Recent progress on metal halide perovskite field-effect transistors

Huihui Zhu, Ao Liu and Yong-Young Noh

Department of Chemical Engineering, Pohang University of Science and Technology (POSTECH), Pohang, Republic of Korea

ABSTRACT
Metal halide perovskite semiconductors could potentially be used to create field-effect transistors (FETs) with high carrier mobilities. This review summarizes progress achieved recently in three-dimensional (3D) lead-based and two-dimensional (2D) tin-based perovskite FETs, and identifies the evolution of electrical characteristics and stability, then discusses outstanding challenges and provides an outlook on the possibilities offered by this electronic material family for use in backplane drivers for active matrix displays.

1. Introduction
Metal halide perovskite semiconductors have attracted great attention over the past decade for optoelectronic applications, e.g. solar cells and light-emitting diodes [1]. Despite their high intrinsic carrier mobilities ($\mu$), their use in field-effect transistors (FETs) has not yet been well explored [2–4]. First-principles theoretical calculations and experimental studies of charge transport, e.g. the Hall effect, time-resolved spectroscopy, and space charge limited current and time of flight measurements, have demonstrated that metal halide perovskites can exhibit high bulk carrier mobilities in the range of $10 \sim 100$ cm$^2$ V$^{-1}$ s$^{-1}$ [5–10]. Moreover, FETs provide a good test bed for studies of carrier transport and structure–property relationships in metal halide perovskite semiconductors [11].

The first perovskite FET used a two-dimensional (2D) layered perovskite, phenylethylammonium tin iodide [(PEA)$_2$SnI$_4$], as the channel layer [12]. The (PEA)$_2$SnI$_4$ FETs showed typical p-channel characteristics with $\mu \approx 0.6$ cm$^2$ V$^{-1}$ s$^{-1}$. Since 2013, a three-dimensional (3D) metal halide perovskite semiconductor, methyl ammonium lead iodide (MAPbI$_3$), has been investigated for use in FETs [13]. Initially, serious ionic migration in 3D perovskite FETs obscured the charge-transport physics of perovskite materials. FET gate modulation could be detected only by suppressing ion movement with low-temperature operation [14,15]. As the perovskite film and devices were improved, the concentration of migrating ionic defects decreased, so good current modulation could be obtained at room temperature (RT) [11,16].

Compared to 3D perovskite analogues, 2D tin-based (PEA)$_2$SnI$_4$ FETs have the advantage of intrinsically suppressing ion migration in the (PEA)$_2$SnI$_4$ channel layer due to their unique layered structure, besides which their organic cation is too large to move easily. In the last five years, several research groups have sought to better understand (PEA)$_2$SnI$_4$ FETs and have tried to increase their electrical properties [17–22]. Recently, organic molecular engineering has been used to increase the chemical stability of 2D tin-based perovskite FETs in air [23].

In this review, we briefly introduce the fundamentals of metal halide perovskite semiconductors, including their structure, and the typical energy bands of lead and tin perovskites. We clarify why lead-based perovskite FETs mostly show ambipolar characteristics, whereas tin-based perovskites exhibit stronger p-channel behavior. Then we review recent progress in the development of 3D lead-based and 2D tin-based perovskite FETs, with emphasis on the evolution of their carrier mobility and stability, and remaining challenges. Finally, we provide an outlook on the opportunities that this family of electronic materials offers.

2. Fundamentals
The typical crystal structure of 3D metal halide perovskite semiconductors can be described with the...
structural formula $\text{ABX}_3$ (Figure 1(a)). $A$ is a monovalent cation, as, for example, organic methylammonium ($\text{MA}^+$), formamidinium ($\text{FA}^+$), inorganic cesium ($\text{Cs}^+$), or rubidium ($\text{Rb}^+$). $B$ is a divalent metal cation, as, for example, $\text{Pb}^{2+}$ or $\text{Sn}^{2+}$. $X$ is a halide anion, generally $\text{I}^-$, $\text{Br}^-$, or $\text{Cl}^-$. Recently, multiple cation and anion systems have also been investigated.

Other derivatives, for example, layered perovskites, can be obtained by inserting large spacer cations to cut the 3D perovskite along crystallographic planes (Figure 1(b)). The spacer can be a monovalent ($A'$) aromatic/aliphatic alkyl ammonium group such as butylammonium ($\text{BA}^+$) or phenethylammonium ($\text{PEA}^+$), or a divalent ($A''$) aromatic/aliphatic alkyl ammonium group such as 1,3-propane-diamine ($\text{PDA}^{2+}$), 3-(aminomethyl)piperidinium ($\text{3AMP}^{2+}$), or 1,4-phenylenedimethanammonium ($\text{PDMA}^{2+}$).

Layered perovskites are divided into $A_2^+A_{n-1}^-B_nX_{3n+1}^{-1}$ RP-structured perovskites and $A'/A_{n-1}^-B_nX_{3n+1}^{-1}$ DJ-structured perovskites (Figure 1(c)) [24]. Both are referred to as 2D perovskites. Due to the van der Waals force in the layered structures, 2D perovskites have higher formation energy than their 3D counterparts. Therefore, 2D structures should be more stable than 3D structures [26]. Organic cation engineering may increase the ambient stability of 2D perovskites, but the insertion in them of an insulating organic spacer could influence their charge carrier transport, particularly along the orthometric direction.

At this stage, the B-site metal cation is typically lead or tin. The elements have similar ionic radii (Pb: 1.19 Å and Sn: 1.18 Å); similar optoelectronic properties, that is, direct band gaps in the visible spectrum; and relatively low effective masses of electrons and holes [27]. Several studies that screened most possible hypothetical halide perovskites revealed that halide perovskite materials based on lead or tin have truly unique optoelectronic properties [28,29]. However, lead-based and tin-based perovskites have slightly different electronic band structures. Lead shows a strong lanthanide contraction effect, but the electrons in Sn 5s are less attracted by the nuclear charge than the electrons in Pb 6s, and thus, the energy level of Sn 5s is shallower than that of Pb 6s (Sn 5s: $-10.1$ eV and Pb 6s: $-11.6$ eV) [30].
As a result, in MAPbI\textsubscript{3}, the valence band maximum (VBM) is dominated by I 5p orbitals with some contribution from the Pb 6s orbitals, whereas the conduction band minimum (CBM) is strongly dominated by the Pb 6p state with slight mixing from the I 5p state. The electronic signature of the organic cation is found deep in the VB; this status implies that the organic part merely stabilizes the perovskite structure and balances the electrostatic charges, but does not directly contribute to the electronic properties. In contrast, in SnI\textsubscript{2}-based perovskites, the VBM is composed of hybrid Sn 5s and I 5p orbitals, and the CBM is dominated by the Sn 5p state with a slight contribution from the I 5p state (Figure 1(d)).

These differences in traits yield different advantages for the two types of perovskite. PbI\textsubscript{2}-based perovskites will show better electron transport than SnI\textsubscript{2}-based perovskites, because the large size and mass of their lead introduces strong spin–orbit coupling (SOC), which makes Pb 6p orbitals deeper and more dispersive than Sn 5p orbitals. In contrast, SnI\textsubscript{2}-based perovskites should show a stronger hole carrier transport property than PbI\textsubscript{2}-based perovskites due to the hybrid Sn 5s and I 5p states in VBM and the low formation energy of the tin vacancies [31,32]. These comparisons of electronic band structures suggest that lead-based perovskites are suitable for ambipolar FETs, and tin-based perovskites are suitable for p-channel FETs (Figure 1(e)). Unipolar \(n/p\)-combinations are generally more preferable for large-area complementary circuits than ambipolar FETs due to their very low off-state current, which means low power dissipation.

3. Lead-based perovskite FETs

In the first bottom-gate, top-contact FETs that used MAPbI\textsubscript{3} films as the channel layer in 2013, the films behaved like ambipolar semiconductors with a hole mobility \(\mu_h = 10^{-5} \text{cm}^2 \text{V}^{-1} \text{s}^{-1}\) at RT. The low \(\mu_h\) was assumed to have resulted from the gate-field screening effects induced by ion migration and accumulation in the perovskite FETs [13]. Two years later, thin-film MAPbI\textsubscript{3} FETs showed much higher \(\mu_h = 2.1 \times 10^{-2} \text{cm}^2 \text{V}^{-1} \text{s}^{-1}\) and electron mobilities \(\mu_e = 7.2 \times 10^{-2} \text{cm}^2 \text{V}^{-1} \text{s}^{-1}\) when operated at 78 K. These results suggest that screening effects associated with ionic transport can be effectively reduced by decreasing \(T_{\text{OP}}\) [33].

Temperature-dependent electrical behavior was also observed in MAPbI\textsubscript{3} FETs with SiO\textsubscript{2} as the dielectric layer in the bottom-gate, top-contact device configuration (Figure 2(a)) [14]. A similar behavior was seen in MAPbBr\textsubscript{3} FETs with different device structures and dielectric layers (Figure 2(b)) [15]. A dramatic decrease in \(\mu\) was seen as \(T_{\text{OP}}\) increased, and the gate modulation ceased at \(T_{\text{OP}} > 240\text{K}\). Dielectric loss spectroscopy of ion migration within the MAPbI\textsubscript{3} film provided evidence that at \(T_{\text{OP}} > 240\text{K}\), transport was affected by the temperature dependence of the ion migration and the associated screening of the gate electric field by negatively-charged iodide ions (Figure 2(c)) [11]. The dependence of \(\mu\) on \(T_{\text{OP}}\) in the films increased as the grain size decreased (Figure 2(d)). High-quality films with a large grain size were achieved by raising the precursor concentration to 0.75 M. Consequently, FET current modulation was observed at RT. Further modification of the gold S-D contact with a thin layer of pentfluorobenzene-tetrafluorobenzene-tetrachlorobenzene (PEIE) decreased it. MAPbI\textsubscript{3} FETs with PEIE-treated Au contacts showed \(\mu_e > 2 \text{cm}^2 \text{V}^{-1} \text{s}^{-1}\) at 100 K (Figure 2(e)) and \(\mu_e \approx 0.5 \text{cm}^2 \text{V}^{-1} \text{s}^{-1}\) at 300 K (Figure 2(f)).

The use of multiple cation/anion systems can also yield a detectable FET current modulation at RT. A mixed halide MAPb\textsubscript{1−x}Cl\textsubscript{x} perovskite exhibited an optimized film morphology with increased conductivity and carrier diffusion length [34]. The top-gate, bottom-contact MAPb\textsubscript{1−x}Cl\textsubscript{x} FET with treated gold S-D contacts showed balanced electron and hole transport with \(\mu \sim 1 \text{cm}^2 \text{V}^{-1} \text{s}^{-1}\) at RT. The S-D electrodes were treated with 2,3,5,6-tetrafluoro-4-(trifluoromethyl)benzenethiol (TTFB). However, the FET device yield and operational stability were limited. Similar FETs fabricated on a bottom SiO\textsubscript{2} gate dielectric layer showed no transistor behavior.

The stability of ambipolar top-gate, bottom-contact perovskite FETs has been increased by using triple-cation Cs\textsubscript{x}(MA\textsubscript{0.17}FA\textsubscript{0.83})\textsubscript{1−x}Pb(Br\textsubscript{0.17}I\textsubscript{0.83})\textsubscript{3} (0 \(\leq x \leq 30\%\)). The triple-cation perovskite films were defect-less and more stable thermally and structurally. The ambipolar perovskite FETs based on those films showed both \(\mu_e\) and \(\mu_h > 2 \text{cm}^2 \text{V}^{-1} \text{s}^{-1}\) at RT (Figure 3(c,d)). A complementary perovskite inverter was also fabricated that achieved an output gain > 20 (Figure 3(e,f)). The triple-cation FET exhibited improved stability under ambient conditions, possibly due to structural stabilization that suppressed the transition from the yellow hexagonal phase to the black phase [35].

The investigation of the mechanism of the chemical instability of perovskite FETs led to the development of a general approach to suppressing the ionic defects. The resulting perovskite FETs showed low hysteresis, high threshold voltage stability (\(\Delta V_t < 2 \text{V}\) over 10 h), and \(\mu > 1 \text{cm}^2 \text{V}^{-1} \text{s}^{-1}\) at RT (Figure 3(g,h)). The study demonstrated that Cs and Rb acted as passivation/crystallization modifying agents and could reduce the vacancy concentration and ion migration in
Figure 2. (a) Transfer characteristics of methyl ammonium lead iodide (MAPbI$_3$) FETs with bottom-gate, top-contact structures, versus their operating temperature ($T_{OP}$). The dielectric layer is SiO$_2$ (150 nm and 23 nF cm$^{-2}$) [14]. (b) $T_{OP}$-dependence of the carrier mobility in thin-film MAPbBr$_3$ perovskite FETs with a top-gate, bottom-contact device with Cytop as the dielectric layer (‘Cytop device’), and with a bottom-gate, bottom-contact device with SiO$_2$ gate as the dielectric layer (‘Si device’). At $T_{OP} > 240$ K, the gate modulation ceased [15]. (c) MAPbI$_3$ thin-film dielectric loss measurement as a function of the frequency and $T_{OP}$. (d) $\mu_{\text{FET}}(T)$ from MAPbI$_3$ FETs fabricated with different precursor concentrations of perovskite solutions and PEIE-modified S-D electrodes. Increases in the concentration yields increased the grain size. (e) Transfer characteristics of MAPbI$_3$ FETs with different S-D contact modifications at 100 K and (f) with PEIE-treated Au contacts at 300 K [11].
perovskite FETs; and that the treatment of the perovskite films with positive azo trope solvents that functioned as Lewis bases or acids further reduced the density of the mobile ionic defects [16].

Ion migrations in MAPbI$_3$ FETs and multiple-cation perovskite FETs were compared directly by monitoring the temporal variation of the channel current and the optical properties of the working device in the trans-diode mode ($V_{GS} = V_{DS} = 60$ V) [16]. An unusual increase in current after the first tens of seconds indicated the presence of ion migration, as the lateral ionic defect movement caused healing of iodine vacancies or a decrease in the contact resistance due to ion accumulation at the metal/perovskite interface, and thereby resulted in an increase in current. The characteristic time scale for the onset of ionic movement was 8 s for a channel length = 20 $\mu$m and 38 s for a channel length = 100 $\mu$m for MAPbI$_3$ FETs, and $10^3$ s for RbCsFAMAPbI$_3$ FETs. These results were consistent with the PL mapping data, which showed photo-induced brightening when iodine was depleted in perovskite FETs under bias (Figure 4(a–c)). Simply put, the difference in brightness before and after biasing is associated with the degree of ion migration. Ion migration was much lower in the RbCsFAMAPbI$_3$ FETs than in the MAPbI$_3$ devices (Figure 4(d,e)). Very recently, the same group developed a solvent-based surface cleaning and passivation technique that includes cleaning-healing-cleaning procedures, for suppressing migrating ionic defects in MAPbI$_3$ films, and demonstrated MAPbI$_3$ FETs that exhibited...
high n- and p-type mobilities of 3.0 and 1.8 cm² V⁻¹ s⁻¹, respectively, at RT [36].

Single crystals have significantly fewer ionic defects than polycrystalline perovskite films. Thus, they can operate in a regime in which structure–property relationships are no longer masked by ion migration, so the charge-transport physics of perovskite semiconductors can be studied cleanly. A spatially confined inverse temperature crystallization strategy can be used to synthesize micrometer-thin MAPbI₃, MAPbBr₃, and MAPbCl₃ single crystals with sub-nanometer surface roughness and very low surface contamination (Figure 5(a)) [37]. This strategy has allowed for the integration of single crystals into ambipolar FETs and has yielded record-high \( \mu_h \) values of up to 4.7 cm² V⁻¹ s⁻¹ and \( \mu_e = 1.5 \) cm² V⁻¹ s⁻¹ at RT in the bottom-gate, bottom-contact FET configuration (Figure 5(b–e)). For single-crystal FETs, the contact resistance \( R_C \) between S-D and the active layer can be high and can reduce the accuracy of the carrier mobility extraction. In this study, the calculated \( R_C \) values were 80, 150, and 195 kΩ cm for the bottom-gate, top-contact FETs; the \( R_C \)s were large enough to suppress the current at a low \( V_{DS} \) in the output curves. Then the device structure was optimized by growing single crystals directly on a substrate with pre-patterned gold electrodes. The resulting bottom-gate, bottom-contact devices excluded the effect of the crystal thickness and showed a record-high \( \mu \).

4. Tin-based perovskite FETs

Tin-based 2D perovskite FETs have benefited from a unique layered perovskite structure that significantly suppressed ion migration, so the current modulation was observable at RT even without elaborate optimization. The first such device was a 2D tin-based layered perovskite \((\text{PEA})_2\text{SnI}_4\) p-channel FET (Figure 6(a)), reported in 1999 [12]. The bottom-gate, bottom-contact \((\text{PEA})_2\text{SnI}_4\) FETs with SiO₂ as the dielectric layer exhibited pure p-channel behavior with a \( \mu_h \sim 0.6 \) cm² V⁻¹ s⁻¹ and a high on/off current ratio = \( 10^4 \) at RT (Figure 6(b)). For practical use, p-channel lead-free \((\text{PEA})_2\text{SnI}_4\) FETs are worthy of intensive efforts to realize low-power dissipation and noise-immune complementary circuits with well-developed n-channel transistors such as metal–oxide transistors, but higher FET performance is needed [38,39].

A series of studies was conducted to improve tin-based 2D perovskite FETs involved the use of a self-assembled monolayers (SAM) layer for \((\text{PEA})_2\text{SnI}_4\) growth, MoOₓ-treated S-D contacts, and Cytop as the dielectric layer in a top-gate, top-contact \((\text{PEA})_2\text{SnI}_4\) device configuration...
The optimized FETs achieved a peak $\mu_h$ of up to 15 cm$^2$ V$^{-1}$ s$^{-1}$, with negligible dual-sweep hysteresis (Figure 6(c)) [17]. By further optimizing $R_C$ and modifying the channel length, the same group demonstrated a $\mu_h = 26$ cm$^2$ V$^{-1}$ s$^{-1}$ and a $\mu_e = 4.6$ cm$^2$ V$^{-1}$ s$^{-1}$ (Figure 6(d)). The group recently developed exfoliated (PEA)$_2$SnI$_4$ crystal FETs that showed a high $\mu_h = 40$ cm$^2$ V$^{-1}$ s$^{-1}$ but a low fabrication yield ($< 1\%$) (Figure 6(e–g)) [20]. These findings suggest that carrier mobilities extracted from FETs do not necessarily represent the fundamental transport limits for the perovskites studied.
To determine the reasons for the limited device yield and to encourage further studies on perovskite FETs, a systematic study on (PEA)$_2$SnI$_4$ FETs, from precursor to film to device, has been conducted (Figures 7(a–c)) [21]. Universal approaches were developed to achieve high-performance and reliable (PEA)$_2$SnI$_4$ FETs. The methods included self-passivation of grain boundaries by excess phenethylammonium iodide, grain crystallization control by Lewis base adducts, and iodide vacancy passivation through oxygen treatment. The grain boundary passivation increased the FET reproducibility and reliability, and the grain-size enlargement boosted the FET performance. Going a step further, the authors demonstrated the first perovskite-based complementary inverter by combining n-channel indium gallium zinc oxide (IGZO) FETs. The inverters showed a high gain $>30$ with an excellent noise margin [21].

5. Perovskite FET stability

Perovskite FET stability mainly includes operational stability and environmental stability. Regarding stable operation, bias stability is important for display applications. For the most recent 3D MAPbI$_3$-based ambipolar FETs, the n-channel characteristics showed a positive threshold voltage ($V_{th}$) shift of $\sim$ 6 V ($V_G = 60$ V) under positive bias stress (PBS) for 12 h due to the trapping of electrons, and the p-channel characteristics showed a negative $V_{th}$ shift of $\sim$ 6 V ($V_G = -60$ V) under negative bias stress (NBS) due to hole trapping [36,41]. The corresponding $\mu_e$ and $\mu_h$ both dropped by $\sim$ 20%. The 2D p-channel (PEA)$_2$SnI$_4$-based FETs exhibited a $\sim$ 30% decrease in their normalized drain current under a constant NBS test ($V_{GS} = -40$ V and $V_{DS} = -40$ V for 2000 s) [21]. As commercialization is still distant, a stricter stability test that involves thermal (for example, 70°C) and illumination conditions, that is, PBT(I)S/NBT(I)S, will also be considered.

Some recent studies have focused on increasing the stability of perovskite FETs under ambient conditions. For example, stable operation of MAPbI$_3$ FETs in ambient air has been achieved by solution-processing MAPbI$_3$ films in a mixed solvent of DMF, DMSO, and hydroiodic acid under ambient air [40]. The tested MAPbI$_3$ FETs showed good operational and air stability for 120 days (Figure 8(a)) mainly due to improved film quality through proper crystallization and grain size, and through oxygen passivation of the grain boundary defects, where ionic migration occurs. The chemical vapor deposition (CVD) method has also been proven to effectively enhance the air stability of perovskite devices by producing smooth and void-free perovskite films. As the moisture absorption at the grain boundaries was reduced, the degradation of the perovskite composition was delayed [41,42].

Tin-based perovskite semiconductors often suffer from tin oxidation. Electrons in Sn 5s are more active than those in Pb 6s because of the shallower energy level of Sn 5s than of Pb 6s. Therefore, in contact with oxygen, Sn$^{2+}$ is easily oxidized to Sn$^{4+}$. Thermogravimetric analysis and in situ x-ray diffraction measurements suggested
Figure 7. (a) A systematic study on FETs that used (PEA)$_2$SnI$_4$ (PEAI) perovskite: (a) optimization of the starting materials; that is, (i) self-passage by adding slightly excess PEAI, (ii) anti-oxidation by Sn powder, and (iii) the halide element effect; (b) grain crystallization control using an adduct approach; (c) the device, including the investigation of trace oxygen doping on FETs and demonstration of the first complementary inverter based on perovskite-IGZO FETs.

Figure 8. (a) Transfer characteristics of FETs using MAPbI$_3$ films processed in air with DMF and with DMF + DMSO. The FETs were measured immediately after fabrication (day 1) and after aging in ambient air for 52 and 120 days [40]. (b) A new 2D perovskite (4Tm)$_2$SnI$_4$ structure and the transfer characteristics for a bottom-gate, top-contact (4Tm)$_2$SnI$_4$ FET with SiO$_2$ as the dielectric layer [23].

that tin-based perovskite materials oxidize and degrade as follows [43]:

$$2\text{SnI}_3 + \text{O}_2 \rightarrow 2\text{AlI} + \text{SnO}_2 + \text{SnI}_4 \quad (1)$$

However, tin oxidation is not an insoluble problem. (4Tm)$_2$SnI$_4$, a new tin-based 2D perovskite that features a linear $\pi$-conjugated oligothiophene ligand, 4Tm, which is a quaterthiophene derivative cation, 2-(3′,4′,5′,6′-dimethyl-[2,2′:5′,2″:5″,2‴-quaterthiophen]-5-yl) ethan-1-ammonium [44]. With (4Tm)$_2$SnI$_4$ as the channel layer, bottom-gate, top-contact perovskite FETs with SiO$_2$ as the dielectric layer achieved a $\mu_h$ of up to
Table 1. Performance of the 3D and 2D perovskites-based FETs.

| Metal halide perovskite | Structure | Dielectric layer | Interlayer/Processing a | Temp. (K) | $\mu_h$ (cm$^2$ V$^{-1}$ s$^{-1}$) | $\mu_e$ (cm$^2$ V$^{-1}$ s$^{-1}$) | Year | Ref. |
|-------------------------|-----------|------------------|-------------------------|----------|--------------------------------|--------------------------------|------|------|
| MAPbI$_3$               | BG-TC b   | SiO$_2$          | /                       | RT       | 10$^{-5}$                      | /                            | 2013 | [13] |
| MAPbI$_3$               | BG-BC     | SiO$_2$          | /                       | 78       | 2.1 × 10$^{-2}$                 | 7.2 × 10$^{-2}$                 | 2015 | [33] |
| MAPbI$_3$               | BG-TG     | SiO$_2$          |                        | 180      | 10$^{-2}$                      | 10$^{-5}$                      | 2015 | [14] |
| MAPbI$_3$               | TG-BC     | SiO$_2$          |                        | 200      | $\sim 3 \times 10^{-4}$        | /                             | 2017 | [11] |
| MAPbI$_3$               | TG-BC     | Cytop            | P3E                     | 300      | /                              | 0.5                          | 2017 | [11] |
| MAPbI$_3$               | TG-BC     | Cytop            | C-H-C                   | 100      | /                              | $>2$                         | 2020 | [36] |
| MAPbI$_3$               | BG-TG     | AlO$_x$          | DEStreated perov         | RT       | 23.2                          | /                            | 2020 | [45] |
| MAPbI$_3$               | BG-TC     | SiO$_2$          | Microplate              | 200      | 0.3                           | /                            | 2016 | [46] |
| MAPbI$_3$               | BG-TC     | SiO$_2$          | Microplate,Gr/BN          | 100      | $\sim 0.14$                    | /                            | 2017 | [47] |
| MAPbI$_3$               | BG-BC     | SiO$_2$          | Microplate,Gr/BN modified | 100      | $\sim 0.22$                    | /                            | 2017 | [47] |
| MAPbI$_3$               | BG-BC     | PVP:HDA          | /                       | 160      | /                              | $<1$                         | 2018 | [48] |
| MAPbBr$_3$              | BG-BC     | SiO$_2$          | SC$^c$,PFBT             | 300      | $0.055$                        | /                            | 2019 | [15] |
| MAPbBr$_3$              | BG-TC     | SiO$_2$          | S-D Au-Al               | 300      | $\sim 5$                      | $\sim 3$                     | 2017 | [49] |
| MAPbCl$_3$              | TG-BC     | Cytop            | TTFB-SAM                | RT       | 1.3                           | 1                            | 2015 | [34] |
| MAPbCl$_3$              | BG-BC     | SiO$_2$          | Sol. annealing          | RT       | $1 \times 10^{-2}$              | 7.3 $\times 10^{-3}$         | 2017 | [51] |
| MAPbCl$_3$              | BG-TC     | SiO$_2$          | Sol. vapor annealing    | RT       | $>10$                         | $>10$                        | 2018 | [52] |
| MAPbX$_3$, X = Cl, Br, I| TG-BC     | PMMA             | DTS                     | RT       | 2.1                           | 2.5                          | 2017 | [35] |
| RbCsFAMAPbI$_3$         | TG-BC     | Cytop            | Sol. treated perov      | RT       | /                             | 1.2                          | 2016 | [16] |
| FAMAPbI$_3$             | TG-BC     | Cytop            | Sol. treated perov      | RT       | 0.02                          | /                            | 2016 | [16] |
| MAPbX$_3$, X = Cl, Br, I| BG-TC     | SiO$_2$          | SC$^{c}$                | RT       | 4.7                           | 1.5                          | 2018 | [37] |
| (PEA)$_2$SnI$_4$        | BG-BC     | SiO$_2$          | /                       | RT       | 0.6                           | /                            | 1999 | [12] |
| (PEA)$_2$SnI$_4$        | BG-BC     | SiO$_2$          | Melt processed          | RT       | 2.6                           | /                            | 2002 | [53] |
| (PEA)$_2$SnI$_4$        | BG-BC     | SiO$_2$          | NH$_3$I-SAM             | RT       | 2.5                           | /                            | 2016 | [17] |
| (PEA)$_2$SnI$_4$        | BG-TC     | SiO$_2$          | NH$_3$I-SAM             | RT       | 5.7                           | /                            | 2017 | [17] |
| (PEA)$_2$SnI$_4$        | TG-TC     | Cytop            | MoO$_x$,NH$_3$I-SAM     | RT       | 15                            | /                            | 2017 | [17] |
| (PEA)$_2$SnI$_4$        | TG-TC     | Cytop            | C$_{60}$,NH$_3$I-SAM     | RT       | 2.1                           | /                            | 2018 | [18] |
| (PEA)$_2$SnI$_4$        | BG-TC     | SiO$_2$          | Vacuum deposition       | RT       | 0.78                          | /                            | 2016 | [54] |
| (PEA)$_2$SnI$_4$        | TG-TC     | Cytop            | MoO$_x$/C$_{60}$,NH$_3$I-SAM | RT     | 26                            | 4.6                          | 2017 | [55] |
| (PEA)$_2$SnI$_4$        | BG-BC     | SiO$_2$          | SC,MoO$_x$,N$_2$ gap at interface | RT | 40 | 35.5 | 2019 | [20] |
| (PEA)$_2$SnI$_4$        | BG-TC     | SiO$_2$          | Semi-CNT hybrid         | RT       | 1.5                           | /                            | 2019 | [22] |
| (PEA)$_2$SnI$_4$        | BG-TC     | SiO$_2$          | Precursor-film-FET optimization | RT | 3.5 | / | 2020 | [21] |
| (PEA)$_2$SnI$_4$        | BG-TC     | SiO$_2$          | Binary sol.             | RT       | 3.8                           | /                            | 2020 | [56] |
| (4-FPEA)$_2$SnI$_4$     | BG-BC     | SiO$_2$          | /                       | RT       | 0.48                          | /                            | 2001 | [57] |
| (4Tm)$_2$SnI$_4$        | BG-TC     | SiO$_2$          | /                       | RT       | 2.32                          | /                            | 2019 | [23] |

Table 1. Performance of the 3D and 2D perovskites-based FETs.

$^a$Thin-film channel layers were solution-processed unless otherwise noted; $^b$bottom-gate, top-contact; $^c$cleaning-healing-cleaning; $^d$single crystal; $^e$Research on perovskite-based FETs is still at a relatively early stage[4,58], and the device test and mobility calculation vary widely from lab to lab. It is suggested that: (i) the drain current be easily distinguishable from the gate leakage current; (ii) the capacitance of the dielectric layer be verified; (iii) the contact resistance between the channel layer and the S-D be small enough for precise mobility extraction; and (iv) the device reproducibility be reasonable.

2.32 cm$^2$ V$^{-1}$ s$^{-1}$, with good air stability for 30 days [23]. In contrast, the pristine (PEA)$_2$SnI$_4$ perovskite-based FETs could not survive for 1 day. There are two main reasons for the increased stability of (4Tm)$_2$SnI$_4$ in air. First, the $\pi$-conjugated 4Tm molecule is hydrophobic and bulky. These traits decrease the penetration of oxygen and moisture through this barrier. Second, the intermolecular interactions between the conjugated 4Tm ligands provide an extra driving force for the stabilization of the hybrid crystal, thereby increasing the intrinsic stability of (4Tm)$_2$SnI$_4$. This study demonstrates that molecular engineering provides new possibilities to achieve highly stable tin-based perovskite electronics. Table 1 summarizes the performance of the 3D and 2D perovskites-based FETs reviewed in this paper.

6. Summary and outlook

We have reviewed the improvements in 3D lead-based and 2D tin-based metal halide perovskite FETs. From the energy-band aspect, lead-based perovskites generally exhibit ambipolar behavior because of their unique band structure and similar hole/electron effective masses. Tin-based perovskites show stronger hole transport because their VBM features I 5p-Sn 5s hybrid states, and their tin vacancies have a lower formation energy than do lead vacancies. Concerted efforts have been made to reduce migrating ionic defects in 3D perovskite FETs, to reveal their genuine charge carrier transport property. Carrier mobilities have been consistently increased. 2D tin-based perovskite FETs can show neat p-channel transport with a low ‘off-state’ current, but their relatively insulating and...
bulky organic part may limit the further increase of $\mu$ at some point because of the trade-off between the carrier mobility and the stabilization of the perovskite structure to protect the inorganic composition from oxidation. 3D tin-based perovskites are potentially promising for the significantly needed high-performance p-channel FETs, which would be comparable to the commercialized metal oxide n-channel devices for complementary electronics. 3D tin-based perovskite FETs have not been reported yet but are expected to be realized soon. The stability of the perovskite phase and the operational/environmental stability of the device should also be considered.

**Disclosure statement**

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**Notes on contributors**

**Huihui Zhu** received her B.S. and M.S. degrees in Microelectronics and Materials Engineering from Qingdao University, China in 2014 and 2016, respectively. She is currently a Ph.D. candidate at the Department of Chemical Engineering of the Pohang University of Science and Technology (POSTECH) under the supervision of Prof. Yong-Young Noh. Her research interests include perovskites and metal oxides for thin-film transistors.

**Ao Liu** received his B.S. and M.S. degrees from the Department of Physics of Qingdao University in China in 2014 and 2017, respectively. He is currently a Ph.D. candidate at the Department of Energy and Materials Engineering of Dongguk University under the supervision of Prof. Yong-Young Noh. His research interests include the development of solution-processed metal oxide/organic materials and their applications in transistors and circuits.

**Yong-Young Noh** is Chair Professor at the Department of Chemical Engineering of the Pohang University of Science and Technology (POSTECH) in Pohang, Republic of Korea. He received his Ph.D. degree in 2005 from the Gwangju Institute of Science and Technology (GIST) in the Republic of Korea, and then worked as a postdoctoral associate at the Cavendish Laboratory in Cambridge, UK. Afterwards, he worked at the Electronics and Telecommunications Research Institute as a senior researcher; at Hanbat National University as assistant professor; and at Dongguk University-Seoul as associate professor. He won the Merck Young Scientist Award in 2013, the Korea President Award in 2014, and the Institute of Electrical and Electronics Engineers (IEEE) George E. Smith Award in 2014, and was selected as the Scientist from the Korea Government in September 2016. He has published over 340 papers in international journals in the field of materials for electronics and optoelectronic devices, particularly OFETs, OLEDs, metal halides, perovskites, carbon nanotubes, and oxide TFTs.

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