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Optoelectronic properties and interfacial interactions of two-dimensional $\text{Cs}_2\text{PbX}_4\text{-MSe}_2$ ($\text{M} = \text{Mo}, \text{W}$) heterostructures

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Constructing 2D inorganic perovskites and TMDs heterostructures is an effective method to design stable and high-performance perovskites optoelectronic applications. Here, we investigate the optoelectronic properties and interfacial interactions of $\text{Cs}_2\text{PbX}_4\text{-MSe}_2$ ($\text{X} = \text{Cl}, \text{Br}, \text{I}$; $\text{M} = \text{Mo}, \text{W}$) heterostructures using first-principles calculations. Firstly, six $\text{Cs}_2\text{PbX}_4\text{-MSe}_2$ interfaces remain stable in energy. With the halogen varying from Cl to I, the interlayer distances of $\text{Cs}_2\text{PbX}_4\text{-MSe}_2$ heterostructures increase rapidly. The CBM and VBM of monolayer Cs_2PbX_4 are all higher than that of monolayer MSe_2 and the charges transfer from Cs_2PbX_4 interfaces to MSe_2 interfaces when they contact. Both $\text{Cs}_2\text{PbX}_4\text{-MSe}_2$ heterostructures are type-II heterostructures, which can drive the photogenerated electrons and holes to move in opposite directions. What's more, $\text{Cs}_2\text{PbCl}_4\text{-MoSe}_2$ heterostructures exhibit the highest charge transport efficiency among $\text{Cs}_2\text{PbX}_4\text{-MoSe}_2$ heterostructures because $\text{Cs}_2\text{PbCl}_4\text{-MoSe}_2$ heterostructures have the lowest exciton binding energies among $\text{Cs}_2\text{PbX}_4\text{-MSe}_2$ heterostructures. In addition, the optical absorptions of all heterostructures are significantly higher than the corresponding Cs_2PbX_4 monolayers and MSe_2 monolayers. The construction of $\text{Cs}_2\text{PbX}_4\text{-MoSe}_2$ heterostructures is beneficial for improving the photoelectric performance of two-dimensional perovskite devices. Lastly, we found that the $\text{Cs}_2\text{PbI}_4\text{-WSe}_2$ heterostructure has the largest PCE (18%) among $\text{Cs}_2\text{PbX}_4\text{-MSe}_2$ heterostructures. The $\text{Cs}_2\text{PbCl}_4\text{-MoSe}_2$ heterostructure exhibits great potential application in photodetector devices and the $\text{Cs}_2\text{PbI}_4\text{-WSe}_2$ heterostructure has great potential application in solar cells.

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Introduction

Solar energy is a renewable and clean energy source that can be used in the field of solar cells. In the early years, perovskites were considered to be one of the most promising solar cell materials.¹ Three-dimensional (3D) lead halide perovskites have been innovatively applied in the field of solar cells with power conversion efficiency (PCE) up to 25.6%.² The high PCE of 3D lead halide perovskites is because of their excellent photoelectric properties, such as high carrier mobility, large light absorption coefficient, long carrier diffusion length and strong photoluminescence.^{3,4} Although the three-dimensional lead halide perovskites have great photoelectric properties, their instability in air limits their widespread use for

commercialization.^{5,6} In recent years, two-dimensional (2D) inorganic halide perovskites with good moisture resistance and stability have attracted attention for optoelectronic applications.⁷ Currently, many theoretical studies and experiments on 2D perovskites have been conducted. For example, Ding *et al.* reported that the number of layers would affect the optical absorption and transport properties of 2D perovskite Cs_2PbI_4 .⁸ In addition, Bala *et al.* also revealed the same laws that the band gap and optical properties of 2D $\text{Cs}_{n+1}\text{Pb}_n\text{X}_{3n+1}$ ($\text{X} = \text{Cl}, \text{Br}, \text{I}$) perovskite vary with layer number.⁹ In experiments, Song *et al.* fabricated 2D CsPbBr_3 nanosheets and used them to produce high-performance photodetectors.¹⁰ These studies found that the bandgaps of 2D monolayer lead halide perovskites are larger than 3D lead halide perovskites, which lead to lower electric properties, optical absorption and low performance of optoelectronic applications.

Fortunately, two-dimensional monolayer lead halide perovskites have a wide range of photoelectric properties adjustable by replacing halogens, changing layers, and working with other two-dimensional materials. In particular, the halide perovskites grown in the other 2D materials provides excellent ability to further adjust band gaps, transport properties, charge carrier dynamics, chemical stability and optical light absorption.¹¹

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Therefore, a great deal of scientific researches of 2D halide perovskites and 2D materials interface engineering have been carried to achieve low-cost, high efficiency, stable heterostructures photoelectric devices. For instance, He *et al.* theoretically studied the effects of different terminals of heterojunctions and found that CsPbI₃/MoS₂ heterostructure had higher electric performances than CsPbI₃/WS₂.¹² The WS₂-CsPbBr₃ heterostructure was applied to perovskite solar cells (PSCs), which significantly improved PSCs stability under constant light and humidity (80%) attack over.¹³ While, the PCE of WS₂-CsPbBr₃ PSCs is only 10.65%.¹³ Recently, inverted perovskite solar cells with WS₂ interlayers had increased PCE up to 21.1%.¹⁴ As we can see, 2D lead halide perovskites applying for PSCs have high stability and can effectively improve PCE by coordinating with two-dimensional transition metal dichalcogenides (TMDs) such as WS₂, MoS₂ and so on.

MoSe₂ and WSe₂ as conventional materials of TMDs are widely used in interface engineering in experiments to improve optoelectronic properties of 2D organic halide perovskites. In experiments, Lu *et al.* fabricated a high-performance WSe₂-CH₃NH₃PbI₃ perovskite photodetector.¹⁵ MoSe₂-CsPbBr₃ Mixed van der Waals nanohybrids shown higher photocurrent than pure CsPbBr₃ nanocrystals.¹⁶ Lee *et al.* found photoluminescence (PL) quenching occurred after the hybridization of perovskites with MoSe₂ and WSe₂ layers, which reflects the charge-transfer effect.¹⁷ The PCE of PEDOT:PSS perovskite solar cells with WSe₂-mediated and without WSe₂-mediated is 16.3% and 13.8%, respectively.¹⁸ These experiments observed the construction of TMDs and perovskite heterostructures can effectively improve the optoelectronic properties of halide perovskites. However, interface electronic transfer and band alignment of Cs₂PbX₄-MSe₂ heterostructures are not studied theoretically. It's worth revealing the effect of halide elements and MSe₂ (M = Mo, W) on Cs₂PbX₄-MSe₂ heterostructures, which can promote the development of 2D inorganic halide perovskite and TMDs optoelectronic applications.

Herein, we constructed the 2D inorganic halide perovskite Cs₂PbX₄ (X = Cl, Br, I) and monolayer MSe₂ heterostructures and discussed their stability, photoelectric properties and charge transfer mechanism through first-principles calculations. We studied electronic structures of the heterostructures with different halide elements and analysed the band alignment type for comparison. Next, we explore the charge transfers mechanism by calculating charge density difference. Finally, optical absorption coefficients of monolayer MSe₂, Cs₂PbX₄ and heterostructures Cs₂PbX₄-MSe₂ were calculated. Our results will be helpful to improve the application performance of two-dimensional lead halide perovskite and TMDs heterostructures.

Computational details

All the density functional theory calculations were performed with the Vienna Ab initial Simulation Package (VASP) code.^{19,20} The projector-augmented wave (PAW) method was referred to electron-ion interactions.²¹ The structure-relaxation, interface binding energy and optoelectronic properties were computed by Perdew, Burke and Ernzerhof's (PBE) exchange correlation

function within the generalized gradient approximation (GGA) formalism.^{22–26} The band gaps of monolayer perovskites and TMDs were further corrected using the screened Heyd-Scuseria-Ernzerhof (HSE) hybrid density functional with the spin-orbital coupling (SOC). The plane wave basis set with a cutoff energy of 450 eV. The convergence criteria were 1×10^{-4} eV for the self-consistent field energy and 0.01 eV Å⁻¹ for the residual forces on each atom, respectively. A vacuum of 20 Å was considered along z direction to avoid artificial interlayer interactions. $3 \times 6 \times 1$ *k*-sampling generated by the Monkhorst-Pack scheme for the Brillouin zone was adopted. The zero damping DFT-D3 method of Grimme is used to account for correcting the van der Waals interaction of Heterostructure.

The interface binding energy is calculated by the following formula:

$$E_b = (E_{\text{heter.}} - E_{\text{Cs}_2\text{PbX}_4} - E_{\text{MSe}_2})/A \quad (1)$$

where *A* represents the interfacial area of Cs₂PbX₄-MSe₂ heterostructures, *E*_{heter.}, *E*_{Cs₂PbX₄}, *E*_{MSe₂} are the total energy of heterostructures Cs₂PbX₄-MSe₂, monolayer Cs₂PbX₄ and MSe₂, respectively.

The plane-averaged charge density difference Δ*ρ* is calculated as the followed equation:

$$\Delta\rho(z) = \rho_{\text{heter.}} - \rho_{\text{Cs}_2\text{PbX}_4} - \rho_{\text{MSe}_2} \quad (2)$$

where *ρ*_{heter.}, *ρ*_{Cs₂PbX₄} and *ρ*_{MSe₂} correspond to the plane-averaged charge density of heterojunctions Cs₂PbX₄-MSe₂, monolayer Cs₂PbX₄ and MSe₂, respectively.

The 2D Mott-Wanier (MW) exciton binding energy (*E*_{eb}) equation is calculated as the followed equation:

$$E_{\text{eb}} = 4 \frac{13.6\mu_{\text{ex}}}{m_0\epsilon^2} \text{eV} \quad (3)$$

where *μ*_{ex} is the effective exciton mass (*μ*_{ex} = *m*_e*m*_h/(*m*_e + *m*_h)), *m*₀ is the electron mass, and ε is the static dielectric constant. The effective masses of electron (*m*_e) and hole (*m*_h) are determined by the curvature of the energy band extremum.

The optical absorption coefficients are obtained from dielectric function, as the followed equation represented:

$$\alpha(\omega) = (\sqrt{2})\omega \left[\sqrt{\epsilon_1(\omega)^2 + \epsilon_2(\omega)^2} - \epsilon_1(\omega) \right]^{1/2} \quad (4)$$

$$\epsilon(\omega) = \epsilon_1(\omega) + i\epsilon_2(\omega) \quad (5)$$

where α, ω correspond to the optical absorption coefficient, the angular frequency and the dielectric function ε(ω) contains real part ε₁(ω) and imaginary part ε₂(ω).

Results and discussion

In recent years, the 2D inorganic perovskites Cs₂PbX₄ (X = Cl, Br, I) and monolayers MSe₂ (M = Mo, W) have been successfully synthesized, which attracted much attention due to its highly stable structures.^{7,27,28} The unit cell of Cs₂PbX₄ perovskites belongs to cubic system and the optimized lattice parameters of



monolayer Cs_2PbCl_4 , Cs_2PbBr_4 , Cs_2PbI_4 is 5.64 Å, 5.91 Å and 6.30 Å, respectively.²⁹ $[\text{CsI}]^0$ interface exhibits more strongly charge transferring than $[\text{PbI}_2]^0$.³⁰ Therefore the $[\text{CsI}]^0$ plane of the monolayer Cs_2PbX_4 is used to form the heterojunctions. The optimized lattice parameters of monolayer MoSe_2 , WSe_2 are 3.32 Å.^{31–33} To minimize the lattice mismatch between the stacking blocks, the supercell of new $\text{Cs}_2\text{PbX}_4\text{--MSe}_2$ heterostructures are built by 3×1 cubic phases Cs_2PbCl_4 and $5 \times \sqrt{3}$ MSe_2 , $\sqrt{8} \times 2$ cubic phases Cs_2PbBr_4 and $5 \times \sqrt{13}$ MSe_2 , and $\sqrt{8} \times \sqrt{2}$ cubic phases Cs_2PbI_4 and $\sqrt{31} \times \sqrt{7}$ MSe_2 , respectively, as seen in Fig. 1 (Table 1).

The lattice mismatches of $\text{Cs}_2\text{PbCl}_4\text{--MoSe}_2$, $\text{Cs}_2\text{PbBr}_4\text{--MoSe}_2$, $\text{Cs}_2\text{PbI}_4\text{--MoSe}_2$, $\text{Cs}_2\text{PbCl}_4\text{--WSe}_2$, $\text{Cs}_2\text{PbBr}_4\text{--WSe}_2$ and $\text{Cs}_2\text{PbI}_4\text{--WSe}_2$ heterostructures are less than 1.10%, 0.62%, 1.79%, 1.02%, 0.65% and 1.82%. The optimized vertical interlayer distances of $\text{Cs}_2\text{PbCl}_4\text{--MoSe}_2$, $\text{Cs}_2\text{PbBr}_4\text{--MoSe}_2$, $\text{Cs}_2\text{PbI}_4\text{--MoSe}_2$, $\text{Cs}_2\text{PbCl}_4\text{--WSe}_2$, $\text{Cs}_2\text{PbBr}_4\text{--WSe}_2$ and $\text{Cs}_2\text{PbI}_4\text{--WSe}_2$ interfaces are 3.07, 3.12, 3.16, 2.90, 3.18 and 3.30 Å, respectively, which increase gradually with the halogen varying from Cl to I. Interface binding energy of $\text{Cs}_2\text{PbCl}_4\text{--MoSe}_2$, $\text{Cs}_2\text{PbBr}_4\text{--MoSe}_2$ and $\text{Cs}_2\text{PbI}_4\text{--MoSe}_2$, $\text{Cs}_2\text{PbCl}_4\text{--WSe}_2$, $\text{Cs}_2\text{PbBr}_4\text{--WSe}_2$ and $\text{Cs}_2\text{PbI}_4\text{--WSe}_2$ interfaces is -16.73 , -16.28 and -16.12 , -17.09 , -17.52 and -15.56 meV Å⁻², respectively. The small interface binding energy and the interlayer distance ranging from 2.9 Å to 3.3 Å indicate that 2D $\text{Cs}_2\text{PbX}_4\text{--MSe}_2$ heterostructures are formed by vdW contact.³³ These E_b values are comparable to E_b of InSe/GaSe (-18.25 meV Å⁻²), suggesting that $\text{Cs}_2\text{PbX}_4\text{--MSe}_2$ heterostructures are stable in energy.³⁴ In addition, $\text{MoSe}_2\text{--CsPbBr}_3$ Mixed van der Waals nanohybrids also have been fabricated in experiments (Table 2).¹⁶

In order to study the band structures of $\text{Cs}_2\text{PbX}_4\text{--MSe}_2$ heterostructures, the band structures of 2D perovskites Cs_2PbX_4 , monolayer WSe_2 and monolayer MoSe_2 were calculated by different functionals, including PBE, PBE with SOC (PBE + SOC), HSE, and HSE with SOC (HSE + SOC) functionals. Monolayer MoSe_2 and monolayer WSe_2 show a direct band gap of 1.45 eV and 1.56 eV using PBE functional. Monolayer

Cs_2PbCl_4 , Cs_2PbBr_4 and Cs_2PbI_4 possess a direct bandgap of 2.59 eV, 2.18 eV and 1.84 eV using PBE functional. We find that the bandgaps of Cs_2PbX_4 computed *via* PBE and HSE + SOC functionals are in good agreements with experiments and other theoretical results.^{9,10,27,28,35–38} The band gap calculated by HSE is larger than PBE and the band gap calculated by PBE + SOC is smaller than PBE. In addition, the band gap of MSe_2 calculated by HSE + SOC is larger than the related experimental data.^{27,28} This is because the HSE functional usually overcorrects slightly the band gap of an intrinsic semiconductor.³⁹ Thus, in order to accurately and closely study the contact characteristics of $\text{Cs}_2\text{PbX}_4\text{--MSe}_2$ heterostructures, PBE is employed in the following heterostructures section. When monolayers Cs_2PbX_4 and MSe_2 ($\text{M} = \text{Mo}, \text{W}$) are contacted to make up $\text{Cs}_2\text{PbX}_4\text{--MSe}_2$ heterostructures, the electronic structures of $\text{Cs}_2\text{PbX}_4\text{--MSe}_2$ heterostructures are shown in Fig. 2. The $\text{Cs}_2\text{PbCl}_4\text{--MoSe}_2$, $\text{Cs}_2\text{PbBr}_4\text{--MoSe}_2$, $\text{Cs}_2\text{PbI}_4\text{--MoSe}_2$, $\text{Cs}_2\text{PbCl}_4\text{--WSe}_2$, $\text{Cs}_2\text{PbBr}_4\text{--WSe}_2$ and $\text{Cs}_2\text{PbI}_4\text{--WSe}_2$ heterostructures have an indirect bandgap with the value of 1.30 eV, 1.48 eV, 1.54 eV, 1.33 eV, 1.67 eV and 1.53 eV, which approach optimal bandgap (1.34 eV) for solar cells.⁴⁰ Moreover, the conduction band minimum (CBM) of $\text{Cs}_2\text{PbX}_4\text{--MSe}_2$ heterostructures are dominated by the MSe_2 layer and the valence band maximum (VBM) of $\text{Cs}_2\text{PbX}_4\text{--MSe}_2$ heterostructures are dominated by Cs_2PbX_4 part.

The analysis of energy level arrangement is of great significance to further study interface properties of $\text{Cs}_2\text{PbX}_4\text{--MSe}_2$ heterostructures. The vacuum energy level (E_v) is set to zero in the precontact state and the Fermi level (E_f) is set to zero in the contact state. In precontact state, the CBM and VBM of monolayer Cs_2PbX_4 are both higher than that of monolayer MSe_2 and both $\text{Cs}_2\text{PbX}_4\text{--MSe}_2$ are type-II heterostructures, as seen in Fig. 3(a). The electrons will diffuse from monolayers Cs_2PbX_4 to monolayers MSe_2 and the holes will move from monolayers MSe_2 monolayer to monolayers Cs_2PbX_4 when they contact. Correspondingly, the holes accumulate in Cs_2PbX_4 monolayers and the electrons accumulate in MSe_2 monolayers. Moreover, the difference between the vacuum level and the Fermi level is defined as the work function.³¹ The work function is the internal dynamics of electron flow. The computed work functions of monolayers Cs_2PbCl_4 , Cs_2PbBr_4 , Cs_2PbI_4 , MoSe_2 and WSe_2 are 4.08 eV, 4.21 eV, 4.30 eV, 4.83 eV and 4.51 eV, respectively. The heterostructures are mainly related to the work function after contact. In order to maintain Fermi levels at the same level after contact, the Fermi level of all Cs_2PbX_4 perovskites moved down and the Fermi level of monolayer MSe_2 moved up after they contact each other.

The energy level diagram of $\text{Cs}_2\text{PbX}_4\text{--MSe}_2$ heterostructures in contact states is given in Fig. 3(b). The $\text{Cs}_2\text{PbX}_4\text{--MSe}_2$ heterostructures are type-II heterostructures, which can drive the photogenerated holes and electrons to move in opposite directions, resulting in spatial separation of holes and electrons on different sides of heterostructures.³³ Thus, $\text{Cs}_2\text{PbX}_4\text{--MSe}_2$ heterostructures are beneficial for improving the photoelectric conversion efficiency of $\text{Cs}_2\text{PbX}_4\text{--MSe}_2$ optoelectronic applications. In these type-II heterostructures, the differences between the VBM of their components (valence band offset, Δv) are crucial for hole blocking and the differences between the CBM

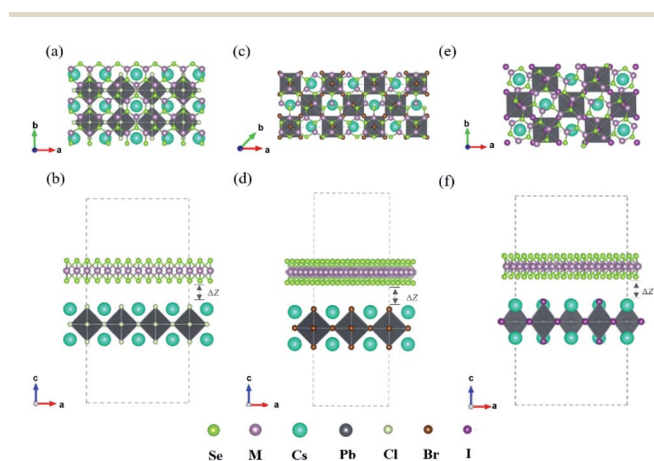


Fig. 1 Top and side views of relaxed $\text{Cs}_2\text{PbX}_4\text{--MSe}_2$ ($\text{M} = \text{Mo}, \text{W}$) heterostructures. (a and b) $\text{Cs}_2\text{PbCl}_4\text{--MSe}_2$ heterostructure. (c and d) $\text{Cs}_2\text{PbBr}_4\text{--MSe}_2$ heterostructure. (e and f) $\text{Cs}_2\text{PbI}_4\text{--MSe}_2$ heterostructure.

Table 1 Optimized lattice parameters (a and b) and expanded cell lattice parameters (a_1 and b_1), lattice mismatch, interlayer distance ΔZ and interface binding energy E_b of the relaxed $\text{Cs}_2\text{PbX}_4\text{-WSe}_2$ heterostructures

Heterostructures	a (Å)	b (Å)	a_1 (Å)	b_1 (Å)	Mismatch (%)	ΔZ (Å)	E_b (meV Å ⁻²)
$\text{Cs}_2\text{PbCl}_4\text{-MoSe}_2$	16.51	5.66	16.75	5.69	1.10%	3.09	-16.73
$\text{Cs}_2\text{PbBr}_4\text{-MoSe}_2$	16.35	11.81	16.65	11.89	0.62%	3.12	-16.28
$\text{Cs}_2\text{PbBr}_4\text{-MoSe}_2$	18.22	8.66	18.14	8.84	1.79%	3.16	-16.12
$\text{Cs}_2\text{PbBr}_4\text{-WSe}_2$	16.41	5.67	16.75	5.69	1.02%	2.90	-17.09
$\text{Cs}_2\text{PbBr}_4\text{-WSe}_2$	16.35	11.81	16.65	11.89	0.65%	3.18	-17.52
$\text{Cs}_2\text{PbBr}_4\text{-WSe}_2$	18.21	8.66	18.20	8.67	1.82%	3.30	-15.56

Table 2 The bandgaps of monolayer Cs_2PbX_4 , MoSe_2 and WSe_2 by different calculation method

Functional	Cs_2PbCl_4	Cs_2PbBr_4	Cs_2PbI_4	MoSe_2	WSe_2
PBE	2.59	2.18	1.84	1.45	1.56
HSE	3.58	2.91	2.57	2.10	2.03
PBE + SOC	1.84	1.46	1.09	1.37	1.33
HSE + SOC	2.78	2.14	1.78	1.89	1.93
Experiment	3.01	2.32	1.86	1.48	1.60

of their components (conduction band offset, Δc) are crucial for electron transport. The large Δv promotes hole extraction from TMDs layers to perovskite layers and the large Δc allows free electrons to move from perovskite to TMDs layers. It is shown that $\text{Cs}_2\text{PbCl}_4\text{-MSe}_2$ heterostructure have the largest Δc and Δv among $\text{Cs}_2\text{PbX}_4\text{-MSe}_2$ heterostructures.¹² Thus, the $\text{Cs}_2\text{PbCl}_4\text{-MoSe}_2$, and $\text{Cs}_2\text{PbCl}_4\text{-WSe}_2$ heterostructures may have the largest charge transport power and are more conducive to reducing the dark current. In addition, $\text{Cs}_2\text{PbX}_4\text{-MoSe}_2$

heterostructures have larger Δc and Δv than $\text{Cs}_2\text{PbX}_4\text{-WSe}_2$ heterostructures.

In addition, the Mott-Wanier theory has been used to approximate exciton binding energies in the vdW heterostructures.⁴¹ Table 3 lists the carrier masses and MW exciton binding energy of $\text{Cs}_2\text{PbX}_4\text{-MSe}_2$ heterostructures. The lower exciton binding energies usually facilitate the splitting of excitons into free charge carriers.³⁴ It is shown that $\text{Cs}_2\text{PbCl}_4\text{-MoSe}_2$ heterostructures exhibit the lowest exciton binding energies among $\text{Cs}_2\text{PbX}_4\text{-MoSe}_2$ heterostructures. Therefore, $\text{Cs}_2\text{PbCl}_4\text{-MoSe}_2$ heterostructures can effectively promotes the separation of excitons and exhibit the highest charge transport efficiency among $\text{Cs}_2\text{PbX}_4\text{-MSe}_2$ heterostructures.

To clearly investigate the recombination rates of electron-hole pairs and transfer of charges between the Cs_2PbX_4 monolayers and MSe_2 monolayers across the interfaces, the plane-averaged charge density difference $\Delta\rho$ are calculated, as shown in Fig. 4. The results demonstrate that the holes mainly accumulated at the Cs_2PbX_4 interfaces and the charges accumulated at MSe_2 interfaces. Additionally, the charges transfer

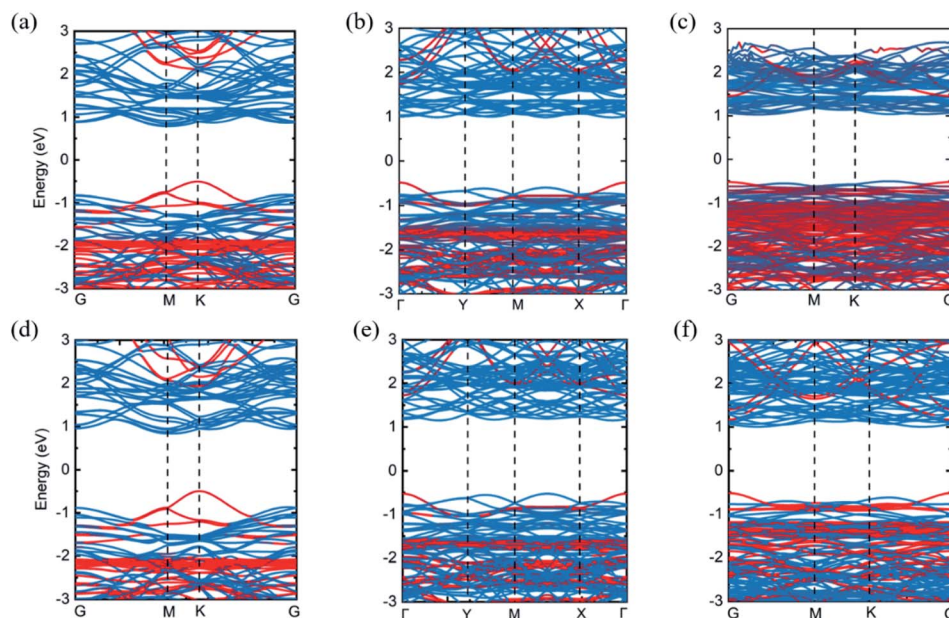


Fig. 2 Band structures of $\text{Cs}_2\text{PbX}_4\text{-MSe}_2$ heterostructures (a) $X = \text{Cl}$, $M = \text{Mo}$; (b) $X = \text{Br}$, $M = \text{Mo}$; (c) $X = \text{I}$, $M = \text{Mo}$; (d) $X = \text{Cl}$, $M = \text{W}$; (e) $X = \text{Br}$, $M = \text{W}$; (f) $X = \text{I}$, $M = \text{W}$. The red and blue lines correspond to Cs_2PbX_4 and MSe_2 , respectively.



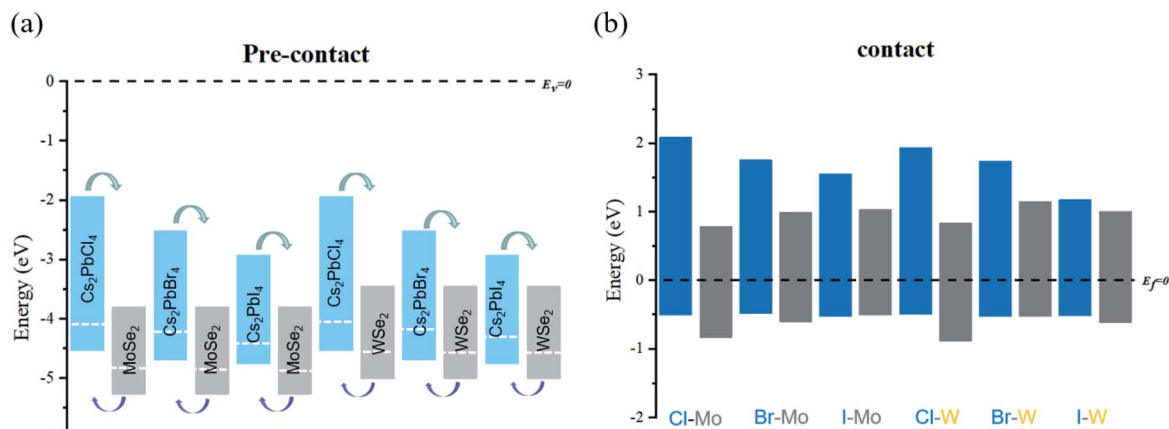


Fig. 3 Energy level graphs of the monolayer MSe₂ and Cs₂PbX₄ in the precontact (a) and contact (b). Blue and gray rectangles represent the monolayer Cs₂PbX₄ and MSe₂. The bottom and top of rectangles correspond to VBM and CBM, respectively.

Table 3 Carrier effective masses (m_e , m_h , and μ_{ex}), static dielectric constant (ϵ) and MW excitonic binding energies (E_{eb})

Heterostructures	m_e (m_0)	m_h (m_0)	μ_{ex} (m_0)	ϵ	E_{eb} (eV)
Cs ₂ PbCl ₄ –MoSe ₂	1.16	0.28	0.23	5.34	0.43
Cs ₂ PbBr ₄ –MoSe ₂	0.79	0.71	0.37	5.44	0.68
Cs ₂ PbI ₄ –MoSe ₂	1.39	0.67	0.45	5.24	1.00
Cs ₂ PbCl ₄ –WSe ₂	1.46	0.27	0.23	4.95	0.50
Cs ₂ PbBr ₄ –WSe ₂	0.75	0.66	0.35	5.08	0.74
Cs ₂ PbI ₄ –WSe ₂	1.03	0.61	0.38	5.26	0.76

from Cs₂PbX₄ interfaces to MSe₂ interfaces while the holes move in the opposite direction. The direction of charge transfer

is consistent with that of band alignment analysis, which shown type-II heterostructures are benefited to the separation of electrons and holes. What's more, detail charge transfers among interlayer spacing are used to quantitatively evaluate charge transferring of heterostructures. The Cs₂PbCl₄–MoSe₂, Cs₂PbBr₄–MoSe₂, Cs₂PbI₄–MoSe₂, Cs₂PbCl₄–WSe₂, Cs₂PbBr₄–WSe₂ and Cs₂PbI₄–WSe₂ heterostructures have charge transfers among interlayer spacing with the values of 0.18, 0.15, 0.16, 0.16, 0.10 and 0.14 $\times 10^{-3}e$. We find that the Cs₂PbCl₄–MSe₂ heterostructures have the highest charge transfers in the six Cs₂PbX₄–MSe₂ heterostructures among interlayer spacing. In addition, Cs₂PbX₄–MoSe₂ heterostructures have larger charge transfers than Cs₂PbX₄–WSe₂ heterostructures. It is because

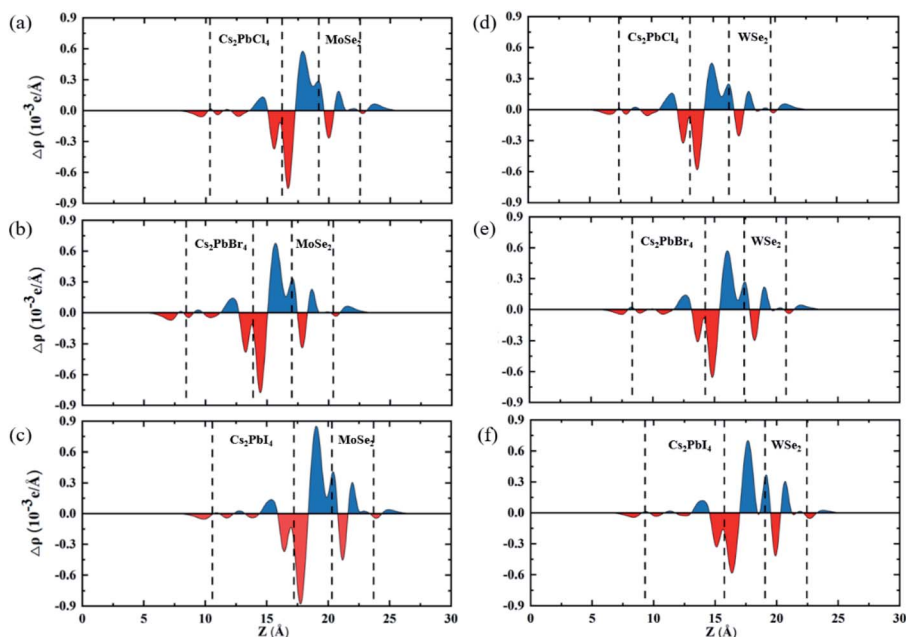


Fig. 4 The planar-averaged differential charge density $\Delta\rho(z)$ of Cs₂PbX₄–MSe₂ heterostructures: (a and d) X = Cl; (b and e) X = Br; (c and f) X = I. Red and blue represent electron depletion and accumulation.



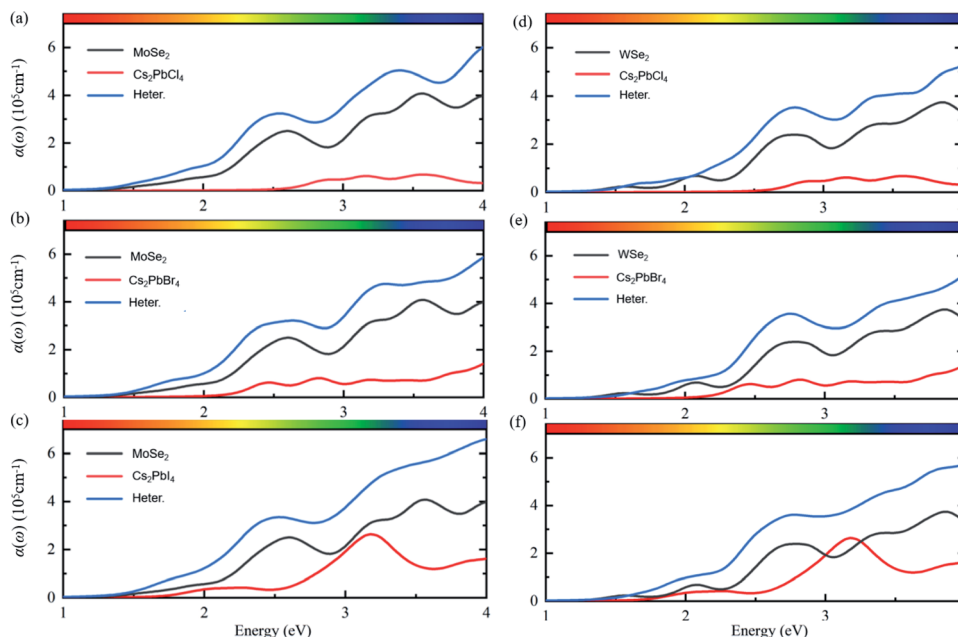


Fig. 5 Optical absorption coefficients of (a and d) $\text{Cs}_2\text{PbCl}_4\text{-MSe}_2$, (b and e) $\text{Cs}_2\text{PbBr}_4\text{-MSe}_2$, and (c and f) $\text{Cs}_2\text{PbI}_4\text{-MSe}_2$ heterostructures. The blue, black and red lines represent optical absorption spectrum of heterostructures, monolayer MSe_2 and Cs_2PbX_4 , respectively.

$\text{Cs}_2\text{PbX}_4\text{-MoSe}_2$ heterostructures have smaller exciton binding energies than $\text{Cs}_2\text{PbX}_4\text{-WSe}_2$ heterostructures.

Except for the electronic structures and charge redistributions of heterostructures, the optical properties in vdW heterostructures have important effects on the performance of perovskite optoelectronic devices.⁴² In order to further study the optical properties of $\text{Cs}_2\text{PbX}_4\text{-MSe}_2$ heterostructures, the optical absorption coefficients were calculated, as shown in Fig. 5. The optical absorption coefficients of MSe_2 monolayers are higher than those of Cs_2PbX_4 monolayers. The optical absorption coefficients of all heterostructures are significantly higher than those of corresponding Cs_2PbX_4 monolayers and

MSe_2 monolayers. This is because the bandgaps of 2D $\text{Cs}_2\text{PbX}_4\text{-MSe}_2$ heterostructures are smaller than the bandgaps of Cs_2PbX_4 monolayers. The bandgaps of $\text{Cs}_2\text{PbX}_4\text{-MSe}_2$ heterostructures approach optimal bandgap (1.34 eV) for solar cells.⁴⁰ Thus, the construction of $\text{Cs}_2\text{PbX}_4\text{-MSe}_2$ heterostructures is beneficial to improve the light absorption of optoelectronic devices.

In addition, the light absorption of $\text{Cs}_2\text{PbX}_4\text{-MoSe}_2$ heterostructures is only slightly greater than that of $\text{Cs}_2\text{PbX}_4\text{-WSe}_2$ in the visible region because the light absorption of MoSe_2 is slightly greater than that of WSe_2 in the visible region. The light absorption coefficients of Cs_2PbX_4 increase rapidly in the visible

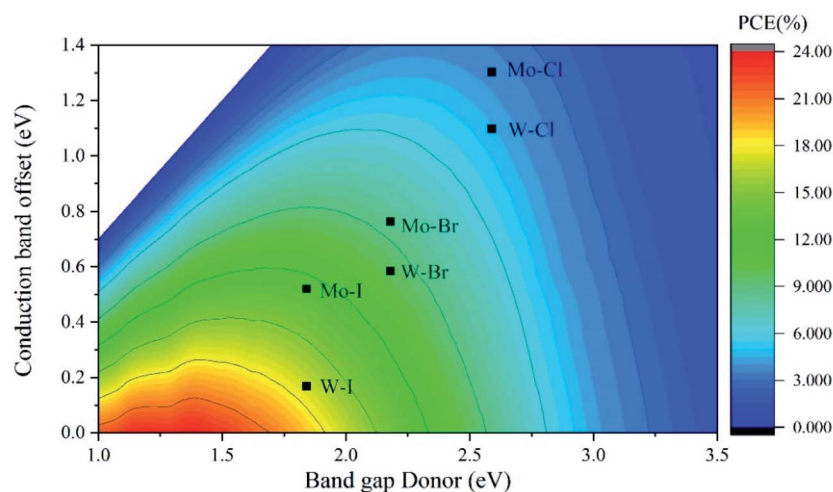


Fig. 6 Contour plot showing the calculated PCE of $\text{Cs}_2\text{PbX}_4\text{-MSe}_2$ heterostructures according to the Cs_2PbX_4 donor band gap and conduction band offset Δ_c .



region with the halogen varying from Cl to I. The light absorption coefficients of $\text{Cs}_2\text{PbX}_4\text{-MSe}_2$ heterostructures increase rapidly in the visible region with the halogen varying from Cl to I. Therefore, the $\text{Cs}_2\text{PbI}_4\text{-MSe}_2$ heterostructure has great potential application in solar cells among $\text{Cs}_2\text{PbX}_4\text{-MSe}_2$ heterostructures.

In order to explore the application of $\text{Cs}_2\text{PbX}_4\text{-MSe}_2$ heterostructures in solar cells, we further calculated the power conversion efficiency of $\text{Cs}_2\text{PbX}_4\text{-MSe}_2$ heterostructures. The PCE depends on the donor band gap and conduction band offset.^{43,44} The PCE of $\text{Cs}_2\text{PbX}_4\text{-MSe}_2$ heterostructures with type-II alignment is shown in Fig. 6 as a contour plot where the x - and y -axes are the donor band gap and conduction band offset, respectively. The PCE of $\text{Cs}_2\text{PbI}_4\text{-MoSe}_2$, $\text{Cs}_2\text{PbI}_4\text{-WSe}_2$ and $\text{Cs}_2\text{PbBr}_4\text{-WSe}_2$ heterostructures is 13%, 18% and 10%, which is greater than 10%. Three heterostructures can be used in solar cells. $\text{Cs}_2\text{PbI}_4\text{-WSe}_2$ heterostructure has the largest PCE (18%) among $\text{Cs}_2\text{PbX}_4\text{-MSe}_2$ heterostructures. The PCE of $\text{Cs}_2\text{PbI}_4\text{-WSe}_2$ heterostructure is larger than that of $\text{WS}_2/\text{CsPbBr}_3$ PSCs (10.65%) and WSe_2 -mediated PSCs (16.3%).^{13,18} With the halogen varying from Cl to I, the PCE of $\text{Cs}_2\text{PbX}_4\text{-WSe}_2$ heterostructures increase rapidly. The PCE variation trend is consistent with the absorption coefficients of $\text{Cs}_2\text{PbX}_4\text{-MSe}_2$ heterostructures varying from Cl to I. In addition, the PCE of $\text{Cs}_2\text{PbX}_4\text{-WSe}_2$ heterostructures is larger than the PCE of $\text{Cs}_2\text{PbX}_4\text{-MoSe}_2$ heterostructures. This is mainly because $\text{Cs}_2\text{PbX}_4\text{-WSe}_2$ heterostructures have smaller Δc than $\text{Cs}_2\text{PbX}_4\text{-MoSe}_2$ heterostructures. Therefore, $\text{Cs}_2\text{PbI}_4\text{-WSe}_2$ heterostructure has great potential application in solar cells.

Conclusions

In summary, we have systematically studied the structural, stability, charge transfer and optoelectronic properties of 2D perovskite $\text{Cs}_2\text{PbX}_4\text{-MSe}_2$ heterostructures with different halide elements based on density functional calculations. All these six heterostructures are stable in energy and the interlayer distances increase gradually with the halogen varying from Cl to I. Electronic structures show $\text{Cs}_2\text{PbX}_4\text{-MSe}_2$ heterostructures belongs to type-II energy level shifts with narrower optical gap than monolayers Cs_2PbX_4 . In addition, the $\text{Cs}_2\text{PbCl}_4\text{-MoSe}_2$, and $\text{Cs}_2\text{PbCl}_4\text{-WSe}_2$ heterostructures may have the largest charge transport power due to their larger band offset, which is beneficial to reduce dark current and improve open circuit voltage. $\text{Cs}_2\text{PbCl}_4\text{-MoSe}_2$ heterostructures can effectively promotes the separation of excitons and exhibit the highest charge transport efficiency among $\text{Cs}_2\text{PbX}_4\text{-MSe}_2$ heterostructures due to their lowest exciton binding energies. $\text{Cs}_2\text{-PbCl}_4\text{-MoSe}_2$ heterostructure has the potential to be applied in photodetectors. Next, the charges transfer from Cs_2PbX_4 interfaces to MSe_2 interfaces while the holes move in the opposite direction. The holes mainly accumulated at the Cs_2PbX_4 interfaces and the charges accumulated at MSe_2 interfaces. $\text{Cs}_2\text{PbX}_4\text{-MoSe}_2$ heterostructures have larger charge transfers than $\text{Cs}_2\text{PbX}_4\text{-WSe}_2$ heterostructures. Finally, optical absorptions of six heterostructures are significantly higher than the corresponding Cs_2PbX_4 monolayers and MSe_2 monolayers. The PCE

of $\text{Cs}_2\text{PbX}_4\text{-WSe}_2$ heterostructures is larger than the PCE of $\text{Cs}_2\text{PbX}_4\text{-MoSe}_2$ heterostructures. $\text{Cs}_2\text{PbI}_4\text{-WSe}_2$ heterostructure has the largest PCE (18%) among $\text{Cs}_2\text{PbX}_4\text{-MSe}_2$ heterostructures. In conclusion, $\text{Cs}_2\text{PbCl}_4\text{-MoSe}_2$ heterostructure exhibits great potential application in photodetectors devices and $\text{Cs}_2\text{PbI}_4\text{-WSe}_2$ heterostructure has great potential application in solar cells. Our findings provide insight into of $\text{Cs}_2\text{PbX}_4\text{-MSe}_2$ heterostructures can effectively improve the performance of perovskite optoelectronic devices.

Conflicts of interest

There are no conflicts to declare.

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Notes and references

- 1 A. T. Barrows, A. J. Pearson, C. K. Kwak, A. D. F. Dunbar, A. R. Buckley and D. G. Lidzey, *Energy Environ. Sci.*, 2014, 7, 2944–2950.
- 2 K. Xiao, R. Lin, Q. Han, Y. Hou, Z. Qin, H. T. Nguyen, J. Wen, M. Wei, V. Yeddu, M. I. Saidaminov, Y. Gao, X. Luo, Y. Wang, H. Gao, C. Zhang, J. Xu, J. Zhu, E. H. Sargent and H. Tan, *Nat. Energy*, 2020, 5, 870–880.
- 3 I. L. Braly, D. W. deQillettes, L. M. Pazos-Outon, S. Burke, M. E. Ziffer, D. S. Ginger and H. W. Hillhouse, *Nat. Photonics*, 2018, 12, 355–361.
- 4 M. M. Lee, J. Teuscher, T. Miyasaka, T. N. Murakami and H. J. Snaith, *Science*, 2012, 338, 643–647.
- 5 J. H. Noh, S. H. Im, J. H. Heo, T. N. Mandal and S. I. Seok, *Nano Lett.*, 2013, 13, 1764–1769.
- 6 Y.-Y. Zhang, S. Chen, P. Xu, H. Xiang, X.-G. Gong, A. Walsh and S.-H. Wei, *Chin. Phys. Lett.*, 2018, 35, 036104.
- 7 S. Chen and G. Shi, *Adv. Mater.*, 2017, 29, 1605448.
- 8 Y. F. Ding, Q. Q. Zhao, Z. L. Yu, Y. Q. Zhao, B. Liu, P. B. He, H. Zhou, K. L. Li, S. F. Yin and M. Q. Cai, *J. Mater. Chem. C*, 2019, 7, 7433–7441.
- 9 A. Bala, A. K. Deb and V. Kumar, *J. Phys. Chem. C*, 2018, 122, 7464–7473.
- 10 J. Song, L. Xu, J. Li, J. Xue, Y. Dong, X. Li and H. Zeng, *Adv. Mater.*, 2016, 28, 4861–4869.
- 11 E. Shi, Y. Gao, B. P. Finkenauer, A. Akriti, A. H. Coffey and L. Dou, *Chem. Soc. Rev.*, 2018, 47, 6046–6072.
- 12 J. He, J. Su, Z. Lin, S. Zhang, Y. Qin, J. Zhang, J. Chang and Y. Hao, *J. Phys. Chem. C*, 2019, 123, 7158–7165.



- 13 Q. W. Zhou, J. L. Duan, X. Y. Yang, Y. Y. Duan and Q. W. Tang, *Angew. Chem., Int. Ed.*, 2020, **132**, 22181–22185.
- 14 J. P. Cao, G. Q. Tang, P. You, T. Y. Wang, F. Y. Zheng, J. Zhao and F. Yan, *Adv. Funct. Mater.*, 2020, **30**, 2002358.
- 15 Hybrid Bilayer WSe₂-CH₃NH₃PbI₃ Organolead Halide Perovskite as a High-Performance Photodetector.
- 16 S. Hassan, P. Basera, S. Bera, M. Mittal, S. K. Ray, S. Bhattacharya and S. Sapra, *ACS Appl. Mater. Interfaces*, 2020, **12**, 7217–7325.
- 17 S.-H. Lee, J. Y. Kim, S. Choi, Y. Lee, K.-S. Lee, J. Kim and J. Joo, *ACS Appl. Mater. Interfaces*, 2020, **12**, 25159–25167.
- 18 Y. Choi, S. Jung, N. K. Oh, J. Lee, J. Seo, U. Kim, D. Koo and H. Park, Enhanced charge transport via metallic 1T phase transition metal dichalcogenides-mediated hole transport layer engineering for perovskite solar cells, *Chemnanomat*, 2019, **5**, 1050–1058.
- 19 G. Kresse and D. Joubert, *Phys. Rev. B: Condens. Matter Mater. Phys.*, 1999, **59**, 1758–1775.
- 20 J. P. Perdew, K. Burke and M. Ernzerhof, *Phys. Rev. Lett.*, 1996, **77**, 3865–3868.
- 21 P. E. Blochl, *Phys. Rev. B: Condens. Matter Mater. Phys.*, 1994, **50**, 17953–17979.
- 22 B. Jia, P. Zhu, S. Sun, L. Han, G. Liu, Y. Wang, G.-D. Peng and P. Lu, *IEEE J. Sel. Top. Quantum Electron.*, 2019, **26**, 1–6.
- 23 S. Lin, J. C. Liu, W. Z. Li, D. Wang, Y. Huang, C. Jia, Z. W. Li, H. Y. Wang, J. N. Song, Z. L. Liu, K. Huang, D. Zu, M. Lei, B. Hong and H. Wu, *Nano Lett.*, 2019, **19**, 6853–6861.
- 24 J. C. Xu, K. Bi, R. Zhang, Y. N. Hao, C. W. Lan, M.-M. Klaus, X. J. Zhai, Z. D. Zhang and S. G. H, *Research*, 2019, **2019**, 9686213.
- 25 X. N. Guan, R. Zhang, B. N. Jia, L. Y. Wu, B. Zhou, L. Fan, G. Liu, Y. Wang, P. F. Lu and G. D. Peng, *J. Non-Cryst. Solids*, 2020, **550**, 120388.
- 26 P. F. Lu, *J. Sichuan Norm. Univ., Nat. Sci.*, 2020, **43**, 1–20.
- 27 X. Wang, Y. Gong, G. Shi, W. L. Chow, K. Keyshar, G. Ye, R. Vajtai, J. Lou, Z. Liu, E. Ringe, B. K. Tay and P. M. Ajayan, *ACS Nano*, 2014, **8**, 5125–5131.
- 28 J.-K. Huang, J. Pu, C.-L. Hsu, M.-H. Chiu, Z.-Y. Juang, Y.-H. Chang, W.-H. Chang, Y. Iwasa, T. Takenobu and L.-J. Li, *ACS Nano*, 2014, **8**(1), 923–930.
- 29 J.-H. Yang, Q. Yuan and B. I. Yakobson, *J. Phys. Chem. C*, 2016, **120**, 24682–24687.
- 30 Y.-Q. Zhao, Y. Xu, D.-F. Zou, J.-N. Wang, G.-F. Xie, B. Liu, M.-Q. Cai and S.-L. Jiang, *J. Phys.: Condens. Matter*, 2020, **32**, 195501.
- 31 Y. Ding, Y. Wang, J. Ni, L. Shi, S. Shi and W. Tang, *Phys. Chem. Chem. Phys.*, 2011, **13**, 15546–15553.
- 32 Y. Zhang, Y. Zhao, Y. Xu and L. He, *Solid State Commun.*, 2021, **327**, 114233.
- 33 B. Liu, M. Long, M.-Q. Cai and J. Yang, *J. Phys. Chem. Lett.*, 2018, **327**, 114233.
- 34 J. H. Chen, X. J. He, B. S. Sa, J. Zhou, C. Xu, C. L. Wen and Z. M. Sun, *Nanoscale*, 2019, **11**, 6431–6444.
- 35 P. Zhu, H. Zhao, X. Li, Y. Zu, L. Wu, C. Chen and G. Liu, *Phys. E*, 2021, **134**, 114908.
- 36 A. Molina-Sanchez, *ACS Appl. Energy Mater.*, 2018, **1**, 6361–6367.
- 37 Z. Zheng, X. X. Wang, Y. W. Shen, Z. Y. Luo, L. G. Li, L. Gan, Y. Ma, H. Q. Li, A. L. Pan and T. Y. Zhai, *Adv. Opt. Mater.*, 2018, **6**, 1800879.
- 38 P. B. Gui, H. Zhou, F. Yao, Z. H. Song, B. R. Li and G. J. Fang, *Small*, 2019, **15**, 1902618.
- 39 R. J. Sutton, M. R. Filip, A. A. Haghighirad, N. Sakai, B. Wenger, F. Giustino and H. J. Snaith, *ACS Energy Lett.*, 2018, **3**, 1787–1794.
- 40 S. Ruhle, *Sol. Energy*, 2016, **130**, 139–147.
- 41 P. Wang, Y. X. Zong, H. Liu, H. Y. Wen, H. X. Deng, Z. M. Wei, H. B. Wu and J. B. Xia, *J. Phys. Chem. C*, 2020, **214**, 23832–23838.
- 42 W. J. Yin, J. H. Yang, J. Kang, Y. Yan and S. H. Wei, *J. Mater. Chem. A*, 2015, **3**, 8926–8942.
- 43 M. C. Scharber, D. Wuhlbacher, M. Koppe, P. Denk, C. Waldauf, A. J. Heeger and C. L. Brabec, *Adv. Mater.*, 2006, **18**, 189.
- 44 M. Bernardi, M. Palummo and J. C. Grossman, *ACS Nano*, 2012, **6**, 10082–10089.

