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Enhanced electrocatalytic hydrogen evolution by molybdenum disulfide nanodots anchored on MXene under alkaline conditions†

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Efficient hydrogen production through electrocatalysis represents a promising path for the future clean energy. Molybdenum disulfide (MoS₂) is a good substitute for platinum-based catalysts, due to its low cost and high activity. However, the limited active sites and low electrical conductivity of MoS₂ hinder its large-scale industrial application under alkaline conditions. Herein, we constructed MoS₂ nanodots anchored on an MXene/nickel foam (MoS₂ NDs/MXene/NF) heterostructure by a cascade polymerization synthesis and *in situ* vulcanization. The prepared heterostructure displays an ultralow overpotential of 94 mV at a current density of 10 mA cm⁻² with a Tafel slope of only 59 mV dec⁻¹ in alkaline (1 M KOH) hydrogen evolution reaction (HER), and is better than conventional MoS₂ electrocatalysts reported so far. Fine structural analysis indicates that MoS₂ NDs are dispersed uniformly on the surface of the heterostructure with consistent orientation, leading to the improvement of MoS₂ conductivity with more paths for electron transfer. Moreover, the orientation of the synthesized MoS₂ NDs was verified to expose the more (002) crystal plane, which exhibits higher activity than other planes. Our results demonstrate that MoS₂ NDs with heterostructure design and preferential growth can serve as high-efficiency noble-metal free electrocatalysts for the HER in alkaline solution.

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Introduction

Hydrogen energy is deemed a promising alternative to fossil fuels, due to its features of high energy capacity and environmental friendliness.¹⁻³ The efficient and economical way of hydrogen production from electrochemical water splitting is a long-standing goal for the energy crisis solution.⁴⁻⁶ Despite the HER being easier under acidic conditions, accompanying corrosion of electrolytic equipment and destruction of catalysts become a huge bottleneck for its industrial application.⁷ Instead, alkaline electrolysis, due to the long durability of the electrolyzer, the high reliability of the production process, and the long life of the catalyst, has attracted tremendous attention in the past decade.⁸ Platinum and its derivatives till now are still the most common HER catalysts, both under acidic and alkaline

conditions.^{9,10} Considering the scarcity of noble-metal catalysts, the development of sustainable catalysts with low cost and abundance has been the focus of intensive research efforts.¹¹

The concentration of hydrogen ions in alkaline medium is ~8 to 14 orders of magnitude lower than that in acidic medium, resulting in the drawbacks of sluggish water adsorption and poor dissociation dynamics during water splitting. Thus, greater efforts on increasing the accessibility of active sites and surface modification with water adsorption components have been adopted to develop high-performance alkaline HER electrocatalysts.¹¹⁻¹⁴ Recently, non-precious Mo-based materials have been proved experimentally to replace platinum and its derivatives,^{11,12} because of the lower Gibbs-energy of hydrogen adsorption analogous to platinum. Typically, MoS₂ with a two-dimensional layered nanostructure, recognized as the most important one, shows catalytic sites at its layer edges during the Volmer reaction.¹⁵⁻¹⁹ The vertical growth of MoS₂ on a flat substrate or nanowire surfaces can expose more edge sites.^{20,21} Meanwhile, MoS₂ nanosheets, even ultrasmall nanodots, are preferable because the accompanying higher specific surface area and more edge atoms strikingly boost the HER catalytic activity.²²⁻²⁴ Worse still, the nanostructures tend to agglomerate after being formed into electrodes, which hinders sustainable electrocatalytic activity, thus far from being satisfactory for practical electrolysis.^{25,26} As for surface modification to enhance the catalyst hydrophilicity and cleave the HO-H bond, nano-assembly hierarchical structures, such as MoS₂-Ni/Co

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compounds with porous media,¹⁴ have been constructed and used as highly efficient electrocatalysts. However, the even worse conductivity of Ni/Co compounds, such as hydroxide, sulfide, and phosphide, causes more energy waste in hydrogen production.

The electrical conductivity enhancement of electrocatalysts can simultaneously accelerate the reaction rate and reduce overpotentials. Due to the semiconducting properties of molybdenum disulfide, a minimized bandgap means less difficulty for carrier migration. Nevertheless, the bulk form of MoS₂ tends to have a smaller bandgap (1.2 eV) than that of few-layer MoS₂ (1.9 eV for monolayer MoS₂). In another word, the MoS₂ nanosheets or nanodots possess poor electrical conductivity. Thus, defect engineering strategies, such as introducing holes, cracks, and crimps, have been extensively developed and utilized to minimize the bandgap, and achieve higher active site exposure^{22,27} at the same time. However, most defects influence slightly the intrinsic property regulation of MoS₂. In this context, introducing a highly conductive substrate could simultaneously optimize the HER performance of MoS₂ nanosheets or nanodots. Of all substrates, the Ti₃C₂T_x MXene shows great promise due to its good hydrophilicity and metallic conductivity (up to 10 000 S cm⁻¹), having been applied in supercapacitors, water treatment, microwave absorption,²⁸ and electromagnetic interference shielding.¹⁰ The structure design between MXene and MoS₂ is assumed to improve the conductivity of the electrocatalyst and deliver enough activity toward the HER, while the remaining challenges of uniform dispersion of catalytically active sites and effective protection of the MXene conductive substrate from surface oxidization and anti-aggregation should be overcome first.^{16,20,29}

Motivated by the above considerations, we developed an MXene/NF substrate with preferential growth MoS₂ nanodots (MoS₂/MXene/NF), of which the more active edge exposure enhances the HER performance. Notably, the chelated molybdate ions and Ti₃C₂T_x MXene flakes encapsulated by polydopamine (PDA), used as a precursor, provide uniformly dispersed MoS₂ nanodots anchored on MXene with ultra-small size and inhibit surface oxidation of MXenes during annealing and sulfurization treatment.³⁰ The synthesized MoS₂/MXene heterostructure, confirmed by morphology, structure, phase, and elemental chemical valence characterization, displays a synergistic effect of active MoS₂ nanodots and conductive Ti₃C₂T_x MXene flakes for fast charge transfer. The optimal MoS₂/MXene/NF yields remarkable HER activities with a Tafel slope of 59 mV dec⁻¹ and an overpotential of 94 mV at 10 mA cm⁻². Due to its excellent stability over 2000 cycles of cyclic voltammetry tests in an alkaline electrolyte, MoS₂/MXene/NF has been certified superior to most non-noble metal catalysts in alkaline solution. This work sheds new insight into nanostructure construction for Mo-based catalysts to enhance their capability of catalytic water decomposition.

Results and discussion

Fig. 1 shows the schematic illustration of a novel strategy for constructing MoS₂ NDs/MXene with abundant exposed sites grown on a 3D self-supporting electrode. First, ultrathin

MXenes were obtained by selectively etching the Al element in the MAX phase (Ti₃AlC₂) with hydrofluoric acid and the subsequent exfoliation procedure (experimental section in the ESI†). During the *in situ* polymerization process of PDA under weakly alkaline conditions, the mixture of few-layer MXenes and molybdate ions was well encapsulated on the NF substrate surface. The molybdate ions were chelated in PDA (Mo-PDA) as the precursor of MoO₃, and few-layer MXenes were fully wrapped with Mo-PDA. Then, a heterostructure of Ti₃C₂T_x MXene sheets decorated with MoO₃ NDs was obtained by annealing Mo-PDA/Ti₃C₂T_x/NF at 650 °C under a N₂ atmosphere, which inhibits the agglomeration of MoO₃ NDs and oxidation of Ti₃C₂T_x due to the confinement effect of PDA. Finally, as-synthesized MoO₃ NDs/MXene was controllably sulfurized to MoS₂ NDs/MXene at sulfurization temperatures from 500 to 800 °C in a N₂ atmosphere.

The products of the schematic step are characterized by scanning electron microscopy (SEM) and transmission electron microscopy (TEM) to verify the synthetic process. The as-prepared MXene shows a layered structure in the SEM image (Fig. S1b†).³¹ The clear few-layered structure of 2D MXene is further confirmed by the TEM image (Fig. S2a†). By *in situ* polymerization of dopamine, the relatively smooth surface of NF with micropores (Fig. 2a) is wrapped by a uniformly layered structure network combined with 2D MXene and Mo-PDA (Fig. S1d†). After annealing at 650 °C, the Mo-PDA transforms into MoO₃ NDs dispersed in the MXene, while the MoO₃/MXene/NF (Fig. 2b) shows a similar morphology to the Mo-PDA/MXene/NF precursor (Fig. S1d†). Moreover, the well-defined ND decorated 2D MXene structure was maintained standing on NF after the sulfurization process (Fig. 2c). The uniform distributions of MoO₃ NDs and MoS₂ NDs in carbide are more clearly seen in the TEM images (Fig. 2d and e). As for Mo-PDA/NF and its derivative MoS₂/NF (Fig. S3a-c†), a flower-like structure with about 400 nm diameter was formed on the NF surface, which is similar to the folded surface in Mo-PDA/MXene/NF (Fig. S1d†). This revealed that the Ti₃C₂T_x with oxygen- or fluorine-containing terminations was identified as a charged substrate to adsorb dopamine hydrochloride chelated Mo precursors, realizing well dispersed nanostructured Mo compounds.^{32–34}

The structures of MXene and its composites were further investigated by X-ray diffraction (XRD) and high-resolution TEM (HRTEM). The XRD pattern of MXenes (Fig. S2c†) shows a strong diffraction peak at 6.9°, corresponding to the typical (002) plane of Ti₃C₂T_x.^{33,35} Meanwhile, the HRTEM image (Fig. S2b†) can also index the lattice fringes of 2D MXene. As for NF (Fig. S4†), three strong peaks located at 44.7, 52.0, and 76.5 can be well identified as Ni (JCPDS no. 04-0850). The diffraction peaks of NF didn't shift or change during the processes of *in situ* polymerization and high-temperature treatment. In addition, the (002) Ti₃C₂T_x diffraction peak disappeared in the XRD pattern of MoO₃/MXene/NF and MoS₂/MXene/NF (Fig. S4†), which should be attributed to the PDA decomposition and carbon layer restacking on the surface of MXene flakes.^{30,36} What's more, no obvious crystal diffraction peaks of MoO₃ and MoS₂ were found in XRD patterns with NF due to the ultra-small size of NDs and the extremely strong diffraction peak of NF.²⁴





Fig. 1 Schematic diagram of the MoS₂ NDs/MXene/NF synthetic route.

The enlarged TEM image (Fig. 2e) exhibits ultrasmall MoO₃ nanodots uniformly decorated on the surface of MXene. Lattice fringes of both MoO₃ and MXene can be found in the HRTEM

image (Fig. 2f). The carbon-coated Ti₃C₂T_x MXene derived from PDA/Ti₃C₂T_x retains the flat surface without the presence of TiO₂ species, which demonstrated that tightly covered PDA on



Fig. 2 SEM images of (a) NF, (b) MoO₃/MXene/NF, and (c) MoS₂/MXene/NF. The insets of (b and c) are images at high magnification. (d and e) TEM images, and (f) HRTEM image of MoO₃/MXene removed from MoO₃/MXene/NF. (g and h) TEM images, (i) HRTEM image, (j) HAADF image, and (k) EDX elemental distributions of MoS₂/MXene removed from MoS₂/MXene/NF.



the MXene surface could inhibit surface oxidation during the annealing process, to maintain excellent electrical conductivity. The morphologies of the $\text{Ti}_3\text{C}_2\text{T}_x$ MXene and molybdenum derivative are slightly changed after high-temperature sulfuration (Fig. 2g). Even smaller MoS_2 nanodots were observed on the MXene surface (Fig. 2h) after the MoO_3 vulcanization process, meaning the existence of multiple reaction points resulting in fission of MoO_3 dots at the end of the vulcanization. The corresponding HRTEM image revealed lattice fringes of 0.62 nm on the nanodot matching with the (002) plane of 2H- MoS_2 (Fig. 2i). The lattice of MoS_2 has an obvious distortion compared with MoO_3 due to the unavoidable vacancies appearing during high-temperature treatment and the two-dimensional microstructure wrinkling in MoS_2 . Further analysis revealed that most of the sulfurized MoS_2 preferentially grows in the vertical direction of $\text{Ti}_3\text{C}_2\text{T}_x$ MXene flakes with (002) crystal plane exposure. The high-angle annular dark field (HAADF) image under scanning TEM mode (Fig. 2j) indicates abundant bright dots anchored on the MXene flakes. The corresponding elemental distribution images under energy-dispersive X-ray spectroscopy (EDX) show the distributions of Ti, C, Mo, and S elements (Fig. 2k), further confirming the uniform MoS_2 NDs/MXene heterostructure.

Given that EDX and HRTEM only provide foundational composition information, the X-ray photoelectron spectroscopy (XPS) analysis is further used to elucidate the electronic valence of compositions and coordination structures. Fig. S5a† shows the full spectrum of $\text{MoO}_3/\text{MXene}$ removed from $\text{MoO}_3/\text{MXene}/\text{NF}$, which evidences the existence of Ti, C, Mo, O, and N. The peak of N may stem from the carbonized PDA. The signal attenuation caused by the active material package is scanned under high-resolution. By calibrating the carbon peak C (284.8 eV), +0.7 eV peak shifted, the high-resolution XPS spectrum of Mo 3d is obtained in Fig. S5b.† The peaks at 229.9, 233.2, and 236.3 eV, which can be attributed to the $\text{Mo}^{4+} 3d_{5/2}$, $\text{Mo}^{6+} 3d_{5/2}$, and $\text{Mo}^{6+} 3d_{3/2}$, are assigned to the existence of MoO_3 after 600 °C high-temperature treatment. In addition, the deconvolution of the Ti 2p XPS spectrum (Fig. S5c†) can be fitted with TiO_2 (Ti^{4+}), Ti-X (Ti^{2+} , X = OH), Ti-C (Ti^{+}), and Ti-X (Ti^{2+} , X = F) at the peaks of 455.6, 458.4, 460.9, and 464.3 eV, respectively. Compared to the brand-new $\text{Ti}_3\text{C}_2\text{T}_x$ MXene after etching, the binding energy of Ti-C and Ti-X positively shifts,³⁷ revealing the obvious electron transfer from $\text{Ti}_3\text{C}_2\text{T}_x$ to MoO_3 nanodots in the heterostructure.

After further vulcanization to obtain $\text{MoS}_2/\text{Ti}_3\text{C}_2\text{T}_x$, the signal peak of sulfur can be found initially in the full spectrum peak (Fig. 3a). In the corresponding Mo 3d spectrum (Fig. 3b), the distinct peaks at 229.6 and 232.7 eV are assigned to the $\text{Mo}^{4+} 3d_{5/2}$ and $\text{Mo}^{4+} 3d_{3/2}$ in MoS_2 , whereas the signals at 226.9 and 234.4 eV can be attributed to the S and Mo-O (Mo^{6+}), respectively. The S 2p convolutional high-resolution spectrum (Fig. 3c) is divided into two peaks of S $2p_{3/2}$ and S $2p_{1/2}$ at 162.1 and 163.5 eV, revealing the -2 oxidation state of S in MoS_2 . These phenomena are consistent with previous studies.^{22,39} As for the Ti 2p XPS spectrum after vulcanization (Fig. 3d), similar to that of $\text{MoO}_3/\text{Ti}_3\text{C}_2\text{T}_x$, signals appear at 455.7, 459.1, 461.2, and 464.8 eV, stemming from TiO_2 (Ti^{4+}), Ti-X (Ti^{2+} , X = OH), Ti-C

(Ti^{+}), and, Ti-X (Ti^{2+} , X = F). Compared to the MoO_3 nanodots on $\text{Ti}_3\text{C}_2\text{T}_x$, the Ti-C and Ti-X binding energies of $\text{Ti}_3\text{C}_2\text{T}_x$ in the $\text{MoS}_2/\text{Ti}_3\text{C}_2\text{T}_x$ heterostructure positively shift about 0.3 eV and 0.7 eV, while the $\text{Mo}^{4+} 3d_{5/2}$ peak shifts negatively from 229.9 to 229.6 eV in Mo-S. The decreasing Mo 3d binding energy and increasing Ti 2p binding energy indicate that more electrons were transferred from $\text{Ti}_3\text{C}_2\text{T}_x$ to MoS_2 nanodots. The above XPS results reveal that MoS_2 nanodots are strongly adhered to the $\text{Ti}_3\text{C}_2\text{T}_x$ MXene surface, accompanied by strong interfacial interactions existing at the $\text{Ti}_3\text{C}_2\text{T}_x$ - MoS_2 heterogeneous interfaces.

To evaluate the electrocatalytic activity of $\text{MoS}_2/\text{MXene}/\text{NF}$ towards water splitting, HER performance was investigated in a traditional three-electrode system, in which a N_2 -saturated 1 M KOH solution was employed as the electrolyte. For comparison, NF, MXene/NF, and MoS_2/NF $\text{MoO}_3/\text{MXene}/\text{NF}$ were also tested as control samples. Fig. 4a shows the steady-state linear sweep voltammograms (LSV) of the above electrodes, whose activity trend is as follows: NF < MXene/NF < $\text{MoO}_3/\text{MXene}/\text{NF}$ < $\text{MoS}_2/\text{MXene}/\text{NF}$. The MXene/NF and NF show relatively worse HER activity, suggesting that both MXene and NF are not very suitable for the HER. The $\text{MoS}_2/\text{MXene}/\text{NF}$ exhibits a low onset overpotential of 94 mV at 10 mA cm^{-2} (η_{10}) vs. reversible hydrogen electrode (RHE), which is much better than those of MoS_2/NF ($\eta_{10} = 164$ mV, Fig. S6a†), $\text{MoO}_3/\text{MXene}/\text{NF}$ ($\eta_{10} = 204$ mV) and MXene/NF ($\eta_{10} = 189$ mV) (Fig. 4c). The Tafel slopes, as a pivotal kinetic parameter to elucidate the HER mechanism, were calculated using the Tafel equation. Here, the Tafel slope of $\text{MoS}_2/\text{MXene}/\text{NF}$ is 59 mV dec^{-1} (Fig. 4b), and the smallest among all the compared samples.

The Tafel slope curves indicate that $\text{MoS}_2/\text{MXene}/\text{NF}$ has extremely fast reaction kinetics. As shown in Fig. 4d and Table S1,† the comparison of HER performance between $\text{MoS}_2/\text{MXene}/\text{NF}$ and other non-noble-metal electrocatalysts from the literature is presented.^{14,23,29,35,38,39} Most of them require larger overpotentials (η_{10} over 100 mV) and higher Tafel slopes (over 60 mV dec^{-1}), which are far worse than the catalytic performance of Pt. Some non-noble metal catalysts may have a low overpotential ($\eta_{10} < 50$ mV), while their poor kinetic conditions (higher Tafel slopes exceeding 50 mV dec^{-1}) also make the performance inexhaustible. Therefore, $\text{MoS}_2/\text{MXene}/\text{NF}$ with an ultra-low overpotential ($\eta_{10} = 94$ mV) and Tafel slope (59 mV dec^{-1}) have great potential to replace Pt in industrial applications. Meanwhile, the stability of the electrocatalyst, which is a critical indicator for catalyst practical application, has also been investigated through cyclic voltammetry (CV) cycles over 200 mV vs. RHE (Fig. 4f) and the chronoamperometry (CP) test at a constant overpotential (Fig. S7d†). The LSV curve of $\text{MoS}_2/\text{MXene}/\text{NF}$ shows a negligible negative shift after 2000 cycles in current density compared to the initial curve. Meanwhile, the $\text{MoS}_2/\text{MXene}/\text{NF}$ exhibits greater long-term stability, only a slight decrease over 60 000 s. SEM and XRD characterization studies of $\text{MoS}_2/\text{MXene}/\text{NF}$ (Fig. S7a-c†) were further conducted to confirm the outstanding durability of its morphology, structure, and phase after the stability test. Similar to MoS_2/NF after electrochemical tests (Fig. S3d†), the SEM image shows that only the folded surface of the PDA



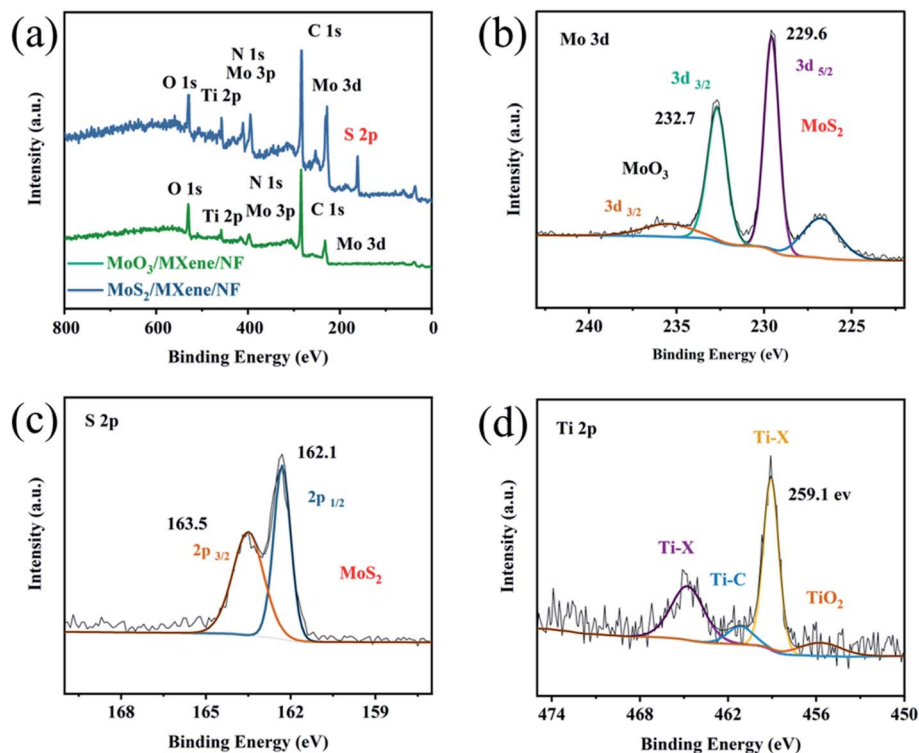


Fig. 3 (a) Full spectra of $\text{MoO}_3/\text{MXene}$ and $\text{MoS}_2/\text{MXene}$ removed from $\text{MoO}_3/\text{MXene}/\text{NF}$ and $\text{MoS}_2/\text{MXene}/\text{NF}$. High-resolution XPS (b) Mo 3d (c) S 2p, and (d) Ti 2p spectra of $\text{MoS}_2/\text{MXene}$.

carbonation layer is destroyed, while the main skeleton structure and phase of $\text{MoS}_2/\text{MXene}/\text{NF}$ remain unchanged. The excellent durability of $\text{MoS}_2/\text{MXene}/\text{NF}$ can be attributed to the confinement effect of carbonated PDA, which protects the ion-exchange interface during the electrocatalytic process.

The above results imply that the HER activity of both $\text{Ti}_3\text{C}_2\text{T}_x$ and MoS_2 is unsatisfactory under alkaline conditions, whereas modulating the heterogeneous structure can result in an optimized HER activity. To better understand the function of the $\text{MoS}_2/\text{Ti}_3\text{C}_2\text{T}_x$ heterostructure during the HER process, the electrochemical impedance spectroscopy (EIS) measurement was performed at 200 mV vs. RHE from 10^6 Hz to 0.01 Hz. Fig. S8† displays the Nyquist plots of the above electrodes, in which the charge transfer resistance (R_{ct}) is related to the electrocatalytic kinetics at the interfaces. The $\text{MoS}_2/\text{MXene}/\text{NF}$ delivers a much smaller R_{ct} than $\text{MoO}_3/\text{MXene}/\text{NF}$ and MXene/NF , evidencing faster electron transfer between the $\text{MoS}_2\text{-Ti}_3\text{C}_2\text{T}_x$ heterostructure. Both the $\text{Ti}_3\text{C}_2\text{T}_x$ flake and MoS_2 nanodot contribute to the higher electrical conductivity and faster HER kinetics. Generally, the effective electrochemically active area (ECSA) is related to the electrocatalyst surface area and the number of active sites, which is estimated from the double-layer capacitance (C_{dl}) by performing cyclic voltammetry with different sweep speeds under the voltage range of the non-Faradic current (electrochemical tests). The C_{dl} value was calculated by the slope between the half capacitive current density at the median of the applied potential and the scan rates. The measured C_{dl} value for $\text{MoS}_2/\text{MXene}/\text{NF}$ is 0.23 F

cm^{-2} (Fig. 4e and S8†), considerably better than those of $\text{MoO}_3/\text{MXene}/\text{NF}$ (0.39 F cm^{-2}), MXene/NF (1.8 F cm^{-2}), and NF (2.6 F cm^{-2}). As the MXene/NF displays the worst ECSA, the HER active sites are assumed from the MoO_3 and MoS_2 nanodots in the heterostructure. Significantly, the ECSA of $\text{MoS}_2/\text{MXene}/\text{NF}$ is larger than that of $\text{MoO}_3/\text{MXene}/\text{NF}$, indicating the extension of the electrochemical active area or the more catalytic active sites obtained after high-temperature sulfurization. Experimental studies and theoretical calculations have certified that MoS_2 active sites are mainly concentrated on the edge.^{15–21} Therefore, the fixed orientation for MoS_2 NDs preferential growth on MXene flake exposes more (002) crystal planes, which exhibit more active sites for excellent HER than other planes.

Obviously, the HER performance is associated with the varied ratio of $\text{MoS}_2/\text{MXene}$ and the construction of $\text{MoS}_2\text{-MXene}$ interfaces. The different ratios of $\text{MoS}_2/\text{MXene}$ can be adjusted by the usage of ammonium molybdate and MXene in the precursor reaction process. Though more Mo active sites are beneficial for HER performance enhancement, excessive Mo will result in larger MoS_2 dots, which have lower conductivity and a smaller percentage of active sites. What's more, a larger ratio of MoS_2 and MXene may be accompanied by the formation of MoS_2 microspheres (Fig. S3†), of which the HER performance is very poor (Fig. S6†). As for the role of the $\text{MoS}_2\text{-MXene}$ interface, it was known that $\text{Ti}_3\text{C}_2\text{T}_x$ itself shows negligible HER activity. So, the function of Ti-based MXene s at the interfaces is to provide a better hydrophilicity and conductivity, which will promote more efficient and direct electron transfer (Fig. 3) to



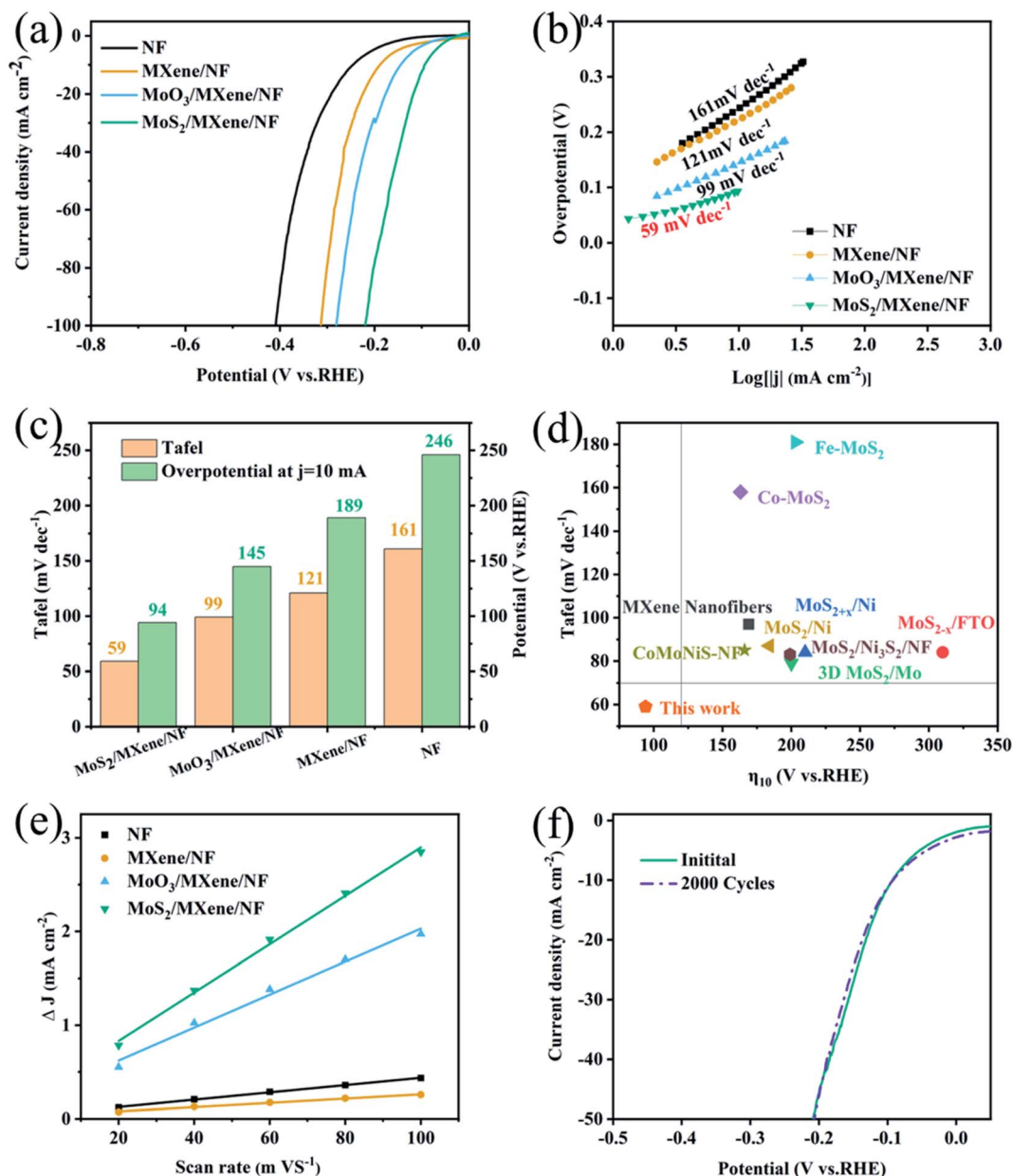


Fig. 4 Electrochemical HER performance. (a) Polarization curves and (b) corresponding Tafel plots of MoS₂/MXene/NF, MoO₃/MXene/NF, MXene/NF, NF in 1 M KOH. (c) Comparison of overpotentials at 10 mA cm⁻² (η₁₀) and Tafel slopes for prepared samples. (d) Electrocatalytic activity comparison of MoS₂ based materials.^{14,23,29,35,38,39} (e) LSV curves of MoS₂/MXene/NF before and after 2000 cyclic potential scans. (f) Linear fits of half capacitive currents vs. scan rates for the extraction of C_{dl}.

improve the electrocatalytic performance. Meanwhile, the exposures of MoS₂ edges in the outside of interfaces provide active sites for the generation of H^{*}.

The effect of sulfurization temperatures (500 °C and 800 °C) on HER activity was evaluated. The SEM image (Fig. 5a) of MoS₂/MXene/NF powder obtained after sulfurization treatment at 500 °C displays negligible change in morphology, and structure, while obvious particles appear in MXene flakes (Fig. 5b), identified as TiO₂ (Fig. S10,† JCPDS no. 21-1271) when increasing the sulfurization temperature to 800 °C. As shown in Fig. 5c, the overpotentials for achieving 10 mA cm⁻² for samples treated at

500 °C and 800 °C are 151 and 128 mV, respectively. Meanwhile, the corresponding Tafel slopes are 128 and 126 mV dec⁻¹ (Fig. 5d), respectively. Obviously, the MoS₂/MXene/NF treated at 650 °C exhibits the best HER activity among the three samples. The deteriorating HER performance of the MoS₂/MXene/NF treated at 800 °C should be attributed to the high temperature induced oxidation of MXene which would reduce the MoS₂/MXene conductivity and block the electron transfer channel. The MoS₂/MXene/NF treated at low temperature (500 °C) has even poorer intrinsic activity than initial MoO₃/MXene/NF (Fig. 4a and b), which may be related to the nanostructure



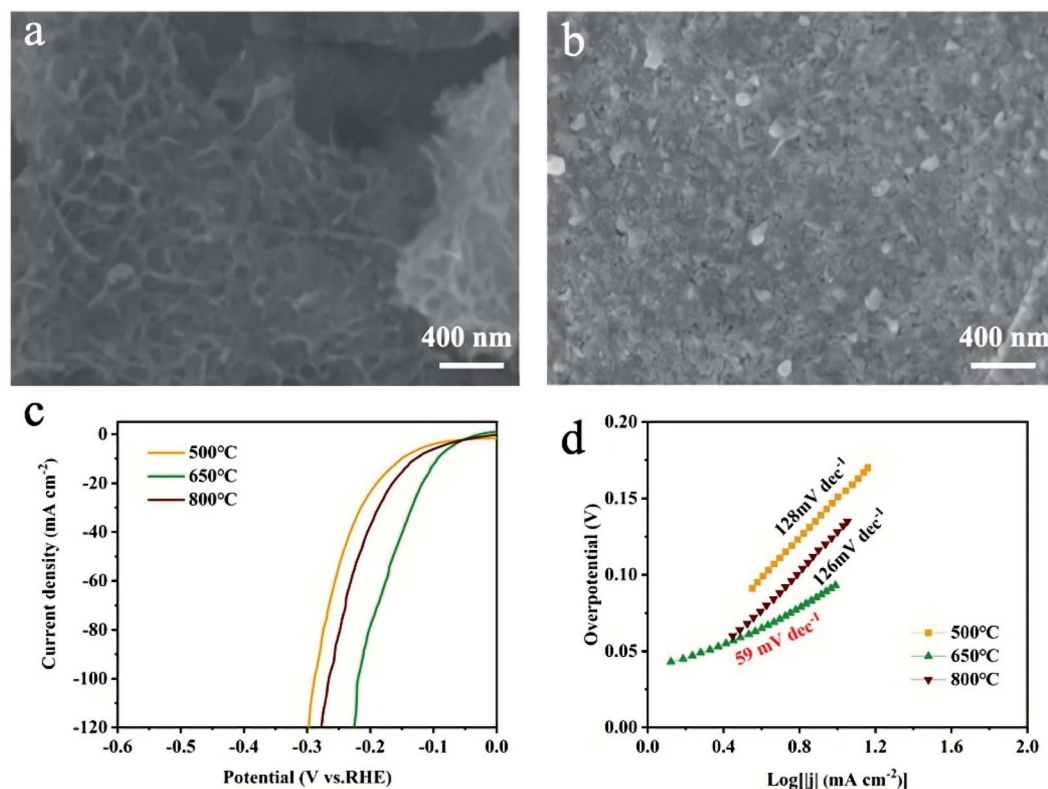


Fig. 5 SEM images of MoS₂/MXene/NF after sulfurization at (a) 500 °C and (b) 800 °C. (c) Polarization curves and (d) corresponding Tafel plots of MoS₂/MXene/NF sulfurized at different temperatures.

destruction of MoO₃ and MoS₂ under insufficient sulfurization. According to the literature,^{32,39,40} temperature can control the degree of sulfurization, and molybdenum disulfide is continuously formed on the surface of molybdenum oxide at high temperatures. For the MoO₃/MXene/NF electrode at low temperature, few active sites on the molybdenum disulfide edge can be exposed. And molybdenum oxide is destroyed in the sulfurization process at the same time, so the active sites of the catalyst are greatly reduced. What's more, the alkaline HER process includes two microscopic steps:^{41,42} (a) H₂O + e⁻ → H* + OH⁻ (Volmer reaction); (2) H* + H₂O + e⁻ → H₂ + OH⁻ (Heyrovsky reaction) or H* + H* → H₂ (Tafel reaction). The Tafel slope of MoS₂/MXene/NF (500 °C) is 128 mV dec⁻¹ (>120 mV), indicating that the Volmer reaction is the rate-limiting step of the HER due to the deficiency of active sites. The Tafel slope of MoO₃/MXene/NF and MoS₂/MXene/NF (650 °C) declined to 99 and 59 mV dec⁻¹, respectively, implying that the Heyrovsky reaction or Tafel reaction has transitioned to the rate-determining step. A comparison experiment further indicated that both MXene flakes with higher conductivity and MoS₂ nanodots with more active site exposure are significant to MoS₂/MXene/NF with extraordinary HER performance.

Based on the aforementioned analysis, the remarkable HER activity of MoS₂/MXene/NF (650 °C) is assumed to originate from the following factors: (1) *in situ* polymerized PDA with chelated molybdate ions plays a key role in avoiding Ti₃C₂T_x aggregation and protecting the structure; (2) PDA chelated

MoO₄²⁻ makes the molybdenum compound highly disperse ultrasmall nanostructures on the surface of the MXene flakes; (3) the special heterostructures of MoS₂ nanodots and Ti₃C₂T_x flakes guarantee excellent HER performance and long-term stability; (4) after sulfurization at an appropriate temperature, MoS₂ nanodots grow vertically on Ti₃C₂T_x MXene flakes, accompanied by the high accessibility of active sites for the HER at the MoS₂ nanodot edge; (5) the intense synergistic effect of the MoS₂-Ti₃C₂T_x heterostructure provides a resistance-less channel for faster electron transfer when the active sites of MoS₂ generate H*.

Conclusion

In summary, a high-activity and high-stability MoS₂/MXene/NF catalyst based on non-precious metal was developed through an *in situ* polymerization strategy and high-temperature sulfurization treatment. The HAADF image and EDX element distribution showed the synthetic MoS₂ anchored on Ti₃C₂T_x MXene flakes with uniformly dispersed ultrasmall nanodots. Electronic microscopies indicated that the constructed MoS₂/MXene heterostructure displayed excellent stability, preventing the aggregation of both the Ti₃C₂T_x flakes and molybdenum compound nanodots, thus providing high accessibility exposure of active sites for the HER. The Ti₃C₂T_x flakes between MoS₂ and NF have multiple functions, which can protect the NF substrate from the sulfurization reaction, increase the stability of the substrate,



and enhance the conductivity of electron transport. The MoS₂ nanodots sulfurized from MoO₃ achieve an exposure of high-activity edges on the (002) plane, providing better HER performance in an alkaline solution. Compared with NF, MXenes, MoO₃, and MoS₂, the prepared MoS₂/MXene/NF exhibited a Tafel slope of 59 mV dec⁻¹ and an overpotential of 94 mV at 10 mA cm⁻², which is better than those of most of the non-noble metal catalysts in alkaline solution. These findings and insights encourage further heterostructure design and coupling effect investigation of HER catalysts, and point to a new pathway for industrial alkaline catalysis for water splitting.

Author contributions

Xiaoyu Wang: experiments and data collection; Wenbin You: characterization and analysis; Liting Yang, Guanyu Chen, Zhengchen Wu, and Chang Zhang: characterization; Qianjin Chen and Renchao Che: experimental design and project lead. The manuscript was written through contributions of all authors. All authors have given approval to the final version of the manuscript.

Conflicts of interest

The authors declare no conflict of interest.

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References

- Z. W. Seh, J. Kibsgaard, C. F. Dickens, I. B. Chorkendorff, J. K. Nørskov and T. F. Jaramillo, *Science*, 2017, **355**.
- F. Dawood, M. Anda and G. M. Shafiullah, *Int. J. Hydrogen Energy*, 2020, **45**, 3847–3869.
- K. T. Moller, T. R. Jensen, E. Akiba and H.-w. Li, *Prog. Nat. Sci.: Mater. Int.*, 2017, **27**, 34–40.
- I. Roger, M. A. Shipman and M. D. Symes, *Nat. Rev. Chem.*, 2017, **1**, 0003.
- Y. Yan, B. Y. Xia, B. Zhao and X. Wang, *J. Mater. Chem. A*, 2016, **4**, 17587–17603.
- B. You, M. T. Tang, C. Tsai, F. Abild-Pedersen, X. Zheng and H. Li, *Adv. Mater.*, 2019, **31**, 1807001.
- M. Mitov, E. Chorbadzhyska, R. Rashkov and Y. Hubenova, *Int. J. Hydrogen Energy*, 2012, **37**, 16522–16526.
- Y. Zheng, Y. Jiao, A. Vasileff and S.-Z. Qiao, *Angew. Chem., Int. Ed.*, 2018, **57**, 7568–7579.
- S. Ghasemi, S. R. Hosseini and S. Nabipour, *J. Iran. Chem. Soc.*, 2019, **16**, 101–109.
- Z. Wu, Z. Yang, C. Jin, Y. Zhao and R. Che, *ACS Appl. Mater. Interfaces*, 2021, **13**, 5866–5876.
- Q. Ding, B. Song, P. Xu and S. Jin, *Chem*, 2016, **1**, 699–726.
- X. Geng, W. Wu, N. Li, W. Sun, J. Armstrong, A. Al-hilo, M. Brozak, J. Cui and T.-p. Chen, *Adv. Funct. Mater.*, 2014, **24**, 6123–6129.
- M. I. A. A. Maksoud, A. G. Bedir, M. Bekhit, M. M. Abouelela, R. A. Fahim, A. S. Awed, S. Y. Attia, S. M. Kassem, M. Abd Elkodous, G. S. El-Sayyad, S. G. Mohamed, A. I. Osman, A. a. H. Al-Muhtaseb and D. W. Rooney, *Environ. Chem. Lett.*, 2021, **19**, 3645–3681.
- Y. Yang, H. Yao, Z. Yu, S. M. Islam, H. He, M. Yuan, Y. Yue, K. Xu, W. Hao, G. Sun, H. Li, S. Ma, P. Zapol and M. G. Kanatzidis, *J. Am. Chem. Soc.*, 2019, **141**, 10417–10430.
- H. Wang, X. Xiao, S. Liu, C.-L. Chiang, X. Kuai, C.-K. Peng, Y.-C. Lin, X. Meng, J. Zhao, J. Choi, Y.-G. Lin, J.-M. Lee and L. Gao, *J. Am. Chem. Soc.*, 2019, **141**, 18578–18584.
- J. Hu, C. Zhang, L. Jiang, H. Lin, Y. An, D. Zhou, M. K. H. Leung and S. Yang, *Joule*, 2017, **1**, 383–393.
- T. F. Jaramillo, K. P. Jørgensen, J. Bonde, J. H. Nielsen, S. Hørch and I. Chorkendorff, *Science*, 2007, **317**, 100–102.
- B. Hinnemann, P. G. Moses, J. Bonde, K. P. Jørgensen, J. H. Nielsen, S. Hørch, I. Chorkendorff and J. K. Nørskov, *J. Am. Chem. Soc.*, 2005, **127**, 5308–5309.
- D. Voiry, R. Fullon, J. Yang, C. d. C. Castro e Silva, R. Koppera, I. Bozkurt, D. Kaplan, M. J. Lagos, P. E. Batson, G. Gupta, A. D. Mohite, L. Dong, D. Er, V. B. Shenoy, T. Asefa and M. Chhowalla, *Nat. Mater.*, 2016, **15**, 1003–1009.
- B. Zhang, J. Liu, J. Wang, Y. Ruan, X. Ji, K. Xu, C. Chen, H. Wan, L. Miao and J. Jiang, *Nano Energy*, 2017, **37**, 74–80.
- G. Zhao, P. Li, K. Rui, Y. Chen, S. X. Dou and W. Sun, *Chem.–Eur. J.*, 2018, **24**, 11158–11165.
- G. Ou, P. Fan, X. Ke, Y. Xu, K. Huang, H. Wei, W. Yu, H. Zhang, M. Zhong, H. Wu and Y. Li, *Nano Res.*, 2018, **11**, 751–761.
- J. Zhang, T. Wang, P. Liu, S. Liu, R. Dong, X. Zhuang, M. Chen and X. Feng, *Energy Environ. Sci.*, 2016, **9**, 2789–2793.
- S. Xu, D. Li and P. Wu, *Adv. Funct. Mater.*, 2015, **25**, 1127–1136.
- S. Anantharaj, S. Noda, V. R. Jothi, S. Yi, M. Driess and P. W. Menezes, *Angew. Chem., Int. Ed.*, 2021, **60**, 18981–19006.
- X. Wang, Y. Zheng, W. Sheng, Z. J. Xu, M. Jaroniec and S.-Z. Qiao, *Mater. Today*, 2020, **36**, 125–138.
- J. Lin, P. Wang, H. Wang, C. Li, X. Si, J. Qi, J. Cao, Z. Zhong, W. Fei and J. Feng, *Adv. Sci.*, 2019, **6**, 1900246.
- X. Li, W. You, L. Wang, J. Liu, Z. Wu, K. Pei, Y. Li and R. Che, *ACS Appl. Mater. Interfaces*, 2019, **11**, 44536–44544.
- J. Zhang, T. Wang, D. Pohl, B. Rellinghaus, R. Dong, S. Liu, X. Zhuang and X. Feng, *Angew. Chem., Int. Ed.*, 2016, **55**, 6702–6707.
- H. Huang, J. Cui, G. Liu, R. Bi and L. Zhang, *ACS Nano*, 2019, **13**, 3448–3456.



- 31 Y. Wei, R. A. Soomro, X. Xie and B. Xu, *J. Energy Chem.*, 2021, **55**, 244–255.
- 32 H. Wang, Y. Lin, S. Liu, J. Li, L. Bu, J. Chen, X. Xiao, J.-H. Choi, L. Gao and J.-M. Lee, *J. Mater. Chem. A*, 2020, **8**, 7109–7116.
- 33 X. Li, X. Lv, X. Sun, C. Yang, Y.-Z. Zheng, L. Yang, S. Li and X. Tao, *Appl. Catal., B*, 2021, **284**, 119708.
- 34 J. Liu, Y. Liu, D. Xu, Y. Zhu, W. Peng, Y. Li, F. Zhang and X. Fan, *Appl. Catal., B*, 2019, **241**, 89–94.
- 35 W. Yuan, L. Cheng, Y. An, H. Wu, N. Yao, X. Fan and X. Guo, *ACS Sustainable Chem. Eng.*, 2018, **6**, 8976–8982.
- 36 X. Zhan, C. Si, J. Zhou and Z. Sun, *Nanoscale Horiz.*, 2020, **5**, 235–258.
- 37 S. Seyedin, E. R. S. Yanza and J. M. Razal, *J. Mater. Chem. A*, 2017, **5**, 24076–24082.
- 38 C. G. Morales-Guio, L. Liardet, M. T. Mayer, S. D. Tilley, M. Graetzel and X. Hu, *Angew. Chem., Int. Ed.*, 2015, **54**, 664–667.
- 39 B. He, L. Chen, M. Jing, M. Zhou, Z. Hou and X. Chen, *Electrochim. Acta*, 2018, **283**, 357–365.
- 40 H. F. Liu, S. L. Wong and D. Z. Chi, *Chem. Vap. Deposition*, 2015, **21**, 241–259.
- 41 Q. Hu, Z. Wang, X. Huang, Y. Qin, H. Yang, X. Ren, Q. Zhang, J. Liu and C. He, *Energy Environ. Sci.*, 2020, **13**, 5097–5103.
- 42 X. Liu, Q. Hu, B. Zhu, G. Li, L. Fan, X. Chai, Q. Zhang, J. Liu and C. He, *Small*, 2018, **14**, 1802755.

