1. Introduction

Lithium-ion batteries (LIBs) and sodium-ion batteries (SIBs) are under rapid development as promising energy storage devices.[1] For anodes, alloying-type (e.g., Si, Sn, Al) and conversion-type materials (e.g., MoO2, MoS2, Co3O4, and MnO) are attractive because they are able to deliver higher capacities than intercalation materials (e.g., graphite and Li4Ti5O12).[2] However, they often suffer from huge volume changes during charge/discharge and have poor cycling stability. A practical way to solve this problem is via hierarchical architecture, forming core–shell, yolk–shell, hollow–sphere structures,[3] or decorating nanoparticles onto stable three-dimensional (3D)/two-dimensional (2D)/ one-dimensional (1D) framework.[4] Such hierarchical architectures can benefit from the components at each length scale: large surface area, short diffusion distance, and fast kinetics from nanostructures; good wetting and open pore channels from meso-/microporous structures; and a stable 3D network that can buffer large volume change and suppress decorated particles from agglomeration/pulverization during cycling.[3a,4]

The most commonly used 3D/2D/1D frameworks are carbon (e.g., nanotubes, graphene, and amorphous carbon), conductive polymers,[5] and Ti-based oxides (TiO2 polymorphs, lithium titanates, and hydrated lithium titanates),[6] because they experience little volume change during electrochemical charge/discharge and have superior stability over the applicable voltage window. The above fact is also consistent with the properties of intercalation compounds: no significant structural change upon lithiation/delithiation. In comparison, alloying-/conversion-type electrodes have never been considered as good candidates for the framework, because of their poor cycling stability. However, their stability can be greatly improved if one limits the cut-off voltage. For example, there is much less capacity fading in silicon/carbon composite if one limits the voltage window to 0.17–0.60 V versus Li/Li+ compared to using 0.05–1.00 V.[7] On the other hand, although not well recognized, switching the application from LIBs to SIBs sometimes helps improve the cycling life of alloying-/conversion-type anodes (meanwhile, capacity is often lowered). For example, MoS2 anode is more stable when used in SIBs (>250 cycles) and less in LIBs (<150 cycles) under same current density of 1 A g–1 (specific capacity in SIBs is half of that in LIBs).[8] With the above two considerations and as a proof of concept, in the present study we show successful application of conversion-type MnO nanorods as a stable anode framework hybridized with high-capacity molybdenum oxide and carbide nanoparticles, which can stably deliver 431 mAh g–1 at 1000 mA g–1 (=2 C) over 600 cycles (43% capacity retention) in LIBs and more surprisingly, 110 mAh g–1 at 8000 mA g–1 (=70 C) over 40 000 cycles (=100% capacity retention) in SIBs. Previously, MnO was not considered for such applications because in LIBs...
it has unsatisfactory stability, while in SIBs it has low capacity (less than 200 mAh g$^{-1}$) and its excellent stability has not been well recognized despite a few reports in the literature.[9]

2. Results and Discussion

The synthesis is as follows (see schematics in Figure 1 and more details in Experimental Section). 1) MnO$\gamma$ nanorods were first synthesized by hydrothermal method at 160 °C for 12 h. 2) MnO$\gamma$ nanorods were then mixed with ammonium molybdate and dopamine hydrochloride in water/alcohol solution, followed by polymerization using ammonia. The as-obtained precursors (denoted as MnO$\gamma$@Mo-PDA hereafter) were washed, collected by centrifuge, and dried. 3) The precursors were heat-treated in flowing argon atmosphere at 800 °C for 3 h. The final product (denoted as MnO@MOC/C hereafter) is a composite of MnO nanorods, decorated with carbon nanosheets and molybdenum oxide/carbide nanoparticles (denoted as MOC/C hereafter) by repeating steps (2, 3) without adding MnO$\gamma$ nanorods and b) MnO nanorods (denoted as MnO hereafter) by repeating steps (1) and (3) without step (2).

The MnO$\gamma$ nanorods synthesize in step (1) are about 100 nm in width and several microns in length (Figure S1, Supporting Information). After coating step, the precursor MnO$\gamma$@Mo-PDA show uniform decorations with a porous structure (Figure S2, Supporting Information). After heat treatment in argon, the final product MnO@MOC/C shows a porous hierarchical structure (Figure 2a,b, more images in Figures S3 and S4, Supporting Information; for specific surface area, see Brunauer–Emmett–Teller (BET) data in Table S1 and Figure S5, Supporting Information), with MnO nanorods (diameter 50–200 nm) coated by ≈10 nm thick carbon nanosheets. Energy-dispersive X-ray spectroscopy (EDS) mappings in Figure 2e–h confirm uniform distribution of Mn, Mo, and O (also see EDS linear scan analysis in Figure S6, Supporting Information). Under high-resolution transmission electron microscopy (HRTEM; Figure 2c), MnO phase is confirmed with 0.26 nm interlayer spacing (corresponding to (111) crystal plane of MnO,

Figure 1. Schematics for material synthesis of MnO@MOC/C hybrid.

Figure 2. Morphology of MnO@MOC/C hybrid, a) SEM, b,c) HRTEM, d) STEM (inset: SAED analysis of the 2D carbon area). e) HRTEM images and EDS mapping of f) Mn, g) Mo, and h) O elements.
According to JCPDS No. 07-0230). The Mo-containing species are more difficult to characterize directly, probably due to ultrafine sizes and good dispersion in carbon matrix. Selected area electron diffraction (SAED) was therefore conducted in the coating layer (not on MnO nanorods). As shown in the inset of Figure 2d, the diffraction patterns circled in orange are associated with (102) lattice plane of Mo2C (JCPDS No. 71-0242) and dispersed diffraction rings from \( \text{Mn}_2\text{O}_3 \) (JCPDS No. 24-0508). (The formation of \( \text{Mn}_2\text{O}_3 \) could be due to partial oxidation of \( \text{MnO} \).) Third, Raman analysis of Mo2C (JCPDS No. 71-0242) and at around 33° (associated with \( \text{Mo}(\text{VI})\text{O}_2\text{C} \) and \( \text{Mo}(\text{VI})\text{O}_2\text{C} \)) represent \( \text{Mo}–\text{O} \) bond in the form of \( \text{Mo}_2\text{C} \), and two peaks at 235.8 and 229.7 eV (associated with \( \text{Mo}(\text{IV})\text{O}_2\text{C} \) and \( \text{Mo}(\text{IV})\text{O}_2\text{C} \)) represent \( \text{Mo}–\text{O} \) bond in the form of \( \text{Mo}_2\text{C} \). For C (Figure 3e,f), bonding states of C−C−O, C=O, and C−Mo are observed. An interesting observation is the absence of Mn peaks in XPS (absence of Mn 2p signals at 630–660 eV in Figure 3c). Since XPS is a surface-sensitive (a few nm) technique, it indicates full decoration of carbon and molybdenum oxides/carbides on MnO nanorods, which blocks injected electrons from MnO in XPS. Last, we conclude the chemical nature of Mo-containing species from the above characterizations. XRD only detected minor \( \text{Mo}_2\text{C} \) phase, much less than MnO phase. But ICP-MS data suggest a Mn/Mo mole ratio of 1:10. Therefore, most Mo-containing species should be amorphous or very nano in size, which cannot be easily observed by XRD or transmission electron microscopy (TEM). This is also consistent with uniform Mo distribution in EDS mapping in Figure 2g. Meanwhile, SAED in Figure 2d show clear diffraction pattern for \( \text{Mo}_2\text{C} \), but only weak dispersed diffraction rings for \( \text{Mo}_2\text{C} \), like species. This agrees with the XRD data, indicating \( \text{Mo}_2\text{C} \) should be the main Mo-containing phase, rather than \( \text{Mo}_2\text{O}_4 \). On the other hand, Raman and XPS analysis found stronger Mo–O signals than Mo–C at the surface, which could be due to partial oxidation of \( \text{Mo}_2\text{C} \) at the surface and is consistent with literature reports.

**Figure 3.** Characterizations of MnO@MOC/C, a) XRD, b) Raman, and c) XPS spectrum. Inset of (b): high frequency Raman data for C–C vibrations. Also shown are fitting results for d) Mo 3d, e) C 1s, and f) O 1s peaks from XPS.
for molybdenum carbides (e.g., Mo$_2$C and MoC, whose surface can be easily oxidized).

The synthesized MnO@MOC/C and two control samples (MOC/C and MnO) were tested in a half-cell configuration using Na metal as the reference and counter electrodes for SIBs. (Detailed morphology and structural analysis for the three control samples are listed in Figures S8–S10, Supporting Information). MnO@MOC/C and MOC/C both show good rate performance, better than typical carbonaceous and Ti-based anodes for SIBs.[18] When the current density increases from 200 to 8000 mA g$^{-1}$, the discharge specific capacities of MnO@MOC/C decreased from 248.5 to 95.2 mAh g$^{-1}$ (Figure 4a; note that the anomalously high discharge capacity at the first cycle is due to the formation of solid electrolyte interphase, SEI), and MOC/C's decreased from 273.7 to 124.4 mAh g$^{-1}$ (Figure S11, Supporting Information). However, the capacities of MnO as well as carbon coated MnO nanorods (denoted as MnO/C hereafter) are unsatisfactory even at relatively low current density (46.9 mAh g$^{-1}$ for MnO and 81.1 mAh g$^{-1}$ for MnO/C at 200 mA g$^{-1}$), indicating that neither MnO nor carbon are good electrode materials for Na$^+$ storage in the tested voltage range (0.01 to 3.00 V vs Na/Na$^+$, Figure S12, Supporting Information). If we estimate the expected value of MnO@MOC/C's capacity from a linear combination of MOC/C's and MnO's, the estimated values are actually similar as the experimental ones we obtained in Figure 4a, especially at small rates (see Table S2, Supporting Information for details). This demonstrates a synergetic effect of all three components (MnO nanorods, carbon nanosheets, and Mo$_2$C/MoO$_x$ active materials). MnO@MOC/C further shows stable cycling with 110 mAh g$^{-1}$ discharge capacity for 40 000 cycles at 8000 mA g$^{-1}$ ($\approx$70 C), with little capacity decay and close to 100% Coulombic efficiency (CE, red curve in Figure 4b). In comparison, even though the control samples MnO (black curve in Figure 4b) and MnO/C also show stable cycling, their capacities are too low to be considered useful (consistent with literature reports[9]), which is not useful in any sense. This is apparently the reason why pure MnO had never been considered as a good candidate in SIBs. On the other hand, MOC/C suffers from poor cyclability, whose discharge capacity decreases from 159 to $\approx$0 mAh g$^{-1}$

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**Figure 4.** Electrochemical analysis of MnO@MOC/C hybrid in SIBs. a) Rate and b) cycling performances at 8000 mA g$^{-1}$ of MnO@MOC/C, MOC/C, and MnO electrodes. c) TEM of MnO@MOC/C after 40 000 cycles. d) Galvanostatic discharge/charge profiles from 100 to 2000 mA g$^{-1}$ of full-cell using MnO@MOC/C as the anode and Na$_3$V$_2$(PO$_4$)$_3$ as the cathode in SIBs. e) Schematic diagram of the fast Na$^+$ transfer behaviors of MnO@MOC/C in SIBs.
after 3500 cycles at 8000 mA g$^{-1}$ (blue curve in Figure 4b). The carbon content of MOC/C is estimated to be 42 wt% by TGA (Figure S7b, Supporting Information), which is higher than that of MnO@MOC/C (16 wt%). So carbon alone cannot completely explain the superior cycling stability of MnO@MOC/C and the poor cycling stability of MOC/C highlights the critical role of MnO nanorods, which provides a rigid framework to suppress agglomeration and coarsening of carbon nanosheets and Mo$_2$C/MoO$_{x}$ nanoparticles. To confirm the microstructural stability of MnO@MOC/C, the electrode materials after 40 000 cycles were investigated under TEM and showed well maintained morphology (of carbon-decorated MnO nanorods, in Figure 4c). Therefore, the superior electrochemical and microstructural stability demonstrates another synergetic effect from the design of such a hierarchical architecture: MnO as a rigid framework, Mo$_2$C/MoO$_{x}$ offering high capacity and probably carbon buffering stress/strain.

As a further proof of the superior electrochemical properties, we assembled full cells using MnO@MOC/C as the anode and Na$_3$V$_2$(PO$_4$)$_3$ as the cathode. Figure 4d illustrates the voltage-to-capacity curves of the full cell at various current densities. With the increasing of current densities from 100 to 2000 mA g$^{-1}$ (for each eight cycles), the discharge specific capacity decreased from 204.8 to 48.9 mAh g$^{-1}$. Due to the irreversible reactions in both cathode and anode together with limited sodium ions in full cells compared with massive Na$^+$ in half cells, the CE is not as good as that in half cells, especially for the first few cycles (only 60%). However, the CE increases to >90% after 20 cycles. After the rate performance tests, the electrochemical reactions became reversible and showed stable cycling at relatively high current density of 500 mA g$^{-1}$ for the next 150 cycles with capacity retention of >60% (Figure S13, Supporting Information). The results demonstrate the rapid electrochemical kinetics of MnO@MOC/C as an anode for SIBs (Figure 4e), but also emphasize the significance of CE in the practical full cells. More efforts should be made to decrease the side reactions in both anodes and cathodes to meet the requirements of real-life SIBs applications.

To understand the reaction mechanism better, we conducted cyclic voltammetry (CV) measurements (from 0.01 to 3.00 V vs Na/Na$^+$ at a scan rate of 0.1 mV s$^{-1}$; results shown in Figure 5a) for MnO@MOC/C. In the first cycle of discharge (from 3.00 to 0.01 V with a negative current), much cathodic reaction takes place below 1.6 V. This is not seen in subsequent cycles and is attributed to the formation of SEI. After the first cycle of discharge, CV curves are mostly unchanged, suggesting a stable redox behavior. Interestingly, the CV curves are flat and diffusive, except small redox peaks at $\approx$1.6 V during discharge (probably MoO$_x$ reduction to Na$_y$MoO$_x$), $\approx$1.7 V during charge (probably Na$_y$MoO$_x$ oxidation to MoO$_x$), and at $\approx$2.3 V during charge (probably Mn oxidation). Such diffusive CV curves could come from pseudocapacitive behavior, or defects/disorders (i.e., a distribution of redox potentials due to a large amount of surface/lattice defects). It is further supported by analyzing peak current density $j_p$ versus scan rate $v$ in a log–log plot (Figure S14, Supporting Information), giving a $b$ value of 0.99 for MnO@MOC/C ($b$ value is 1.0 for purely surface-controlled process, i.e., a capacitor-like behavior; and 0.5 for purely diffusion-controlled process, i.e., a battery-like behavior). In comparison, $b$ value of control samples is 1.0 for MOC/C and
0.45 for MnO (Figure 5b). The transition from more diffusion-controlled process to more surface-controlled one may improve the cycling stability of conversion-type Mo2C/MoOx—which contributes most capacity in MnO@MOC/C—especially at high rates. This effect is known in LiFePO4 as cathode in LIBs: nano-LiFePO4 at slow charge/discharge rates or micro LiFePO4 show two-phase behavior and are prone to degradation, while nano-LiFePO4 at fast rates show solid-solution behavior and have excellent cycling stability.[23]

The above $b$-value analysis also suggests diffusion is likely less a problem in our MnO@MOC/C, which offers much better kinetics and hence rate performance than bulk Mo2C/MoOx.[24] To further demonstrate this, we conducted galvanostatic intermittent titration technique (GITT, Figure 5c) and calculated the apparent diffusion coefficient $D_{Na}$ according to[25]

$$D_{Na} = \frac{4}{\pi} \left(\frac{V_{Na}}{AFZ}\right) \left[t_0 \left(\frac{dE}{d\delta}\right) \left(\frac{dE}{dI}\right)\right]^{-1}$$

(1)

where $V_{Na}$ is molar volume of active material (here we use 22 cm$^3$ mol$^{-1}$ of MoC for simplification). $A$ is total contact area between electrolyte and electrodes, $F$ is Faraday constant, $I_0$ is applied constant electric current, $Z_i$ is valence of species, $dE/d\delta$ is determined from initial stoichiometry, $L$ is the thickness of electrode, $E$ is electrode voltage, $\delta$ is the deviation from the stoichiometric ratio, and $t$ is the time during constant current pulse. The obtained apparent diffusivity coefficient $D_{Na}$ (Figure 5d) is surprisingly high, in the range of $10^{-12}$–$10^{-10}$ cm$^2$ s$^{-1}$, which are higher than typical high-rate LIB electrodes such as LiFePO4 ($10^{-11}$–$10^{-10}$ cm$^2$ s$^{-1}$)[26] and Li4Ti5O12 ($10^{-10}$–$10^{-9}$ cm$^2$ s$^{-1}$).[27] and comparable to typical high-rate SIB electrode P2-Na2/3[Ni1/3Mn2/3]O2 ($10^{-10}$–$10^{-9}$ cm$^2$ s$^{-1}$).[28] (The above diffusivity data are all calculated from GITT measurements.) A closer inspection on the GITT curve and $D_{Na}$ also suggests a relatively poor kinetics at the end of charge (2.0–3.0 V vs Na/Na$^+$, Figure S15, Supporting Information). It may be because higher-valence electronically-insulating Mo$^{6+}$ forms at such voltages, imposing a kinetic issue for electron transport. This problem may be solved by transitional metal doping (e.g., Co–N doping)[29] of as-synthesized MoC/MoOx nanoparticles, which should improve our materials further. Also to note in the GITT data is that the overpotential of MnO@MOC/C increases at higher voltage in the desodiation process (similar trend also holds in LIBs; see Figure S16, Supporting Information). This could be due to two reasons. First, at higher voltages, Mo ions are oxidized to higher valence so the electronic conductivity may become worse, which increases the overpotential. Second, ionic diffusion in MoO2 and Mo2C should follow an interstitial mechanism. So at higher voltages, the concentration of charge carriers (lithium interstitial) becomes lower, which also increases the overpotential.

The three electrode materials (MnO@MOC/C, MOC/C, and MnO) were further tested in LIBs, in a half-cell configuration using Li metal as the reference and counter electrode. The voltage-to-capacity curves at the current density of 50 mA g$^{-1}$ are shown in Figure 6a, with 888.3 mAh g$^{-1}$ for MnO@MOC/C, 992.4 mAh g$^{-1}$ for MOC/C, and 679.3 mAh g$^{-1}$ for MnO. Compared with the smooth curves of MnO@MOC/C and MOC/C with apparent plateaus, MnO showed a plateau at ~0.6 V during lithiation, indicating a Faradaic two-phase transformation process. MnO@MOC/C has a good rate performance, with discharge capacity decreased from 888.3 to 508.3 mAh g$^{-1}$ when the current density increased from 50 to 2000 mA g$^{-1}$ (Figure 6b; Figure S17, Supporting Information). In comparison, the specific capacity of the MOC/C decreased much faster from 985.8 to 354.7 mAh g$^{-1}$ (MOC/C) and MnO’s decreased from 643.8 to 375.6 mAh g$^{-1}$ when the current density increased from 50 to 2000 mA g$^{-1}$. The capacity and rate performance of MnO@MOC/C in LIBs is highly competitive with MnO-, Mo$_2$C- or MoO$_2$-based anodes reported in literature references (Table S3, Supporting Information). However, few of MnO@MOC/C, MOC/C, and MnO show good stability in prolonged cycling of >1300 cycles (Figure 6c) in LIBs, which is different from the stable cycling of MnO@MOC/C in SIBs. (See comparison with recently reported MnO-, Mo$_2$C- and MoO$_2$-based anodes in SIBs references in Table S4, Supporting Information.) The reason could be due to much larger capacity of MnO in LIBs (as shown in Figures 4 and 6), resulting in large volume change and serious pulverization of the hybrid materials. TEM of MnO@MOC/C after 500 cycles in LIBs (Figure S18, Supporting Information) also shows thicker SEI than that after 40 000 cycles in SIBs (see Figure 4c for comparison). Therefore, limiting the capacity (hence less volume change and higher stability) of conversion-type anodes is important when using them as the framework in hybrid architectures.

![Figure 6](https://www.afm-journal.de/)

Figure 6. Electrochemical analysis of MnO@MOC/C hybrid in LIBs. a) Galvanostatic discharge/charge profiles at 50 mA g$^{-1}$, b) rate performances, and c) cycling performances at 500 mA g$^{-1}$ of MnO@MOC/C, MOC/C, and MnO electrodes.
3. Conclusions

To summarize, we designed and synthesized 1D@2D/0D hierarchical architectures with 1D MnO nanorods, 2D carbon nanosheets, and 0D Mo2C/MoOx nanoparticles. Our work demonstrated successful application of conversion-type MnO nanorod as a super-stable anode framework in SIBs. It can stably deliver \( \approx 110 \text{ mAh g}^{-1} \) at \( \approx 70 \text{ C} \) over 40 000 cycles with no apparent capacity decay. Our work suggests that conversion reaction is not inherently related to fast degradation. If one limits the capacity in use (e.g., switching the application of MnO from LIBs to SIBs as in the present work), conversion-type materials may also be potentially useful as a stable anode framework. We believe this to be a general strategy applicable to many conversion-/alloying-type electrode materials and it provides new opportunities to explore the synthesis and to tune the interfacial/bonding characteristics between the host framework and the loaded high-capacity nanomaterials.

4. Experimental Section

Materials Synthesis—Synthesis of MnO nanorods: 90 mg KMnO4 and 27 mg of polyvinylpyrrolidone were dissolved in 30 mL of deionized (DI) water under vigorous magnetic for 10 min, forming a homogeneous solution in purple. The solution was then transferred into a 50 mL Teflon-lined stainless-steel autoclave, sealed, and maintained at 170 °C for 12 h for a hydrothermal reaction. After hydrothermal reaction, the precipitates were collected and washed with DI water and ethanol for several times.

Synthesis of MnO@MOC/C, MOC/C, MnO, and MnO/C: For the synthesis of MnO@MOC/C, 8 mg as-synthesized MnO nanorods were dispersed into 12 mL deionized water with ultrasonic treatment to form a suspension, followed by addition of 24 mg ammonium molybdate ((NH4)6Mo7O24·4H2O), 8 mg dopamine hydrochloride, and 8 mL ethanol under intense stirring. 20 µL ammonium hydroxide (25–28 wt% ammonia in water) was next added into the above suspension, followed by intense stirring for 6 h. The collected materials were dried and heated at 800 °C in flowing argon for 3 h with a heating rate of 5 °C min\(^{-1}\) to obtain the final product MnO@MOC/C. For the synthesis of reference sample MOC/C, 24 mg ammonium molybdate and 8 mg dopamine hydrochloride were added into 8 mL ethanol under intense stirring. 20 µL ammonium hydroxide (25–28 wt% ammonia in water) was next added into the above suspension, followed by intense stirring for 6 h. The collected materials were dried and heated at 800 °C in flowing argon for 3 h with a heating rate of 5 °C min\(^{-1}\) to obtain the final product MOC/C. For the synthesis of reference sample MnO nanorods, the as-synthesized MnO nanorods were heated at 800 °C in flowing argon for 3 h with a heating rate of 5 °C min\(^{-1}\) to obtain the final product MnO. For the synthesis of reference sample carbon coated MnO nanorods MnO/C, 8 mg as-synthesized MnO nanorods were dispersed into 12 mL deionized water with ultrasonic treatment to form a suspension, followed by addition of 8 mg dopamine hydrochloride and 8 mL ethanol under intense stirring. 20 µL ammonium hydroxide (25–28 wt% ammonia in water) was next added into the above suspension, followed by intense stirring for 6 h. The collected materials were dried and heated at 800 °C in flowing argon for 3 h with a heating rate of 5 °C min\(^{-1}\) to obtain the final product MnO/C.

Characterizations: Scanning electron microscope (SEM, MERLIN VP Compact), transmission electron microscopy (TEM, Hitachi-H7700), HRTEM (JEM-2100F), and scanning TEM (STEM, JEM-2100F) were used to analyze the morphology and structures. X-ray diffraction (XRD, Rigaku D/Max-B X; Cu Kr radiation \( \lambda = 1.5418 \text{ Å} \)) was used to identify the phase. Raman spectrometer (HR800, HORIBA; 633 nm line of helium-neon beam) was used for Raman spectrum analysis. XPS (Escalab 250Xi system, Thermo Fisher Scientific) was used to analyze surface chemistry. Transition metal ratios were obtained from inductively coupled plasma mass spectroscopy (ICP-MS, Thermo Fisher Scientific). TGA analysis was conducted on thermogravimetric analyzer (STA 449F3) instrument in flowing air at 5 °C min\(^{-1}\) ramping rate from 50 to 600 °C. BET measurements (Autosorb-iQ2-MS, Quanta Chrome) were used to calculate specific surface areas.

Electrochemical Measurements: To prepare the electrode composite, 70 wt% active materials, 20 wt% Super P, and 10 wt% poly(vinylidene fluoride) binder were mixed in N-methyl-2-pyrrolidinone. The obtained slurry was casted on Cu (for MnO@MOC/C, MOC/C, MnO/C, and MnO anodes) or Al foil (for Na\(_3\)V\(_2\)(PO\(_4\))\(_3\) cathodes) and dried at 110 °C under vacuum overnight. For half-cell tests, sodium and lithium metal were used as the counter and reference electrodes in SIBs and LIBs, respectively. For full-cell tests in SIBs, Na\(_3\)V\(_2\)(PO\(_4\))\(_3\) was used as the cathode and MnO@MOC/C was used as the anode. Glass fibers (Whatman Inc.) and microporous membranes (Celgard 2400) were used as separators in SIBs and LIBs, respectively. Ethylene carbonate-dimethyl carbonate (1:1 in volume) with 1.0 M NaClO\(_4\) and 0.5 wt% fluoroethylene carbonate additive and ethylene carbonate-dimethyl carbonate (1:1 in volume) with 1.0 M LiPF\(_6\) were used as electrolytes in SIBs and LIBs, respectively. Mass loading of all anode samples is about 1.0 mg cm\(^{-2}\). Mass loading of Na\(_3\)V\(_2\)(PO\(_4\))\(_3\) cathodes is about 4.5 mg cm\(^{-2}\), providing >50% excess area-capacity compared with MnO@MOC/C anodes. Cyclic voltammetry were performed using IM6 (Bas-Zahnner) electrochemical workstation. GITT measurements were conducted at 50 mA g\(^{-1}\) with a duration of 40 min, followed by 20 h relaxation. GITT data were obtained in cell coins after five cycles at 100 mA g\(^{-1}\). All half/full-cell tests were conducted using LAND 2001A Cell test system between 0.01–3.00 V (for half cells) and 0.01–4.00 V (for full cells) at ambient temperature.

Supporting Information

Supporting Information is available from the Wiley Online Library or from the author.

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Conflict of Interest

The authors declare no conflict of interest.

Keywords

lithium-ion batteries, MnO, molybdenum, nanomaterials, sodium-ion batteries

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