Decoupling engineering of formamidinium–cesium perovskites for efficient photovoltaics

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ABSTRACT

Although pure formamidinium iodide perovskite (FAPbI₃) possesses an optimal gap for photovoltaics, their poor phase stability limits the long-term operational stability of the devices. A promising approach to enhance their phase stability is to incorporate cesium into FAPbI₃. However, state-of-the-art formamidinium–cesium (FA–Cs) iodide perovskites demonstrate much worse efficiency compared with FAPbI₃, limited by the different crystallization dynamics of formamidinium and cesium, which result in poor composition homogeneity and high trap densities. We develop a novel strategy of crystallization decoupling processes of formamidinium and cesium via a sequential cesium incorporation approach. As such, we obtain highly reproducible, highly efficient and stable solar cells based on FA₁₋ₓCsₓPbI₃ (x = 0.05–0.16) films with uniform composition distribution in the nanoscale and low defect densities. We also revealed a new stabilization mechanism for Cs doping to stabilize FAPbI₃, i.e. the incorporation of Cs into FAPbI₃ significantly reduces the electron–phonon coupling strength to suppress ionic migration, thereby improving the stability of FA–Cs-based devices.

Keywords: formamidinium–cesium, perovskite solar cell, decoupling engineering, sequential cesium incorporation, uniform composition distribution

INTRODUCTION

Metal–halide perovskites with superior photophysical properties and low-cost solution technology have emerged as promising candidates for different optoelectronic devices, including solar cells, light-emitting diodes, etc. [1–6]. For perovskite solar cells (PSCs), the certified power-conversion efficiency (PCE) has reached 25.7%, which is comparable to the current commercial crystalline silicon solar cells. ABX₃ perovskites with tailoring compositions, where A is an organic or inorganic cation, B is a metal cation and X is a halide anion, have been attempted for high efficiency and stable photovoltaic devices. Among these, formamidinium lead iodide (FAPbI₃) has exhibited great potential as the absorber layer, due to its optimal band gap of ∼1.5 eV and excellent thermal stability [7–9]. However, the photoactive FAPbI₃ phase would easily transform into a non-photoactive yolk–δ-FAPbI₃ phase at room temperature, especially under humid conditions [10,11]. The poor phase stability challenges both the efficiency and long-term stability of the PSCs based on FAPbI₃ [12,13]. It is generally believed that the phase instability of FAPbI₃ perovskites originates from its unsuitable tolerant factor. To address this problem, alloying FA⁺ with MA⁺/Cs⁺ cations or partially substituting I⁻ with Br⁻ ions has been employed to tune the tolerant factor [14,15]. The resulting mixed-ion FA-based perovskites exhibit improved resistance to phase transition.

Among these different alloying approaches, formamidinium–cesium mixed-cation pure iodide (FA₁₋ₓCsₓPbI₃) perovskites are particularly promising, because they avoid the concerns about volatile MA cations and phase segregation induced by mixed halide ions (Br–I) [16–20]. However, because of the complex crystallization kinetics of
formamidinium and cesium, these pure iodide FA–Cs perovskites fabricated by one-step crystallization suffer from poor composition homogeneity and high defects/traps densities [21,22]. The PSCs based on these films are therefore facing relatively low efficiencies. Especially, strong non-radiative recombination in all reported FA–Cs-based PSCs limited the open-circuit voltage (\(V_{oc}\)) of the resulting devices [23–25].

Herein, we develop a novel sequential Cs incorporation (SCI) strategy to decouple the crystallization processes of formamidinium and cesium, and achieve highly efficient pure iodide \(\text{FA}_1\_x\text{Cs}_x\text{PbI}_3\) \((x = 0.05–0.16)\) perovskites (denoted as \(\text{SCI-FA}_1\_x\text{Cs}_x\text{PbI}_3\)). The ratio of FA and Cs in \(\text{FA}_1\_x\text{Cs}_x\text{PbI}_3\) can be straightforwardly tuned by introducing different concentrations of cesium \((0.05, 0.09, 0.16)\) perovskites (denoted as SCI-FA0.91Cs0.09PbI3). As a result, the champion SCI-FA0.91Cs0.09PbI3 PSCs yield a record PCE of 24.7% (certified 23.8%) with improved \(V_{oc}\) and fill factor, which is the highest value for the pure iodide \(\text{FA}_1\_x\text{Cs}_x\text{PbI}_3\) perovskites. Compared with FAPbI3, the SCI-FA0.91Cs0.09PbI3 perovskite films show reduced electron–phonon coupling and lattice fluctuations, which suppress the formation of iodide-rich clusters and finally contribute to the excellent operational stability of the \(\text{FA}_0\_x\text{Cs}_x\text{PbI}_3\)-based PSCs.

**RESULT AND DISCUSSION**

Figure 1a shows the schematic diagram of SCI-FA1-xCsPbI3 perovskite films prepared by decoupling the crystallization processes of formamidinium and cesium. A FAPbI3 precursor film was first deposited by a typical anti-solvent method followed by annealing for 1 min. The Cs cation is sequentially introduced onto the FA perovskite film by spin-coating HCOOCs isopropanol (IPA) solution, followed by further annealing. For comparison, we employed different concentrations of HCOOCs solution \((2.5, 5\) and \(10\) mg mL\(^{-1}\)) to fabricate SCI-FA1-xCsPbI3 perovskites. The final ratios of incorporated Cs in the above SCI-FA1-xCsPbI3 perovskites films, i.e. the value of \(x\), are \(0.05, 0.09\) and \(0.16\), as confirmed by inductively coupled plasma–mass spectrometry (ICP–MS) analysis (Supplementary Table 1). The corresponding SCI-FA1-xCsPbI3 perovskite films are noted as \(x = 0.05\), \(x = 0.09\) and \(x = 0.16\) in Fig. 1.

Optical and structural measurements of perovskite films indicate that Cs\(^+\) from HCOOCs has successfully been incorporated into the lattice of FAPbI3 perovskites. Figure 1b shows the ultraviolet–visible (UV–vis) spectra of SCI-FA1-xCsPbI3 perovskite films, in which the absorption edges of SCI-FA1-xCsPbI3 perovskites strongly depend on the amount of Cs\(^+\) incorporation. When \(x\) increases from 0 to 0.16, the absorption edges of SCI-FA1-xCsPbI3 perovskites gradually blue-shift from 816 to 802 nm, and the corresponding photoluminescence (PL) peaks shift from 809 to 797 nm. The X-ray diffraction (XRD) measurements are carried out to investigate the crystal structure evolution of SCI-FA1-xCsPbI3 perovskites (Fig. 1c). All the SCI-FA1-xCsPbI3 perovskites exhibit stronger peak intensity than the pure FAPbI3 at around both 14° and 28°, corresponding to (001) and (002) perovskite crystal planes. The inset image of Fig. 1c shows that the peak between 13.8° and 14.1° shifts to a higher degree, indicating that Cs ions are incorporated into the perovskite lattice. The lattice parameter decreases with increasing the amount of Cs (Supplementary Fig. 1), further confirming the successful mixing of Cs\(^+\) in the perovskite lattice. The tolerance factor of SCI-FA1-xCsPbI3 perovskites is also reduced compared with pure FA perovskite, potentially contributing to an stable perovskite structure (Supplementary Fig. 2).

The Cs incorporation also significantly improves the film morphologies (Fig. 1d). All SCI-FA1-xCsPbI3 films show enlarged and pinhole-free grains compared with the FAPbI3 film, which shows coarse grains and pinholes. As shown in the cross-sectional scanning electron microscopy (SEM) images, SCI-FA1-xCsPbI3 perovskite films \((550–600\) nm) with vertical growth of grains benefit efficient charge extraction.

By adopting these Cs-incorporated perovskites as light absorber layers, we fabricate PSCs with a configuration of fluorine-doped tin oxide (FTO)/electron-transport layer/perovskite/hole-transport layer/Au. All SCI-FA1-xCsPbI3 \((x = 0.05, 0.09, 0.16)\)-based PSCs exhibit improved device efficiency compared with the FAPbI3-based devices (Supplementary Fig. 3). Considering that \(x = 0.09\) provides the optimal photovoltaic (PV) performance, we then chose this composition (denoted as SCI-FA0.91Cs0.09PbI3) for detailed investigations on Cs incorporation and its role on film and device properties.

X-ray photoelectron spectroscopy spectra are conducted to explore the effect of SCI on the elements and their chemical states in perovskite films. All core-level peaks are assigned to Cs, Pb,
Figure 1. Spectroscopic, structural and morphological characterizations of SCI-FA$_{1-x}$Cs$_x$PbI$_3$ films. (a) Schematic diagram of the SCI-FA$_{1-x}$Cs$_x$PbI$_3$ perovskite films fabricated by decoupling the crystallization processes of formamidinium and cesium. (b) UV–vis absorption and normalized PL spectra and (c) XRD patterns of FAPbI$_3$ and SCI-FA$_{1-x}$Cs$_x$PbI$_3$ perovskites. Inset pattern corresponding to the characteristic peaks of (001) perovskite crystal planes. (d) Top-surface SEM images of FAPbI$_3$ and SCI-FA$_{1-x}$Cs$_x$PbI$_3$ perovskite films. The inset presents the cross-sectional morphology of the corresponding perovskite films. Scale bars: 1 μm.

N and I (Supplementary Fig. 4) elements. The characteristic Cs signals in the SCI-FA$_{0.91}$Cs$_{0.09}$PbI$_3$ perovskite locate at 738.5 and 724.7 eV, and show a 1.1-eV shift compared with the Cs in HCOOCs. Such a large shift is attributed to the formation of chemical bonds between Cs$^+$ and [PbI$_6$]$^{4-}$. For the Pb 4$f$ spectra in FAPbI$_3$, two peaks corresponding to Pb 4$f_{7/2}$ and Pb 4$f_{5/2}$ are observed at 138.3 and 143.2 eV. However, in the SCI-FA$_{0.91}$Cs$_{0.09}$PbI$_3$ perovskite, both Pb 4$f_{7/2}$ and Pb 4$f_{5/2}$ shift by 0.1 eV toward higher binding energy, originating from the stronger bond energies between Cs$^+$ and [PbI$_6$]$^{4-}$ than that between FA$^+$ and [PbI$_6$]$^{4-}$. Additionally, the N and I elements representing the formamidine component have undergone a small shift. These results further confirm that Cs has been successfully incorporated into the FAPbI$_3$ perovskite lattice to form SCI-FA$_{0.91}$Cs$_{0.09}$PbI$_3$ perovskite.

For comparison, we also fabricate one-step crystallized 1S-FA$_{1-x}$Cs$_x$PbI$_3$, where Cs$^+$ (from HCOOCs) is directly mixed with FA$^+$ in the perovskite precursor. For 1S-FA$_{1-x}$Cs$_x$PbI$_3$ PSCs (Supplementary Fig. 5), the optimal efficiency is also obtained from the x = 0.09 sample, which hence will be used as the control sample for comparison with SCI-FA$_{0.91}$Cs$_{0.09}$PbI$_3$. The absorption edge, characteristic XRD peaks and surface morphologies of the 1S-FA$_{0.91}$Cs$_{0.09}$PbI$_3$ perovskite films are consistent with SCI-FA$_{0.91}$Cs$_{0.09}$PbI$_3$ perovskite (Supplementary Fig. 6).

The synchrotron-based grazing-incidence wide-angle X-ray scattering (GIWAXS) and time-of-flight secondary ion mass spectrometry (ToF-SIMS) measurements are further employed to explore the crystal structure and internal composition in perovskite films. As shown in Fig. 2a and b, there is a signal ring at $q_{xy} = 8.8$ nm$^{-1}$ corresponding to $\delta$-FAPbI$_3$ in 1S-FA$_{0.91}$Cs$_{0.09}$PbI$_3$ perovskite, while the GIWAXS result of SCI-FA$_{0.91}$Cs$_{0.09}$PbI$_3$ exhibits high phase purity and obvious crystal orientation.
Characterizations of FA0.91Cs0.09PbI3 perovskite films. GIWAXS data of (a) SCI-FA0.91Cs0.09PbI3 and (b) 1S-FA0.91Cs0.09PbI3 films. (c) 3D distribution in the SCI-FA0.91Cs0.09PbI3 and 1S-FA0.91Cs0.09PbI3 film by ToF-SIMS analysis, where the signal intensity in two kinds of perovskites is normalized. (d) TRPL decay curves of FAPbI3, SCI-FA0.91Cs0.09PbI3 and 1S-FA0.91Cs0.09PbI3 perovskite films.

Figure 2. Characterizations of FA0.91Cs0.09PbI3 perovskite films. GIWAXS data of (a) SCI-FA0.91Cs0.09PbI3 and (b) 1S-FA0.91Cs0.09PbI3 films. (c) 3D distribution in the SCI-FA0.91Cs0.09PbI3 and 1S-FA0.91Cs0.09PbI3 film by ToF-SIMS analysis, where the signal intensity in two kinds of perovskites is normalized. (d) TRPL decay curves of FAPbI3, SCI-FA0.91Cs0.09PbI3 and 1S-FA0.91Cs0.09PbI3 perovskite films.

without any obvious phase impurities. Since the distribution of Cs in SCI-FA1–xCsxBixPbI3 perovskite has a significant effect on both the phase stability and traps/defects [22,26], we proceed to investigate the distribution of Cs in the SCI-FA0.91Cs0.09PbI3 and 1S-FA0.91Cs0.09PbI3 perovskites. Both energy-dispersive spectroscopy mapping and ToF-SIMS have established that the incorporated Cs+ cations homogeneously distribute in the surface and bulk of SCI-FA0.91Cs0.09PbI3 perovskite, which is completely different from the inhomogeneous Cs+ cation distributions in 1S-FA0.91Cs0.09PbI3 perovskite (Fig. 2c and Supplementary Fig. 7). The Cs aggregation at the top surface of 1S-FA0.91Cs0.09PbI3 reveals the phase separation in SCI-FA0.91Cs0.09PbI3, which is consistent with the GIWAXS data. Such uniform distribution of Cs ions in the SCI-FA0.91Cs0.09PbI3 perovskite originates from the decoupled crystallization kinetics by the SCI strategy, which is beneficial for enhancing the phase stability and reducing defects/traps density. Different from Cs+ cations, other ions, including FA+, Pb2+ and I−, are uniformly distributed throughout both the SCI-FA0.91Cs0.09PbI3 and 1S-FA0.91Cs0.09PbI3 perovskite films (Supplementary Fig. 8). Although most of the HCOO− anions escape from the FA0.91Cs0.09PbI3 perovskite during the annealing process, residual traces of HCOO− still exist at the bottom interface between the perovskite and the substrate (Supplementary Fig. 9), which can further passivate the defects on the bottom interface and improve the charge carriers’ dynamics in the devices [27,28].

Uniform incorporation of Cs ions into FAPbI3 has two positive effects: enhanced phase stability and decreased trap/defect densities. The enhanced phase stability is evidenced from the absence of color/structural changes under humid conditions. Under 60% relative humidity, the SCI-FA0.91Cs0.09PbI3 perovskite maintains the black phase for 7 days without any changes (Supplementary Fig. 10), showing significant improvement compared with FAPbI3. The decreased trap/defect densities are demonstrated from photophysical measurements. The photoluminescence (PL) intensities of SCI-FA0.91Cs0.09PbI3 perovskites are much stronger than those of 1S-FA0.91Cs0.09PbI3 and pure FA perovskite films (Supplementary Fig. 11). In addition, the time-resolved PL (TRPL) spectra in Fig. 2d show that the PL lifetime (τ) of SCI-FA0.91Cs0.09PbI3 is much longer (413.84 ns) than that of 1S-FA0.91Cs0.09PbI3 (240.41 ns) and FAPbI3 (165.49 ns). Enhanced PL intensity and improved PL lifetime indicate that non-radiative recombination is suppressed in SCI-FA0.91Cs0.09PbI3, attributed to decreased trap densities. Consistently, we calculated the Urbach energy (Eu) according to the equation: A = A0 exp(E/Eu), where A is the absorbance, A0 is a constant for data fitting and E is the excitation energy. The Urbach energy (Supplementary Fig. 12) is decreased from 22.2 meV in FAPbI3 and 21.6 meV in 1S-FA0.91Cs0.09PbI3 to 18.3 meV in SCI-FA0.91Cs0.09PbI3. The smaller Urbach energy in SCI-FA0.91Cs0.09PbI3 corresponds to a lower density of trap states. These results suggest that, with sequential incorporation of Cs+, the SCI-FA0.91Cs0.09PbI3 perovskite exhibits significantly reduced non-radiative recombination via defects/traps.

Benefitting from these advantages of crystallization decoupling engineering, the resulting SCI-FA0.91Cs0.09PbI3 perovskite shows much enhanced device performance. Figure 3a compares the current density–voltage (J–V) characteristics of champion PSCs based on FAPbI3, SCI-FA0.91Cs0.09PbI3 and 1S-FA0.91Cs0.09PbI3 perovskites, respectively. The SCI-FA0.91Cs0.09PbI3-based PSC exhibits an impressive PCE of 24.7% compared with 22.6% for FAPbI3, representing a new record for pure iodide FA1–xCsxBixPbI3-based PSCs (Supplementary Table 2). A certified PCE of 23.8% with negligible hysteresis has been obtained in the SCI-FA0.91Cs0.09PbI3-based PSCs (Supplementary Fig. 13). The most striking difference is the Voc, which increases from 1.09 V in FAPbI3 to 1.18 V in SCI-FA0.91Cs0.09PbI3. The counterpart PSCs based on 1S-FA0.91Cs0.09PbI3 perovskite yield a
Figure 3. Photovoltaic and device characterization. (a) The J–V curves of the champion devices of FAPbI3, SCI-FA0.91Cs0.09PbI3, and 1S-FA0.91Cs0.09PbI3-based PSCs with 0.1 cm² of effective cell area. (b) IPCE spectra of FAPbI3, SCI-FA0.91Cs0.09PbI3, and 1S-FA0.91Cs0.09PbI3-based PSCs. (c) The PV performance distribution of FAPbI3-, SCI-FA0.91Cs0.09PbI3-, and 1S-FA0.91Cs0.09PbI3-based PSCs from 18 devices, respectively. (d) Steady-state efficiency of FAPbI3 and SCI-FA0.91Cs0.09PbI3 and 1S-FA0.91Cs0.09PbI3-based PSCs. (e) J–V characteristics of PSCs based on FAPbI3, SCI-FA0.91Cs0.09PbI3, and 1S-FA0.91Cs0.09PbI3, with 1 cm² of effective cell area under simulated AM 1.5-G solar illumination of 100 mW cm⁻² in the reverse scan. (f) EQE EL of FAPbI3-, SCI-FA0.91Cs0.09PbI3-, and 1S-FA0.91Cs0.09PbI3-based PSCs vs. the current density.

lower efficiency of 23.1%. This comparison indicates that our crystallization decoupling engineering is beneficial for enhancing SCI-FA1.3CsPbI3-based PSCs performance. The incident photon to electron conversion efficiency (IPCE) (Fig. 3b) is similar for both devices, with a high value of >90% in the wavelength range of 450–650 nm. The short-circuit current density (Jsc) of the SCI-FA0.91Cs0.09PbI3 device is slightly decreased compared with the FAPbI3 device, mainly due to slight increase in the band gap upon Cs incorporation. Figure 3c compares the PV parameters of FAPbI3-, SCI-FA0.91Cs0.09PbI3-, and 1S-FA0.91Cs0.09PbI3-based PSCs for 18 devices, respectively, indicating that sequential Cs incorporation also improves the device reproducibility. In addition, the SCI-FA0.91Cs0.09PbI3-based PSCs exhibit a smaller hysteresis (Supplementary Figs 13 and 14), resulting in a stabilized output power of 24.4% (Fig. 3d).

In addition to small-area PSCs, the large-area PSCs based on these sequential Cs-incorporated perovskite films also exhibit significantly improved device performance. The champion SCI-FA0.91Cs0.09PbI3 device, fabricated on 2.5 × 2.5 cm² substrates with an effective cell area of 1 cm² (Fig. 3e), displays a PCE of 22.4%, which is far higher than the FAPbI3 (~19.9%) and 1S-FA0.91Cs0.09PbI3 (~20.7%)-based devices.

The significantly enhanced Voc of the SCI-FA0.91Cs0.09PbI3 device is mainly due to suppressed non-radiative recombination, which can be quantified by measuring the external quantum efficiency of electroluminescence (EQE EL) values [29]. As shown in Fig. 3f, at the injection current densities corresponding to Jsc, the EQE EL value of the SCI-FA0.91Cs0.09PbI3 device is 6.38%, while that of the FAPbI3 device is 0.16%. We calculate the voltage losses due to non-radiative recombination (ΔVoc non-rad) based on the formula [30]:

$$\Delta V_{oc \text{ non-rad}} = -\frac{kT}{q} \ln \text{EQE}_{EL}$$

where k, T, and q represent the Boltzmann constant, temperature, and elementary electric charge, respectively. The difference in ΔVoc non-rad (0.09 V) matches well with the difference of device Voc (0.09 V).

Suppressed non-radiative recombination in the SCI-FA0.91Cs0.09PbI3 device is consistent with previous photophysical measurements on the films, which indicate that the sequential Cs incorporation can reduce the defects/traps. Further measurements on the devices also reach similar conclusions. The trap-filled limiting voltage in the space-charge limited current measurements decreases from 0.13 V in the FAPbI3 device to 0.09 V in SCI-FA0.91Cs0.09PbI3 device (Supplementary Fig. 15), indicating suppressed traps/defects upon Cs sequential incorporation [31,32]. These results are also consistent with transient photovoltage (TPV) decay and transient photocurrent (TPC) decay results (Supplementary
In summary, we successfully develop a novel SCI strategy to tackle the critical challenge of different crystallization dynamics of different cations in developing FA\textsubscript{1-x} Cs\textsubscript{x} PbI\textsubscript{3} perovskite PSCs. The resulting pure iodide SCI-FA\textsubscript{1-x} Cs\textsubscript{x} PbI\textsubscript{3} perovskites show more uniform composition distribution and reduced defects/traps density than FAPbI\textsubscript{3} and one-step crystallized 1S-FA\textsubscript{0.91} Cs\textsubscript{0.09} PbI\textsubscript{3}. Compared with FAPbI\textsubscript{3}, the SCI-FA\textsubscript{0.91} Cs\textsubscript{0.09} PbI\textsubscript{3} exhibits reduced electron–phonon coupling and lattice fluctuations, minimizing ion migration and hence enhancing the stability. As such, we have been able to achieve highly stable PSCs with a high efficiency of 24.7%, which is a record for SCI-FA\textsubscript{1-x} Cs\textsubscript{x} PbI\textsubscript{3} PSCs. This work opens up new possibilities to develop high-quality PSCs with enhanced stability.

**CONCLUSION**

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mixed-cation perovskites, presenting a milestone towards the development of highly efficient and highly stable perovskites for various applications, including solar cells, light-emitting diodes and lasers.

METHODS
Materials
Lead iodide (PbI₂, 99.9985%), cesium formate (HCOOCs, 98%), methylamine hydrochloride (MACl, 99%) and tin (IV) oxide colloid precursor (SnO₂, 15% in H₂O colloidal dispersion) were purchased from Alfa Aesar. Formamidinium iodide (FAI) was purchased from Xi’an Polymer Light Technology Corp. Other materials were purchased from Sigma-Aldrich and used as received without any purification. N,N-dimethylformamide (DMF, anhydrous, 99.8%), dimethyl sulfoxide (DMSO, anhydrous, 99.7%), chlorobenzene (anhydrous, 99.8%) and isopropanol alcohol (IPA, 99.5%) were purchased from J&K Scientific Ltd. Ammonium nitrate (Natl Sci Rev 2022, Vol. 9, nwac127)

Device fabrication
A compact TiO₂ layer (20 nm) was deposited using the spray pyrolysis method using a titanium bis(ethyl acetoacetate)-disopropoxide/1-butanol solution (1:9 volume ratio). The cleaned patterned fluorine-doped tin oxide (FTO, 7 Ω sq⁻¹) substrate was placed on a 450°C plate during the spray process followed by 1 h of annealing. The SnO₂ colloid precursor/ammonium solution (1:9 volume ratio) was spin-coated on a TiO₂ layer at 3000 rpm for 30 s followed by annealing for 30 min at 180°C. The SCI-FAₓ₋ₓCsₓPbI₃ perovskite precursor was prepared by mixing PbI₂, FAI and MACl (30 mol%) in DMF:DMSO = 9:1 solution to form 1.5 M FAPbI₃ precursor. The perovskite precursor films were deposited by spin-coating on FTO/TiO₂/SnO₂ substrate at 5000 rpm for 15 s. During spin-coating, 150–200 µL of chlorobenzene was dripped at the end of 10 s. The perovskite films annealed at 150°C for 10 min and 100°C for 10 min. The spiro-OMeTAD layer was spin-coated by dissolving 90 mg of spiro-OMeTAD in 1 mL of chlorobenzene and mixing with 39.5 µL of 4-tert-butylpyridine (TBP), 23 µL of Li-bis(trifluoromethanesulfonyl) imide (Li-TFSI)/acetonitrile (520 mg/mL) and 10 µL of tris(2-(1h-pyrazol-1-yl)-4-tert-butylpyridine)-cobalt(III)tris(bis(trifluoromethylsulfonyl)imide) (FK209)/acetonitrile (375 mg mL⁻¹) at 4000 rpm for 25 s. Finally, a 80-nm thick gold electrode was thermal evaporated on the spiro-OMeTAD layer. The perovskite films fabrication was operated in a dry box with <15% humidity.

SUPPLEMENTARY DATA
Supplementary data are available at NSR online.

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AUTHOR CONTRIBUTIONS
Y.Z. and F.G. designed and directed the study. H.C. and Y.F. conceived and performed the device fabrication work. Y.W. performed photophysical measurements and analysed the results. H.C., Y.F., Y.W., Y.C., Y.M., Z.Q., X.W., X.L., K.Z., F.G. and Y.Z. participated in characterization and data analysis. All authors contributed to the discussions. H.C., Y.F., Y.W., F.G. and Y.Z. wrote the manuscript with input from all authors. All authors reviewed the paper.

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