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# Enhancement of thermal stability for perovskite solar cells through cesium doping†

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Organic–inorganic hybrid perovskite solar cells are found to be sensitive to moisture, oxygen, UV light, light soaking, heat, electric field, etc. Among all these factors, thermal stability is one of the most challenging concerns affecting PSCs stability, since it is hard to avoid a temperature increase for solar cells during operation. In this work, we systematically studied the thermal stability of  $\text{Cs}_x\text{MA}_{1-x}\text{PbI}_3$  film and solar cells. The introduction of Cs into the precursor solution would inevitably accelerate the film deposition rate, resulting in decreased grain size and more Cs atoms in the film than in the precursors. The study on thermal stability illustrated that perovskite degradation was highly related to the amount of oxygen in the air. A small amount of Cs doping ( $x = 0.09$ ) was beneficial for better thermal stability. In addition, Cs doping also enhanced the device performance. The improvement of short-circuit currents came from the increased film thickness, which was due to the faster deposition rate for Cs doped samples. Besides, Cs doping was vital to suppress the trap states in the film since the trap states were related to halide deficiency during thermal annealing. At last, the final performance of  $\text{Cs}_{0.09}\text{MA}_{0.91}\text{PbI}_3$  reached 18.1%, with a  $J_{\text{SC}}$  of  $22.57 \text{ mA cm}^{-2}$ ,  $V_{\text{OC}}$  of 1.06 V, FF of 0.76.

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## Introduction

The last several years have witnessed the rapid development of organic–inorganic perovskite materials with striking photonic and electronic properties, as well as successful demonstrations of their applications in solar cells, light emitting diodes, photodetectors, and lasers. The power conversion efficiency (PCE) of perovskite solar cells (PSCs) has increased from 3.81% to 22.1% in seven years.<sup>1–4</sup> During this period, many reports have focused on the film fabrication process, composition control, and device structure.<sup>5–8</sup>

Almost at the same time, many researchers found the instability of perovskites despite their excellent electronic properties.<sup>9,10</sup> As far as we are concerned, perovskites are found to be sensitive to moisture, oxygen, UV light, light soaking, heat, electric field, and other potential factors.<sup>11–13</sup> Thereby, massive efforts have been made to study the degradation mechanism and stability enhancement strategies. Moisture and oxygen could cause the formation of hydrate intermediates and damage the perovskite accordingly. The moisture and oxygen instability could be well avoided by device encapsulation, including fluoropolymer encapsulation or metal oxide to protect the perovskite.<sup>14,15</sup> In addition, PSCs could be protected from UV light through adding UV filters onto devices. The

stability toward an electric field, which is also denoted as the typically observed hysteresis phenomenon, could be solved by suppressing the trap states of perovskites, by using  $\text{PC}_{61}\text{BM}$  or other fullerene derivatives.<sup>16,17</sup>

Currently, from our views, thermal stability and light soaking are the most challenging concerns affecting PSCs stability, since it is hard to avoid temperature increase and light illumination for solar cells during operation. Formamidinium ( $\text{HC}(\text{NH}_2)_2$ ), abbreviated as FA, has been demonstrated more stable than methylammonium ( $\text{CH}_3\text{NH}_3$ , shortened as MA) under high temperature.<sup>18,19</sup> However, as an organic molecules, degradation could still be observed for FA-based perovskite, especially when oxygen was present, which was caused by the oxidation of FA. Moreover, FA is also more sensitive to moisture than MA due to its high hygroscopicity, thus requiring more rigorous encapsulation and increasing the cost. In terms of the intrinsic property, Cs-based perovskite should exhibit higher thermal stability than FA- and MA-counterparts. Some recent papers indeed found that  $\text{CsPbBr}_3$ ,  $\text{CsPbI}_2\text{Br}$ ,  $\text{CsPbIBr}_2$  could endure harsh conditions with temperature as high as  $300^\circ\text{C}$ .<sup>20–22</sup> However, due to the non-ideal band gaps ( $>1.7 \text{ eV}$ ), the PCE for Cs-based perovskite is relatively lower than  $\text{MAPbI}_3$  and  $(\text{MAPbBr}_3)_x(\text{FAPbI}_3)_{1-x}$ . For single junction solar cells, researchers started to utilize mixture of Cs–MA, Cs–FA, Cs–MA–FA etc., to compose perovskite.<sup>23–25</sup> The role of Cs was found to be effective to stabilize the black phase of  $\text{FAPbI}_3$ , decrease trap states, and benefit device performance. However, there is still no detailed studies on thermal stability of Cs-based perovskite materials.

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XRD patterns and SEM images were used to study the composition and morphology change of the film before and after thermal degradation. As shown in Fig. 2a, all the films demonstrated strongest intensity along  $\langle 110 \rangle / \langle 002 \rangle$  directions. Consistent with our previous results, as Cs doping increased, the peak for  $(112) / (200)$  planes became more obvious, which was due to that Cs atoms would preferentially precipitate and result in preferred crystal orientation along the more thermodynamically stable directions of  $\langle 112 \rangle / \langle 200 \rangle$ .<sup>23</sup> After thermal treatment (Fig. 2b), there was an additional new peak at  $12.6^\circ$ , which attributed to  $\text{PbI}_2$ . For the control sample ( $x = 0$ ), the peak for

## Results and discussions

Firstly, we prepared perovskite films with different Cs content. During the experiments, one-step method was used to construct the films, since according to our experience, two-step deposition typically led to needle-like crystals atop the film. CsI, MAI, and  $\text{PbI}_2$  were mixed together in the precursor solution, dissolving into a mixed solvent of GBL : DMSO = 7 : 3. Cs salts were hard to dissolve, thereby we heated the whole solution at 60 °C. X-ray photoelectron spectroscopy (XPS) was used to quantitatively study the content of Cs in the films (Fig. S1†). As shown in Table 1, Cs content in the films were always higher than the precursors, which was due to the preferentially

| Cs content in precursors<br>(Cs/Cs + MA) | Pb    | N     | I     | Cs    | Cs content in<br>films (Cs/Cs + MA) |
|------------------------------------------|-------|-------|-------|-------|-------------------------------------|
| 0                                        | 10.5% | 11.5% | 44.2% | NA    | NA                                  |
| 5%                                       | 10.7% | 10.2% | 44.2% | 1%    | 9%                                  |
| 10%                                      | 11%   | 9%    | 46%   | 1.7%  | 15.9%                               |
| 15%                                      | 10%   | 9.5%  | 43.4% | 2.8%  | 22.7%                               |
| 20%                                      | 10%   | 4.5%  | 37.5% | 4%    | 47%                                 |
| 30%                                      | 10%   | 2.7%  | 36.9% | 7%    | 72%                                 |
| 40%                                      | 12%   | 1.5%  | 43%   | 9.36% | 86%                                 |

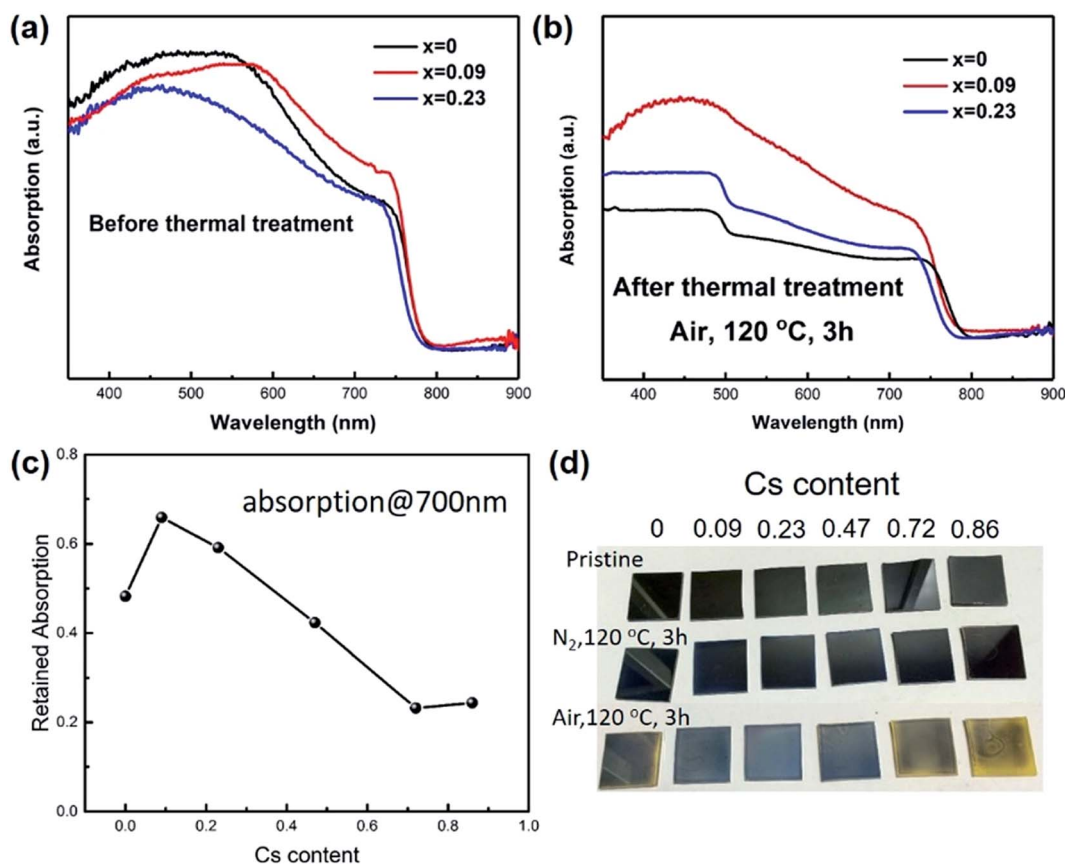


Fig. 1 Thermal stability study of perovskite films with different Cs content. (a) Perovskite films with compositions of  $\text{Cs}_x\text{MA}_{1-x}\text{PbI}_3$  before thermal treatment, where  $x$  represents Cs content. (b) Perovskite films after thermal treatment at 120 °C for 3 hours exposing to air. (c) Retained absorption at 700 nm for perovskite films with  $x$  ranging from 0 to 0.86. (d) Photographs of perovskite films exposing to different atmosphere.

$\text{PbI}_2$  was much higher than perovskite, demonstrating most of  $\text{MAPbI}_3$  has degraded to  $\text{PbI}_2$ . For  $x = 0.09$ , the formation of  $\text{PbI}_2$  was effectively suppressed, due to the better thermal stability. For  $x = 0.23$ , in contrast, the peak for perovskite was even lower than that of  $x = 0$ .

Fig. 2c and d shows the morphology of the films. Before thermal treatment, all the films were flat with no pinholes, while Cs content could lead to smaller grain size, probably due to the increased nucleation sites in the early stage of film formation. In contrast, after thermal treatment, the films became rugged, especially for  $x = 0$ . The pits and pinholes were caused by the loss of MAI or  $\text{CH}_3\text{NH}_2$ , leaving the film recrystallized as  $\text{PbI}_2$ . When Cs was introduced into the film, the pinholes were effectively suppressed. In addition, as Cs content increased, the pinholes became smaller due to the much more stable A site (Cs) in the films and decreased crystal volume.

We assembled perovskite solar cells with  $x = 0$  and  $x = 0.09$ , since if  $x > 0.09$ , the thermal stability was not good enough. As shown in Fig. 3a and b, the cross-sectional SEM images exhibited that the thickness of perovskite ( $x = 0.09$ ) was  $\sim 500$  nm, including the mesoporous  $\text{TiO}_2$  layer and top perovskite layer, higher than that of  $x = 0$  (405 nm). During the experiment, all the parameters were kept the same except the composition of precursor solution. The higher thickness of  $x = 0.09$  was probably due to the increased film formation rate when

Cs was introduced into the film. If the film formation rate was fast, more perovskite would precipitate out when chlorobenzene was added onto the film to quench the precursor solution. Due to the increased film thickness, the short-circuit current ( $J_{\text{SC}}$ ) of  $x = 0.09$  was higher than that for  $x = 0$ . As shown in Table 2 and Fig. 3c, the champion device of  $x = 0.09$  showed a  $J_{\text{SC}}$  of  $22.57 \text{ mA cm}^{-2}$ , a  $V_{\text{OC}}$  of 1.06 V, FF of 0.76, leading to a final power conversion efficiency of 18.1%. In contrast, the champion device of control sample exhibited a  $J_{\text{SC}}$  of  $20.59 \text{ mA cm}^{-2}$ , a  $V_{\text{OC}}$  of 1.04 V, FF of 0.74, and a PCE of 15.8%. Fig. 3d shows the IPCE spectra of the champion devices. Although Cs atoms could inevitably increase the band gap of perovskite, the absorption region was almost the same for  $x = 0$  and 0.09 due to the small doping ratio. The integrated photocurrent for  $x = 0$  and 0.09 was 18.6 and  $19.9 \text{ mA cm}^{-2}$ , respectively, slightly lower than the value derived from  $J$ - $V$  curves. The reason is that our devices exhibited hysteresis effect, while IPCE recorded the steady state current. We have also averaged data from 12 devices to obtain statistic information. The average  $J_{\text{SC}}$  was increased from  $20.34 \pm 1.20 \text{ mA cm}^{-2}$  for  $x = 0$  to  $22.42 \pm 0.33 \text{ mA cm}^{-2}$  for  $x = 0.09$ . Additionally, besides the increase of  $J_{\text{SC}}$  after Cs introduction, the fill factors have also been increased from  $0.70 \pm 0.02$  to  $0.74 \pm 0.01$ . During this work, we employed mesoporous  $\text{TiO}_2$  as anode and found hysteresis effect for the fabricated devices (Fig. S3†). Further improvement on





In order to explain the charge recombination effect by Cs introduction, we measured XPS spectra for the films (Fig. 4a). There are two main peaks for Pb 4f spectra, assigning to Pb 4f<sub>7/2</sub> and Pb 4f<sub>5/2</sub>. For  $x = 0$  sample, the peaks at 138.4 eV and 143.2 eV are due to Pb element from perovskite. The presence of

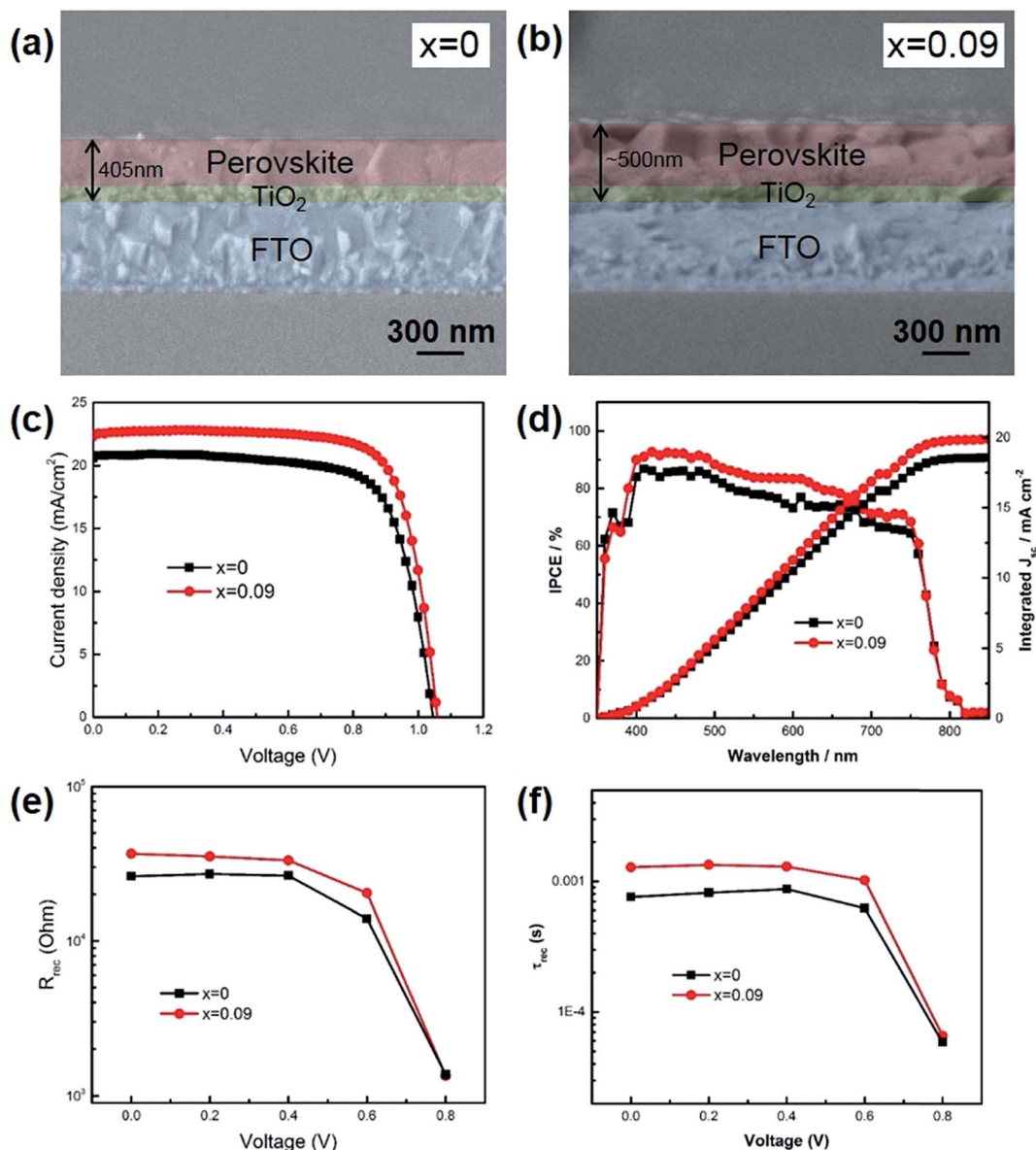


Fig. 3 (a, b) Cross-sectional SEM images of perovskite films with  $x = 0$  and 0.09. (c)  $J$ - $V$  characteristics of champion devices with different Cs content. (d) IPCE and integrated photocurrent for the champion devices. (e) Recombination resistance derived from electrochemical impedance spectra (EIS). (f) Electron recombination time from EIS spectra.

Table 2 Performance summary for perovskite solar cells with different Cs content

|            | $J_{\text{SC}}$ ( $\text{mA cm}^{-2}$ ) | $V_{\text{OC}}$ (V) | FF              | $\eta$ (%)     |
|------------|-----------------------------------------|---------------------|-----------------|----------------|
| $x = 0$    | $20.34 \pm 1.2$                         | $1.05 \pm 0.01$     | $0.70 \pm 0.02$ | $14.9 \pm 1.3$ |
| Champion   | 20.59                                   | 1.04                | 0.74            | 15.8           |
| $x = 0.09$ | $22.42 \pm 0.33$                        | $1.05 \pm 0.01$     | $0.74 \pm 0.01$ | $17.5 \pm 0.5$ |
| Champion   | 22.57                                   | 1.06                | 0.76            | 18.1           |

small peaks at 136.9 eV and 141.3 eV could be assigned to metallic Pb ( $\text{Pb}^0$ ), which was due to unsaturated Pb, according to recent studies. Henry Snaith and co-workers demonstrated that the presence of unsaturated Pb atoms was related to the iodide deficiencies, and metallic lead species could act as

recombination sites, leading to poor performance.<sup>26</sup> In our assumption, during thermal annealing process, the loss of iodide was accompanied with the loss of methyl ammonium. When we replaced some MA with Cs atoms, due to the better thermal stability and thus less loss of molecular groups from A site and iodide atoms, unsaturated Pb was effectively suppressed. For XPS spectra of  $x = 0.09$ , there are no additional peaks from  $\text{Pb}^0$ . Furthermore, we also measured transient photoluminescence spectra to study the defect density in the film (Fig. 4b). The decay curves could be well fitted by a biexponential decay function. The fast component is attributed to the surface recombination, while the slow component is caused by the recombination in the bulk of perovskite. We average the two components according to their amplitude to obtain the



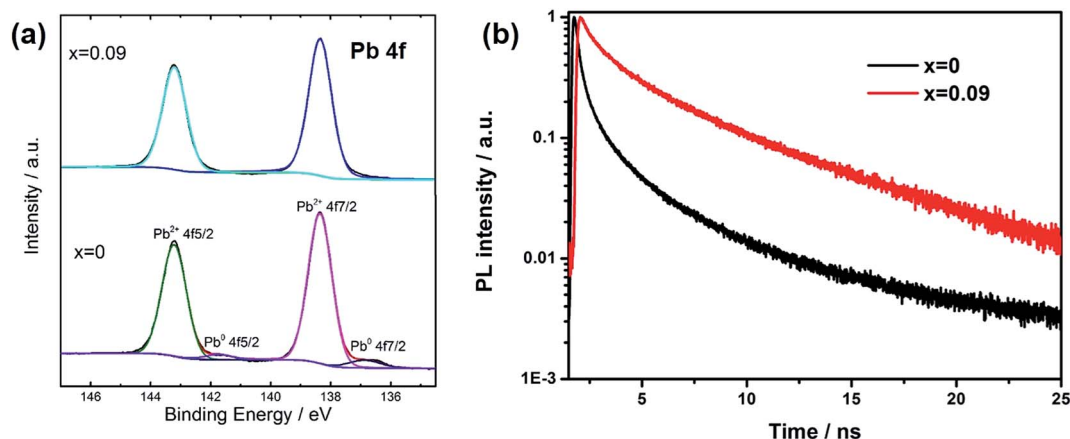


Fig. 4 (a) XPS study of Pb4f for  $x = 0$  and 0.09 films. (b) Transient PL spectra for perovskite films with different Cs content.

lifetime of carriers. The average corresponding decay time was 1.9 and 2.9 ns for  $x = 0$ , and 0.09 respectively, as shown in the inset of Fig. 4b. The increased lifetime when Cs was introduced into the film reflected the reduction of trap density, which was indicative of fewer non-radiative recombination sites.

At last, we compared the thermal stability of assembled perovskite solar cells for  $x = 0$  and 0.09, as shown in Fig. 5. The accelerated stability of the devices is evaluated in air at 85 °C without encapsulation. Each time before measurement for  $J-V$  curves, the devices were cooled down to room temperature naturally. For  $x = 0$ , less than 40% of the original performance was maintained after 60 min aging test. In contrast, for  $x = 0.09$ , nearly 80% of the initial performance was retained after the same durability test. The data were averaged from 6 devices in one batch. There are two causes for the performance declination, one is coming from the degradation of perovskite, and the other one is from the degradation of spiro-MeOTAD under heat treatment. The glass transition temperature of spiro-MeOTAD is around 125 °C.<sup>11</sup> We believe mixture of Cs and MA indeed

improved the thermal stability. However, the existence of spiro-MeOTAD would inevitably lower the overall stability. Better stability toward thermal conditions could be obtained by replacing spiro-MeOTAD with other stable hole transport materials.

## Conclusions

In a word, we have successfully fabricated perovskite films with composition of  $\text{Cs}_x\text{MA}_{1-x}\text{PbI}_3$ . The introduction of Cs into the precursor solution would inevitably accelerate the film deposition rate. The results were that the grain size became smaller and the film contained more Cs atoms than in the precursors. The study on thermal stability illustrated that perovskite degradation was highly related with oxygen in air. Small amount of Cs doping ( $x = 0.09$ ) was beneficial for the better thermal stability. In addition, Cs doping also enhanced the device performance. The improvement of short-circuit currents came from the increased film thickness, which was due to the faster deposition rate for Cs doped sample. Besides, Cs doping was vital to suppress the trap states in the film since the trap states was related to the halide deficiency during thermal annealing. At last, the final performance of  $\text{Cs}_{0.09}\text{MA}_{0.91}\text{PbI}_3$  reached 18.1%, with a  $J_{\text{SC}}$  of 22.57  $\text{mA cm}^{-2}$ ,  $V_{\text{OC}}$  of 1.06 V, FF of 0.76. In addition, the thermal stability of the unencapsulated devices for  $x = 0.09$  was also significantly improved compared to  $\text{MAPbI}_3$ .

## Experimental details

### Solar cell fabrications

The  $\text{TiO}_2$  compact layer was deposited onto FTO glass through atomic layer deposition (Beneq TFS 200) documented in previous reports. The mp- $\text{TiO}_2$  layer was prepared by depositing nanocrystalline  $\text{TiO}_2$  paste (18NRT from Dyesol Company; diluted to w/w 14.3%) onto the compact layer at 6000 rpm for 30 s, followed by heating at 500 °C for 1 h. The perovskite precursors were prepared by dissolving specific amounts of  $\text{PbI}_2$ , MAI, and CsI in the mixed solvent ( $\gamma$ -butyrolactone : DMSO =

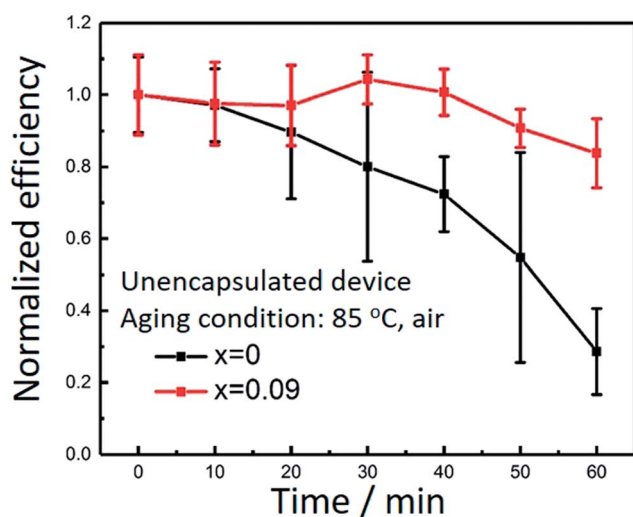


Fig. 5 Thermal stability of assembled perovskite solar cells from Cs doping and undoped perovskite.



7 : 3 vol%). The total concentration of Pb was set as 0.96 M for all the solutions. The solutions were then coated onto the substrate by two consecutive spin-coating steps, at 1500 rpm for 10 s, and 5000 rpm for 30 s. During the second step, 0.3 mL chlorobenzene was poured onto the substrate. Then the film was heated at 90 °C for 10 min. Spiro-MeOTAD solution was prepared by dissolving 102.7 mg spiro-MeOTAD in 1 mL chlorobenzene, to which 9.85  $\mu$ L 4-*tert*-butyl pyridine and 42.22  $\mu$ L lithium bis(trifluoromethanesulfonyl)imide solution (170 mg LITFSI in 1 mL acetonitrile) were added. Spiro-MeOTAD was deposited on the substrate at 2000 rpm for 45 s. Then the films were left in air overnight. Finally, 50 nm gold electrode were thermally evaporated under vacuum of  $\sim 10^{-6}$  Torr, at a rate of  $\sim 0.2$  Å s $^{-1}$ .

### Characterization

X-ray diffraction (XRD) spectra was measured with smart LAB instruments Cu K $\alpha$  beam ( $\lambda = 1.54$  Å). UV-Vis absorption spectra were obtained with a Hitachi U-3010 spectroscope. SEM images were measured by JEOL JSM-7401F and TEM images by Hitachi HT7700 with an acceleration voltage of 100 kV. The composition of the film was measured by ICP-AES (Thermo IRIS intrepid II). *J*-*V* curves were measured by a Keithley 2400 source meter under one sun illumination (AM 1.5G, 100 mW cm $^{-2}$ ), simulated by a solar simulator (ORIEL 81193) calibrated with an NREL-calibrated silicon solar cell. External quantum efficiency (EQE) spectra were recorded by a setup, consist of a xenon light source, a monochromator, and a potentiostat. EIS were measured by an electrochemical workstation (Zahner, CIMPS) with frequency from 10 to 10 $^6$  Hz under applied voltage with an amplitude of 10 mV under light condition. The active area of each cell was 0.16 cm $^2$  with a mask of 0.09 cm $^2$ . Time-resolved PL decay was measured by a FV1200, excited at 488 nm.

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