pb and N/Pb ratios for the pristine perovskite surfaces (*e.g.*, 0 ALD cycle) were found to be 4.0 and 1.3, respectively. These numbers are greater than those in the stoichiometric  $\text{MAPbI}_3$ , which suggests that the initial perovskite surface contained excessive iodine and MAI.<sup>45</sup> Although the I-rich MAI surface termination is more stable than the  $\text{PbI}_2$  termination for the  $\text{MAPbI}_3$  surface,<sup>46–48</sup> excessive iodine can result in deep electronic traps, thereby deteriorating the solar cell performance.<sup>49</sup> Apparently, as shown in Fig. 1f, the surface reactions of the I-rich  $\text{MAPbI}_3$  with the ALD precursors of TMA and  $\text{H}_2\text{O}$  can reduce the surface I/Pb ratio, by forming volatile  $\text{CH}_3\text{I}$ , as previously identified *via* *in situ* QMS.<sup>27</sup> As shown in Fig. 1f, the 10- and 20-cycle ALD at 25 °C or 75 °C reduced the I/Pb ratio to 3.7 and 3.6, respectively. As for the ALD at 125 °C, the I/Pb ratio drastically decreased to 2.2 and 1.7 after 10 and 20 ALD cycles, respectively, which is consistent with the formation of metallic Pb as previously suggested from the BE change. Concurrently, a significant drop of the N/Pb ratio to 0.1 was found for the 20-cycle ALD at 125 °C (Fig. 1g). Such a low N/Pb ratio indicates a significant loss of the  $\text{CH}_3\text{NH}_3^+$  ions, which eventually collapsed the perovskite structure to form  $\text{PbI}_2$  (as will be shown later). However, we also noticed a pronounced increase in the N/Pb ratio (from 1.3 to 1.7) upon the first TMA dose (0.5 cycle) for all the temperatures. Given the surface-sensitivity of the XPS technique, the increase in the N/Pb ratio suggests that a certain amount of  $\text{CH}_3\text{NH}_2$  accumulated near the perovskite surface.  $\text{CH}_3\text{NH}_2$  could be

afforded at the perovskite surface and grain boundaries; for the latter, the afforded  $\text{CH}_3\text{NH}_2$  diffused outward through the grain boundaries to the top of the perovskite layer. This situation is analogous to a technique called methylamine-induced defect-healing (MIDH), which aims to reduce the perovskite defects by recrystallizing it *via* an intermediate  $\text{MAPbI}_3 \cdot x\text{CH}_3\text{NH}_2$  liquid.<sup>50,51</sup> Certainly, the *in situ* formed  $\text{CH}_3\text{NH}_2$  in our case was only of a small amount and thus not enough to form a full liquid intermediate, but the defects at the surface can be healed by the same scheme. In addition,  $\text{CH}_3\text{NH}_2$  can also passivate the surface I vacancies by coordinating to  $\text{Pb}^{2+}$ , forming a Pb-N bond which is much stronger than Pb-OH.<sup>52</sup>

The thickness of the ALD  $\text{Al}_2\text{O}_3$  layer can be estimated from the degree of attenuation of the photoelectron signal from the perovskite underneath (see Experimental). Fig. 1h plots the extracted equivalent  $\text{Al}_2\text{O}_3$  thickness with respect to the ALD cycle number. For all the deposition temperatures, the initial  $\text{Al}_2\text{O}_3$  growth was delayed for at least 2 cycles, and after 5 cycles, the thickness *versus* cycle number exhibited a superlinear relation, which suggests that the per-cycle growth rate accelerated as the deposition proceeded. This behavior is suggestive of an inhibited initial growth mechanism.<sup>53</sup> Inhibited initial growth is usually because of the lack of ALD-reactive moieties on the substrate surface, and it would normally result in the growth of islands on the substrate.<sup>53</sup> Given the small equivalent thickness of ALD- $\text{Al}_2\text{O}_3$  after 10 cycle ALD at 25–125 °C, the afforded  $\text{Al}_2\text{O}_3$  islands should be no greater than several nanometers in size and therefore elude direct observation by SEM or AFM. Although detailed chemistry of the ALD  $\text{Al}_2\text{O}_3$  growth on  $\text{MAPbI}_3$  is still unknown, the ALD  $\text{Al}_2\text{O}_3$  growth normally starts from TMA reacting with surface hydroxyls to afford Al-O bonds. Presumably, the same reaction scheme is also plausible if the surface defect sites adsorb  $\text{H}_2\text{O}$ . Recently, Choudhury *et al.*<sup>35</sup> and Koushik *et al.*<sup>40</sup> proposed another reaction scheme that TMA reacts with  $\text{MAPbI}_3$  to afford a  $\text{PbI}_3\text{-Al}(\text{CH}_3)_2$  complex; however, our XPS results did not show the presence of Al after the first TMA half-cycle. Therefore, the growth from the surface defect sites is more likely to be the case. Given that the surface defects are often detrimental to the perovskite optoelectronic properties,<sup>3,4</sup> fine-tuning the ALD conditions to well passivate these defects is of crucial importance to achieve high-performance solar cells.

X-ray diffraction (XRD) was employed to investigate the bulk part of the perovskite upon ALD. Fig. 2a compares the obtained XRD patterns for the pristine perovskite and the perovskites with 10-cycle ALD  $\text{Al}_2\text{O}_3$  grown at different temperatures. The pattern of the pristine perovskite contains the characteristic  $\text{MAPbI}_3$  diffraction peaks of (110) and (200) at 14.2° and 20.1°, respectively,<sup>18,54</sup> and the pattern remained almost the same after the 10-cycle ALD at 25 °C or 75 °C. However, for the ALD at 125 °C, a prominent  $\text{PbI}_2$  (001) peak appeared at 12.8°, while the perovskite (110) peak became weaker. These results indicate that a considerable portion of the perovskite was converted to  $\text{PbI}_2$  upon the ALD at 125 °C. Notably, thermal annealing at 125 °C itself was not enough to drive this conversion, as the vacuum annealed sample shows an almost unchanged XRD pattern. Therefore, the surface reactions with the ALD



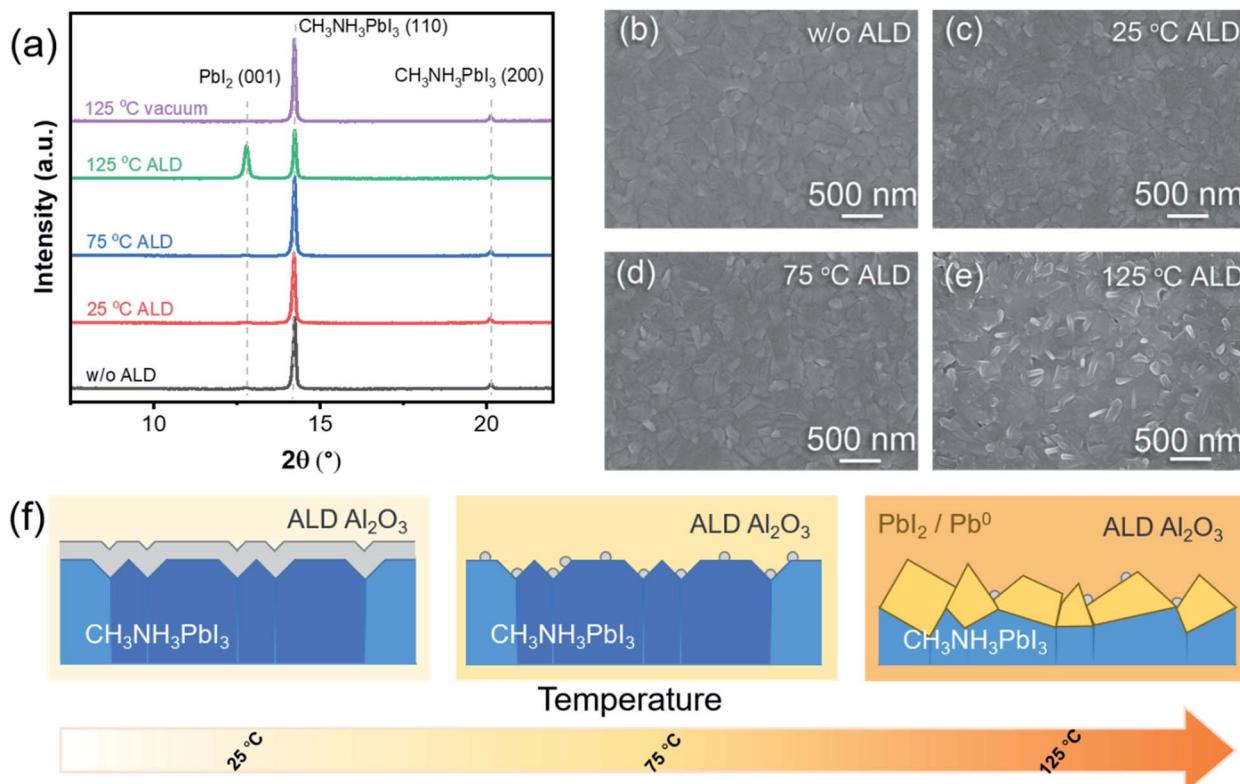


Fig. 2 (a) XRD patterns and (b–e) SEM images taken on a pristine  $\text{MAPbI}_3$  film and the films with 10-cycle ALD of  $\text{Al}_2\text{O}_3$  on top. The deposition temperatures were 25 °C, 75 °C and 125 °C. For comparison, the XRD pattern for a  $\text{MAPbI}_3$  film annealed under vacuum at 125 °C is also shown. (f) Schematic of the evolution of  $\text{MAPbI}_3$  upon the ALD of  $\text{Al}_2\text{O}_3$  at different temperatures.

precursors must be the promoter to the degradation of the perovskite structure. Scanning electron microscopic (SEM) measurements were performed to examine the morphology of the above perovskite samples. The pristine perovskite layer was found to be fairly smooth and uniform (Fig. 2b), and the surface morphology remained almost unchanged after the ALD at 25 °C or 75 °C (Fig. 2c and d) or by vacuum annealing at 125 °C (Fig. S3†). However, the ALD at 125 °C considerably roughened the surface and created pinholes through the perovskite layer (Fig. 2e). On the basis of the above-mentioned analysis, Fig. 2f schematically summarizes the evolutions of  $\text{MAPbI}_3$  upon the ALD of  $\text{Al}_2\text{O}_3$  at different temperatures. While the ALD at 125 °C considerably roughened the perovskite surface and formed metallic lead and  $\text{PbI}_2$ , the degradation was largely minimized as the ALD temperature was lowered to 25 or 75 °C. Compared to 25 °C, the initial ALD growth at 75 °C was apparently slower, and therefore, the growth could occur more selectively on the surface defects.<sup>55</sup> As such, wide-band gap  $\text{Al}_2\text{O}_3$  could be more selective to only passivate the surface defects while not to impede the charge transport across the interface. Therefore, 75 °C was determined to be the best ALD temperature for perovskite passivation. In order to compare the hydrophobicity of the  $\text{MAPbI}_3$  surface, we carried out the contact angle measurement. The contact angle increased from 52.79° to 75.81° after 10 cycles of ALD- $\text{Al}_2\text{O}_3$  at 75 °C (Fig. S4†), which

indicates that the  $\text{MAPbI}_3$  surface became more hydrophobic after ALD of  $\text{Al}_2\text{O}_3$ .

To further demonstrate the ALD passivation effect, we fabricated a series of inverted planar  $\text{MAPbI}_3$  solar cells with ALD  $\text{Al}_2\text{O}_3$  and measured their photovoltaic performance. The cells had a configuration of FTO/ $\text{NiO}_x$ / $\text{MAPbI}_3$ /ALD- $\text{Al}_2\text{O}_3$ /PCBM/PPDIN6/Ag, which is schematically shown in Fig. 3a. The influence of the ALD temperature was first investigated by 10-cycle ALD of  $\text{Al}_2\text{O}_3$  directly on  $\text{MAPbI}_3$  (Fig. 3b). The benchmarking solar cells (*i.e.* without ALD) exhibited an average power conversion efficiency (PCE) of 18.8% (champion of 19.2%). The ALD at the temperatures of 25 °C and 75 °C were both found to increase the PCE. As shown in Fig. 3b and S2,† the open-circuit voltage ( $V_{oc}$ ) and fill factor were both increased by ALD, probably because of the surface defect passivation by the ALD of  $\text{Al}_2\text{O}_3$ . However, the variations in the PSC parameters are considerably larger for 25 °C than for 75 °C, which suggests that the ALD passivation at 25 °C is less stable. However, the ALD at 125 °C was found to drastically deteriorate the efficiency to 14.2%, probably by forming the more insulating  $\text{PbI}_2$  at the surface, which hindered the interface charge transfer and therefore reduced the short-circuit current ( $J_{sc}$ ).

Given the optimal ALD temperature of 75 °C, the effect of the total ALD cycle number was also investigated. As shown in Fig. S5,† the cells with 10-cycle ALD of  $\text{Al}_2\text{O}_3$  gave out the highest PCE, as compared to the 5- and 20-cycle samples. Notably, the



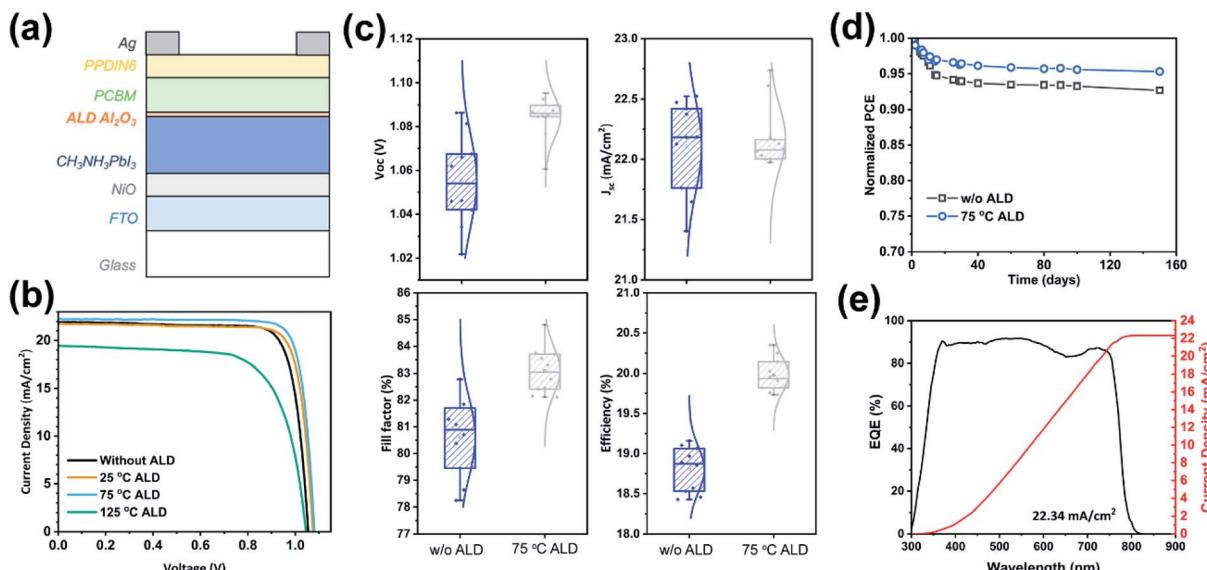


Fig. 3 (a) Schematic architecture of the ambient-prepared PSCs with an ALD Al<sub>2</sub>O<sub>3</sub> interlayer: FTO/NiO<sub>x</sub>/MAPbI<sub>3</sub>/ALD-Al<sub>2</sub>O<sub>3</sub>/PCBM/PPDIN6/Ag. (b) Curves of current density versus voltage ( $J-V$ ) for the PSCs with and without 10 cycles of ALD of Al<sub>2</sub>O<sub>3</sub> deposited at 25 °C, 75 °C and 125 °C, and the corresponding statistical distributions of performance parameters are shown in Fig. S2.† (c) Statistical distributions of  $V_{oc}$ ,  $J_{sc}$ , fill factor, and PCE for the PSCs with and without ALD Al<sub>2</sub>O<sub>3</sub> deposited at 75 °C. (d) Long-term stability of PSCs in a N<sub>2</sub> atmosphere. (e) Incident photon-to-electron conversion efficiency (IPCE) spectrum and the corresponding integrated current density for the PSCs with 10 cycles of ALD of Al<sub>2</sub>O<sub>3</sub> at 75 °C ((b): 75 °C ALD).

equivalent thickness of 10-cycle ALD of Al<sub>2</sub>O<sub>3</sub> was only about 1 Å (Fig. 1h). With this extremely small number, it is highly probable that ALD Al<sub>2</sub>O<sub>3</sub> grew only selectively on the defect sites of the perovskite surface. This selective growth is desirable for effectively passivating the defects while not impeding the charge transport at the interface. As a result, 10-cycle ALD of Al<sub>2</sub>O<sub>3</sub> 75 °C gave out a much-improved average PCE of 20.0% and a champion PCE of 20.4% (Fig. 3c). Moreover, the hysteresis was small for both of the PSCs with and without 10 cycles of ALD of Al<sub>2</sub>O<sub>3</sub> at 75 °C (Fig. S6†).

The stability of the fabricated solar cells was also evaluated in a N<sub>2</sub>-filled glove box. As shown in Fig. 3d, after the initial “burn-in” process,<sup>29,56</sup> the normalized PCE of the solar cells remained fairly stable over the later test period up to 150 days. With the addition of ALD Al<sub>2</sub>O<sub>3</sub>, the eventual retention of the PCE increased from 92.7% to 95.3%, which suggests a good stability of the ALD-passivated perovskite. The maximum power point tracking result is shown in Fig. S7,† in which the PSCs with and without ALD-Al<sub>2</sub>O<sub>3</sub> both have high operational stability. The incident photon-to-electron conversion efficiency (IPCE) spectrum was also measured for the ALD-passivated champion cell. As shown in Fig. 3e, the spectrum shows a rather flat plateau at ~90% over the wavelength range from 380 to 760 nm, and the integrated current density was found to be 22.34 mA cm<sup>-2</sup>, which agrees well with the cell  $J-V$  curve of sample 75 °C ALD in Fig. 3b.

To further understand the charge transport at the perovskite/Al<sub>2</sub>O<sub>3</sub> interface, we carried out both steady-state photoluminescence spectroscopy (PL) and time-resolved photoluminescence spectroscopy (TRPL) experiments on the perovskite films with PVK/Al<sub>2</sub>O<sub>3</sub> and without ALD Al<sub>2</sub>O<sub>3</sub> (PVK).

The observed PL signals (Fig. 4a) reflect the radiative direct recombination of the photo-generated free carriers. With the presence of defects, the photo-generated carriers can be trapped at the defects and recombine non-radiatively *via* a Shockley–Read–Hall (SRH) process, resulting in poor PL efficiency.<sup>57,58</sup> As shown in Fig. 4a, compared to the bare perovskite film, the perovskite with ALD Al<sub>2</sub>O<sub>3</sub> shows appreciably enhanced PL intensity, which suggests that the defects at the MAPbI<sub>3</sub> surface were well passivated by the ALD of Al<sub>2</sub>O<sub>3</sub>. At the same time, a small blue-shift from 768 to 766 nm was also observed for the PL peak of the PVK/Al<sub>2</sub>O<sub>3</sub> sample, and this blue-shift could be well ascribed to the reduction of the defect traps, which recover the band edge transition.<sup>59,60</sup> Reinforcing evidence was also obtained from the TRPL results. Fig. 4b and c show the TRPL spectra of the samples with the laser illumination from the air side (*i.e.* the Al<sub>2</sub>O<sub>3</sub> side) and the glass side, respectively. Note that the light penetration depth (~100 nm (ref. 61)) is smaller than the perovskite layer thickness, and therefore, the TRPL spectra of different sides can provide information about the location of the defects. As shown in Fig. 4b and c, the TRPL curves were all fitted by a bi-exponential decay function, where a longer lifetime of decay corresponds to fewer defects.<sup>4,62,63</sup> Table S1† displays the detailed fitting results. For the air-side illumination, the average lifetime increases significantly from 20.88 ns to 73.39 ns after the ALD of Al<sub>2</sub>O<sub>3</sub>, which again suggests that the ALD of Al<sub>2</sub>O<sub>3</sub> can effectively passivate the defects on the perovskite surface. In comparison, the glass-side TRPL curves give out a smaller increase in the average lifetime from 49.42 ns to 77.52 ns after the ALD. These results suggest that most of the problematic defects are located near the top surface of the



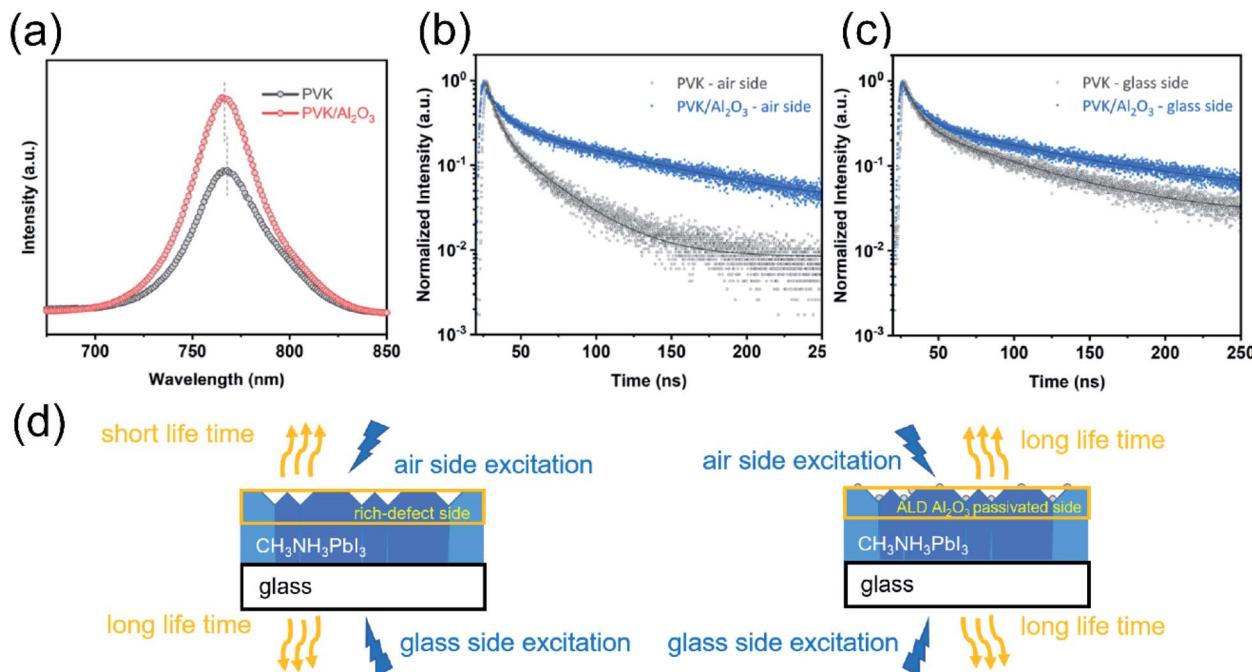


Fig. 4 (a) Steady-state photoluminescence (PL) spectra and (b and c) time-resolved photoluminescence (TRPL) spectra of the perovskite-on-glass samples with and without ALD Al<sub>2</sub>O<sub>3</sub>. (d) Schematic of the relationship between the carrier life time and laser excitation side.

perovskite and they can be effectively passivated by the ALD of Al<sub>2</sub>O<sub>3</sub> (Fig. 4d).<sup>64</sup>

## Conclusions

In summary, the atomic layer deposition (ALD) technique offers a generally promising approach to effectively passivate the perovskite surface defects to improve the power conversion efficiency and stability of PSCs. However, owing to the sensitive chemical reactivity of the perovskite surface, discrepant and conflicting observations of the ALD passivation were often reported in the literature. To unveil the involved crucial mechanism, we herein utilized the *in situ* XPS technique to directly investigate the evolution of a representative high-performance MAPbI<sub>3</sub> perovskite surface during the ALD process of Al<sub>2</sub>O<sub>3</sub>. For the ALD at 125 °C, we observed pronounced degradation of the perovskite, as there formed metallic lead at the perovskite/Al<sub>2</sub>O<sub>3</sub> interface and a considerable portion of the bulk perovskite converted to PbI<sub>2</sub>. As the ALD temperature was lowered to 25 or 75 °C, the degradation was largely minimized, and we observed the formation of CH<sub>3</sub>NH<sub>2</sub>, which could possibly heal the perovskite surface defects. The initial ALD growth at 75 °C was apparently slower than that at 25 °C, and therefore Al<sub>2</sub>O<sub>3</sub> grew more selectively on the surface defects, which was desirable to not impede the charge transport across the interface. Therefore, 75 °C was determined to be the best ALD temperature for perovskite passivation. Following this conclusion, inverted planar perovskite solar cells were prepared with the ALD interlayers. Indeed, the cells with the ALD Al<sub>2</sub>O<sub>3</sub> interlayer deposited at 75 °C exhibited a significantly enhanced PCE from 18.8% (champion 19.2%) to 20.0% (champion 20.4%), along

with improved  $V_{oc}$  and fill factor. In contrast, the ALD at 125 °C drastically deteriorated the cell PCE, which signifies the crucial importance to restrain the in-ALD-process degradation of the perovskites for the high-performance PSCs. Finally, we carried out steady-state and time-resolved PL to explicitly characterize the perovskite layer, and the results indicated that it was the ALD layer that reduced the defect states at the perovskite surface. The harmony of the PSC performance with the implications from the mechanism investigation well highlights the importance of the *in situ* interface investigation toward rational designing of the perovskite solar cells. Considering the great representativeness and broad applicability of MAPbI<sub>3</sub> and ALD Al<sub>2</sub>O<sub>3</sub>, the mechanism and strategy reported herein should be of significant value for the PSC interface engineering in general.

## Author contributions

R. Zhao, K. Zhang, and J. Zhu contributed equally to this work. Prof. S. Yang, S. Xiao and X. Wang supervised this work. R. Zhao, K. Zhang, J. Zhu and S. Xiao designed and conducted the experiments. X. Wang, R. Zhao and J. Zhu took charge of the work of ALD. R. Zhao measured and analysed the *in situ* XPS data, and wrote the manuscript. R. Zhao and J. Zhu optimized the ALD technology and analysed the relevant data. W. Xiong conducted measurement of SEM. S. Yang, S. Xiao and K. Zhang took charge of the work of perovskite solar cells. S. Xiao and K. Zhang conducted device fabrication, characterization and analysis of passivation mechanism, and prepare relevant data. J. Wang analysed defects of perovskite. T. Liu, G. Xing, and K. Wang carried out the measurement and analysis of TRPL. All authors participated in the writing or revise of the manuscript.



## Conflicts of interest

The authors declare no conflict of interest.

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