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Introduction

Currently, lithium-ion batteries (LIBs) are widely used in various energy storage systems such as mobile phones, computers, and power grids.^{1,2} However, the growth of LIBs has encountered limitations³ such as the slow increase in energy density and the safety problem, but the most serious reason is the lack of lithium resources.⁴ Based on the sustainable development strategy consideration, secondary batteries based on inexpensive metal ions⁵ such as sodium-ion batteries (SIBs),⁶ potassium-ion batteries (KIBs),⁷ magnesium-ion batteries (MIBs)⁸ and aluminum-ion batteries (AIBs)⁹ have received increased attention. Among these battery systems, AIBs have attracted more and more attention owing to their unique advantages.

Aluminum is the most abundant metal element on the earth and is widely distributed. Moreover, aluminum metal is also a kind of energy carrier with high specific capacity due to its three-electron transfer nature. Its volumetric specific capacity reaches 8046 mA h cm⁻³, which is about three times higher than that of lithium (2062 mA h g⁻¹);¹⁰ hence, rechargeable aluminum-ion secondary batteries have a broader pro-



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Aluminum-ion batteries (AIBs) are supposed to be one of the energy storage systems with great potentialities on account of their high safety, low cost and high theoretical volumetric capacity. Herein, we report a novel rod-shaped $Cu_{1.81}$ Te cathode material for AIBs. At 40 mA g⁻¹, the initial discharge capacity can reach 144 mA h g⁻¹. The diffusion coefficient of Al³⁺ calculated by the galvanostatic intermittent titration technique (GITT) and cyclic voltammetry (CV) tests at different scan rates is larger than that in sulfides, indicating that telluride has faster kinetics. The results of *ex situ* X-ray photoelectron spectroscopy (XPS), *ex situ* X-ray diffraction (XRD) and ²⁷Al nuclear magnetic resonance (NMR) prove that the mechanism of the charging and discharging processes is the reversible intercalation and deintercalation of Al³⁺, which is very important for the subsequent researchers to understand and investigate the mechanism of the Al/Cu_{1.81}Te battery. This work also proves that telluride can also be used as a cathode material for aluminum storage.

> spect.¹¹ Since AIBs can directly use aluminum metal as the anode, the main factor limiting the energy density of the system is the cathode material. However, there are few materials that can reversibly conduct the insertion and extraction of Al³⁺ due to the slow diffusion rate of Al³⁺, and the performance is not as good as that of the lithium-ion batteries. Dai et al.¹² developed an AIB system in 2015 with a threedimensional graphitic foam as the cathode and an ionic liquid (IL) as the electrolyte, which could be stably cycled over 7500 cycles at 4000 mA g^{-1} with a coulombic efficiency (CE) of 97%. This is the first AIB prototype with good reversibility. Since then, various new AIB cathode materials have been reported, like graphite materials, $^{13-16}$ transition metal oxides (V₂O₅, $^{17-20}$ TiO_2 ,^{21,22} SnO₂@C,²³ and Li₃VO₄²⁴), conductive polymer materials,^{25,26} and transition metal chalcogenides (CuS@graphene,²⁷ SnS₂@RGO,²⁸ and CoSe₂²⁹). Among them, the capacity of carbon-based materials is limited to about 60 mA h g⁻¹ due to the insertion/extraction of monovalent AlCl₄⁻ ions rather than trivalent Al³⁺ ions. For transition metal oxides, the oxygen-ion framework is not an ideal host framework owing to the strong coulombic effect of trivalent Al^{3+} . In contrast, a transition metal chalcogenide material is a more promising cathode material because of its higher capacity and lower polarity. In 2016, Wang et al.30 prepared a cathode material of Ni₃S₂@graphene for AIBs. The initial capacity reached 350 mA h g^{-1} with the discharge platform up to 1.0 V at 100 mA g⁻¹. In 2018, Hu et al.³¹ designed a cathode material of cobalt sulfide@carbon nanotubes for AIBs, which showed extremely high initial capacity (315 mA h g^{-1}) and remarkable



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Paper

cycling performance; after 200 cycles, it maintained 297 mA h g^{-1} . Our group previously reported nonstoichiometric $Cu_{2-x}Se$ nanorods³² for AIBs. The unique crystal structure resulted in electronic conductivity of about 3000 times that of stoichiometric Cu_2S .³³ The initial capacity reached 277 mA h g^{-1} and the batteries could be stably cycled over 200 times. Although transition metal sulfides and selenides have been studied in AIBs, to the best of our knowledge, transition metal tellurides have not been reported as the cathode materials in AIBs. Finding new materials suitable as the cathode materials for AIBs and exploring the aluminum storage mechanism are also crucial for the development of AIBs.

Compared to metal sulfides or selenides, metal tellurides tend to have a larger unit cell volume to accommodate ion embedding and have a smaller volume effect due to Te having a larger ionic radius.³⁴ Meanwhile, owing to the weaker electronegativity of Te, the interaction with Al³⁺ is weaker, which is more conducive to the diffusion of Al^{3+,35} Moreover, the electronic conductivity of telluride tends to be higher, which is more favorable for electrochemical performance. Although telluride has been initially studied in LIBs,^{34,36–40} it has not been reported in AIBs. Considering the advantages of telluride, it will make sense to study its aluminum storage properties.

On this basis, herein, we first reported rod-shaped $Cu_{1.81}$ Te prepared by a simple hydrothermal method as the cathode material of AIBs; this is also the first time that a transition metal telluride material has been used as the cathode for AIBs. Our results showed that $Cu_{1.81}$ Te exhibited a good electrochemical performance; X-ray photoelectron spectroscopy (XPS), X-ray diffraction (XRD), ²⁷Al nuclear magnetic resonance (NMR) and transmission electron microscopy (TEM) were used to prove that the mechanism of the charging and discharging processes is the reversible intercalation and deintercalation of Al³⁺.

Experimental section

Preparation of Cu_{1.81}Te nanorods

According to a previous report,^{41,63} the material was synthesized by using a hydrothermal method. First, 13.46 g KOH (analytical reagent, AR) was added to 40 mL deionized water to form a 6 mol L^{-1} solution. Then, 0.25 g Te powder (metal basis) and 0.68 g CuCl₂·2H₂O (AR) were added to the solution successively. After stirring for 10 minutes, 8 mL N₂H₄·H₂O (AR) was added, and the mixture was quickly transferred to a 100 mL Teflon vessel. The mixture was heated at 200 °C for 4 h and then cooled to 25 °C. The precipitate was washed with deionized water and dehydrated ethanol for 3 times and then transferred to an oven at 70 °C to dry overnight. Finally, the black precipitate was calcined at 250 °C in a mixed H₂ and Ar (volume ratio = 1:19) atmosphere for 8 h. The as-prepared Cu_{1.81}Te nanorods were synthesized.

Preparation of the electrolyte

The electrolyte was a mixture of 1-ethyl-3-methylimidazolium chloride ([EMIm]Cl, AR) and anhydrous AlCl₃. First, [EMIm]Cl

was heated at 130 °C for 12 h. Then, anhydrous AlCl₃ was added to [EMIm]Cl (n/n = 1.3:1) under stirring. After stirring for 12 h, the electrolyte for AIBs was obtained. All operations were carried out in an argon-atmosphere glovebox.

Electrochemical measurements

In our tests, the electrochemical performance was tested by the soft package batteries. The cathode electrode was prepared by mixing 70 wt% $Cu_{1.81}$ Te powder, 20 wt% acetylene black, and 10 wt% polyvinylidene difluoride (PVDF) binder in *N*-methyl-2-pyrrolidinone (NMP) and then casted on the Mo foil. To remove residual NMP, the electrode plates were dried in a vacuum oven at 80 °C for 12 h. Acetylene black, PVDF and NMP were all battery grade. The assembly process of the soft package batteries was in an Ar-atmosphere glovebox. The CV measurement was conducted on a CHI 660E electrochemical workstation. The constant current (CC) and GITT tests were measured by cycling between 0.2 and 1.5 V *vs.* Al/AlCl₄⁻ on the Neware battery test system.

Characterization

The XRD curves were recorded with Rigaku Ultima IV with Cu K α radiation. The SEM patterns were characterized by HITACHI S-4800. The TEM, HRTEM, and SAED patterns were tested with JEM-2100, while EDS mapping was obtained by Tecnai F30. The XPS data were acquired from PHI Quantum 2000. The ²⁷Al NMR spectra of the electrolyte were tested with a Bruker AVANCE II 400 spectrometer. The Cu element content of the sample was tested by ICP-AES, and the experiments were implemented by Plasma 1000 (NCS Testing Technology).

Results and discussion

We synthesized hexagonal Cu_{1.81}Te by a simple hydrothermal method and then increased the crystallinity of Cu_{1.81}Te by high-temperature calcination in a mixed H₂ and Ar atmosphere. The structural composition and crystallinity of $Cu_{1.81}$ Te were tested by XRD (Fig. 1a). The intensive diffraction peaks at 24.71°, 27.53°, 43.20° and 45.28° correspond to the (006), (023), (220) and (029) planes of Cu_{1.81}Te (PDF no. 83-1320, JCPDS, 2004). There were no other intensive diffraction peaks in the XRD results, indicating that the synthesized materials were of high purity and good crystallinity. The molar ratio of Cu to Te measured by inductively coupled plasmaatomic emission spectrometry (ICP-AES) was 1.76, which was close to 1.81. Through the results of these two characterizations, we determined that the synthesized material was Cu_{1.81}Te. In order to better express the structure of Cu_{1.81}Te, the stick model of Cu_{1.81}Te is shown in Fig. 1b. The green balls represent Cu atoms, and the yellow balls represent Te atoms. It shows that the crystal system of $Cu_{1.81}$ Te is hexagonal and the space group is P3m1. Scanning electron microscopy (SEM) and transmission electron microscopy (TEM) were used to characterize the morphology of $Cu_{1.81}$ Te. The SEM images



Fig. 1 (a) XRD spectrum of $Cu_{1.81}$ Te powder. (b) Ball-and-stick model of the crystal structure of $Cu_{1.81}$ Te. (c and d) SEM images of $Cu_{1.81}$ Te powder. (e) TEM image and (f) HRTEM pattern with the SAED pattern (inset figure) of $Cu_{1.81}$ Te powder.

(Fig. 1c and d) show that the synthesized material possesses rod-like morphology with a diameter between 500 nm and 2 μ m. The TEM test results (Fig. 1e and f) were consistent with that of SEM. The energy-dispersive spectroscopy (EDS) mapping result (Fig. S1†) also demonstrated the uniformity of the distribution of the Cu and Te elements.

For correctly evaluating the electrochemical performance of Cu_{1.81}Te, electrochemical measurements were obtained with the soft-packed batteries to avoid possible side reactions in stainless steel coin cells. Due to the strong corrosive nature of chlorine in the electrolyte, the conventional current collectors such as Cu foils and Al foils corrode at a low potential during tests, which can significantly affect the electrochemical results.⁴² Therefore, we performed linear sweep voltammetry (LSV) tests on different current collectors to find a suitable one. Fig. S2[†] reveals that Mo, Ti and carbon paper remain stable from 0.01 to 1.8 V. When the voltage reached 1.8 V or higher, the polarization current of Mo was the smallest, indicating that the kinetics of side reactions was the slowest; hence, Mo was selected as the current collector. To prove that the Mo current collector did not provide any capacity during the electrochemical experiment, the galvanostatic dischargecharge and cyclic voltammetry (CV) tests were conducted using an aluminum foil as the anode, IL as the electrolyte, and the Mo foil as the cathode without active materials. The result of the CV test (Fig. S3a[†]) illustrated that there was no obvious reduction or oxidation peak between 0.01 and 1.8 V, and the CC electrochemical performance test result (Fig. S3b[†]) further demonstrated that the capacity provided by the Mo current collector was negligible. Therefore, Mo is suitable as a current collector for AIBs.

The galvanostatic discharge and charge curves in the initial cycle at 40 mA g^{-1} are demonstrated in Fig. 2a. The initial discharge capacity was 144 mA h g^{-1} with CE of 101.9%. Fig. 2b exhibits the electrochemical performance of the second to fifth cycles. The capacities were maintained at 190, 164, 166, and 159 mA h g^{-1} , and the corresponding CE values were 88.3%, 97.5%, 95.1% and 95.9%. The generation of irreversible capacity was due to the fact that some Al³⁺ species could not escape from the host lattice of Cu_{1.81}Te and the formation of a solid electrolyte interface (SEI).^{27,28} Fig. 2c shows the cycle performance at 40 mA g⁻¹. The specific capacity was retained at 50 mA h g⁻¹ after 66 cycles. Owing to the continuous insertion and extraction of Al³⁺ during the charging and discharging processes, the structure of the active material will be broken, due to which some of the Al³⁺ species previously trapped in the lattice will be gradually extracted during the subsequent charging process.⁴³ This explains that during the later cycles, the charge specific capacity is higher than the discharge specific capacity and CE is higher than 100%. When the current density increased to 100 mA g^{-1} , the initial capacity was reduced to 71 mA h g^{-1} and remained at 30 mA h



Fig. 2 Constant current charge/discharge curves of (a) the first cycle and (b) from the second to the fifth cycle at 40 mA g^{-1} . (c) Cycling performance at 40 mA g^{-1} . (d) CV curves at 0.1 mV s⁻¹. (e) CV patterns of Cu_{1.81}Te ranging from 0.1 to 1 mV s⁻¹ with fitted lines of log (i_p) and log (v) (inset figure). (f) Fitted lines of i_p and $v^{1/2}$ of reduction peak a and oxidation peak b.

 g^{-1} after 230 cycles (Fig. S4a[†]). In case the first cycle was activated at 10 mA g⁻¹, the subsequent capacity of the second cycle could reach 189 mA h g^{-1} at 100 mA g^{-1} but after 110 cycles, the capacity was also maintained at 30 mA h g⁻¹ (Fig. S4b[†]). After the first cycle is activated at a small current density, the aluminum-ion diffusion channel will be opened, and the diffusion energy barrier will be reduced, which will facilitate the diffusion of more aluminum ions. At the same time, activation will also build a more stable electrochemical interface, which is also conducive to the electrochemical cycle.44-46 Fig. 2d shows the CV curves of Cu1.81 Te with a sweep rate of 0.1 mV s⁻¹ between 0.2 and 1.5 V. There are three reduction peaks at 0.55 V, 0.8 V and 1.3 V and three oxidation peaks at 0.55 V, 0.9 V and 1.35 V, which are consistent with the discharge and charge platforms. Table S1[†] shows the electrochemical performance comparison between Cu1.81Te and other cathode materials that have been reported. It can be found that Cu_{1.81}Te has a long cycling performance and high initial capacity.

The ex situ XPS technique is a very effective means to explore the mechanism of the discharging and charging processes. Fig. 3a-c show the Cu 2p spectra in the first cycle. Fig. 3a displays the Cu 2p spectrum of original Cu_{1.81}Te. Cu $2p_{1/2}$ and Cu $2p_{3/2}$ belong to two sharp peaks at 953.2 eV and 933.1 eV, respectively.^{47,48} The two weaker peaks at 961.6 eV and 941.3 eV correspond to the satellite peaks of Cu^{2+,49} These results indicated that both Cu⁺ and Cu²⁺ were present in original Cu_{1.81}Te.⁵⁰ As shown in Fig. 3b, during the discharging process, the satellite peaks of Cu²⁺ weaken gradually compared to the original state, indicating that some Cu²⁺ species are reduced to Cu⁺ or Cu⁰. Because the binding energies of Cu 2p in Cu⁺ and Cu⁰ are similar, it is hard to judge the presence of Cu^{0.27} Subsequently, during the charging process, the satellite peaks of Cu²⁺ returned to their original state, indicating that Cu⁺ or Cu⁰ was oxidized to Cu²⁺.³² The Al 2p spectrum of the $Cu_{1.81}$ Te electrode in the initial cycle is shown in Fig. S5.[†] The result indicated that during the discharging process, the content of the Al element increased and decreased with the deepening of the charging process. It should be noted that the electrochemical test starts from the discharging process, suggesting a cation insertion mechanism. The above-mentioned results showed that during the discharging process, Al^{3+} was intercalated in $Cu_{1.81}$ Te, accompanied by Cu^{2+} being reduced to Cu⁺ or Cu⁰. During the charging process, Al³⁺ was extracted from the cathode material, along with Cu⁺ or Cu⁰ being oxidized to Cu²⁺.



Fig. 3 XPS spectra of Cu 2p in (a) pristine $Cu_{1.81}$ Te powder, (b) full discharge state and (c) full charge state.

The ex situ XRD patterns of Cu_{1.81}Te at first and second cycles were obtained to further explore the electrochemical reaction mechanism. Fig. 4b shows that the two sharp peaks at 58.76° and 73.99° represent the Mo current collector (Fig. S6,† PDF no. 01-1205, JCPDS, 2004), and the peak positions do not change and migrate during the electrochemical process, indicating the accuracy of the ex situ XRD pattern. During the discharging process in the first cycle, no new peaks appear, indicating that no new phase is generated in this process. It can be found from the PDF standard card of Cu (PDF no. 01-1241, JCPDS, 2004) that the diffraction peaks of Cu completely coincide with the diffraction peaks of Cu_{1.81}Te and therefore, it is difficult to judge the existence of Cu⁰ by the result of XRD. It can be seen from Fig. 4c that the diffraction peak at 24.71° representing the (006) plane moves to a higher angle by 0.15° and the diffraction peak at 27.83° corresponding to the (023) plane moves to a higher angle by 0.19°. The results indicated that with the insertion of Al³⁺, the interlayer spacing of Cu_{1.81}Te decreased due to the strong electrostatic attraction of Al³⁺. After charging to 1.5 V, in addition to the diffraction peaks of Cu_{1.81}Te in the XRD spectrum, the diffraction peaks of Te at 38.23° and 40.60° (PDF no. 01-0714, JCPDS, 2004) also appeared. Fig. S7a and S7b† show the Te 3d spectra in the original and fully charged states. At the initial state, the binding energies of 582.5 eV and 572.9 eV are ascribed to Te 3d_{3/2} and Te $3d_{5/2}$ of Te²⁻, respectively.⁵¹⁻⁵³ However, in the fully charged state, the two 3d peaks of Te move toward the high binding energy direction and reach 586.5 eV and 576.2 eV, respectively, indicating the existence of Te^{4+,54} Moreover, there was a smooth platform at 1.45 V in the initial charging curve. These results indicated that Te²⁻ was partially oxidized during the charging process, which also explained the high CE of the first cycle. This phenomenon has also been reported in the case of sulfides and selenides as the cathodes for AIBs.^{30,55} No new diffraction peak appeared for the discharging and charging processes at the second cycle (Fig. 4b), and the variation of the second cycle was similar to that of the first cycle, indicating that Cu_{1.81}Te had not undergone phase transition since the second cycle. According to the above-mentioned results, we can draw some conclusions. First, Cu_{1.81}Te was used as an intercalation-type cathode material rather than a conversion cathode material. If a conversion reaction occurs, the active material should break after cycling. However, as shown in Fig. S8a and S8b,[†] the morphology of the active material is



Fig. 4 (a) Discharge/charge curves at 40 mA g^{-1} of the initial cycle. (b) XRD spectra of electrodes at different discharge/charge states. (c) Enlarged view of the selected region (24–28°).

well maintained after the initial charging and discharging processes; hence, we think that it is the mechanism of Al^{3+} insertion and extraction. Second, the side reaction of Te oxidation occurs during the charging process in the first cycle, which provides some capacity and leads to higher CE.

CV analyses of Cu_{1.81}Te at different scan rates were used to measure the diffusion kinetics of Al³⁺. Fig. 2e illustrates the CV curves at different scan rates from 0.1 to 1 mV s⁻¹. These curves have similar redox peaks; as the scan rates increase, the oxidation peaks gradually shift to a higher potential, while the reduction peaks gradually shift to a lower voltage, indicating that the polarization of the material increases. According to the Randles–Sevcik formula, the connection between peak currents (i_p) and scan rates (ν) can be represented by the formula given below:

$$i_{\rm p} = a v^b \tag{1}$$

$$\log i_{\rm p} = a + b \log \nu \tag{2}$$

here, *a* and *b* are the adjustable parameters. When *b* approaches 0.5, the electrochemical process is controlled by the ion diffusion step.^{56,57} After calculations, the *b* values at peaks a and b were 0.5153 and 0.5631, respectively, which were close to 0.5, indicating that the diffusion of Al³⁺ was the rate-determining step of the charging and discharging processes. Based on the Randles–Sevcik formula, the relationship between peak currents (i_p) and half power of scan rates ($\nu^{1/2}$) can be described by the following formula:

$$i_{\rm p} = 0.4463 n FAC \left(\frac{n F \nu D}{RT}\right)^{1/2} \tag{3}$$

here, *n* is the charge number, *F* is the Faraday constant, *A* is the surface area of the electrode, *C* is the Al³⁺ concentration in the electrode, *R* is the gas constant, *T* is the Kelvin temperature, *v* is the scan rate and *D* is the diffusion coefficient.⁵⁸⁻⁶⁰ The diffusion coefficient of Al³⁺ can be obtained based on formula (3). Fig. 2f shows the fitted curve. During the discharging and charging processes, the diffusion coefficients of Al³⁺ were calculated and found to be 2.806×10^{-14} and 1.945×10^{-14} cm² s⁻¹, respectively; they were in accordance with the results of the galvanostatic intermittent titration technique (GITT) (Fig. S9†). The results indicated that the diffusion of Al³⁺ in telluride was faster than that for some sulfides (Mo₆S₈: $10^{-17}-10^{-19}$ cm² s⁻¹, Co_3S_4 : $10^{-18}-10^{-20}$ cm² s⁻¹), which was consistent with our prediction that telluride has faster kinetics.

²⁷Al NMR was used to measure the changes in the ion concentration and composition of room-temperature IL during the charging and discharging processes. As shown in Fig. 5, the peaks at 103.95 and 98.18 ppm are assigned to $AlCl_4^-$ and $Al_2Cl_7^{-,61,62}$ respectively. These two peaks can be observed in both the original state and the charging/discharging processes, indicating that $AlCl_4^-$ and $Al_2Cl_7^-$ are the components of room-temperature IL in any process. The calculated molar ratio of the $AlCl_4^-$ and $Al_2Cl_7^-$ ions in the original state was

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Fig. 5 ²⁷Al spectra of the electrolyte at different states.

6.98. The molar ratio decreased to 4.45 after the discharging process and increased to 5.76 after charging. The change in molar ratio was due to the reversible conversion of the $AlCl_4^-$ and $Al_2Cl_7^-$ ions (7 $Al_2Cl_7^-$ + $Al \rightleftharpoons 4 AlCl_4^-$ + 3 e^{-1}) in the IL during the charging and discharging processes, further confirming the insertion and extraction of Al^{3+} in $Cu_{1.81}$ Te.

The morphology and structure of $Cu_{1.81}$ Te at different states were tested by transmission electron microscopy (TEM). Fig. 6a, d and g show the TEM images of $Cu_{1.81}$ Te at the original state, the discharged state and the charged state of the



Fig. 6 (a–c) TEM, HRTEM and SAED images of original $Cu_{1.81}$ Te, respectively. (d–f) TEM, HRTEM and SAED images of the discharged electrode. (g–i) TEM, HRTEM and SAED images of the charged electrode.

first cycle, respectively. The result showed that the rod-shaped morphology of Cu_{1.81}Te was well preserved during the discharging and charging processes. Fig. 6b, e and h show the corresponding high-resolution TEM (HRTEM) images. The lattice fringe spacings of original Cu_{1.81}Te were 0.27 nm, 0.19 nm and 0.32 nm, corresponding to the (210), (128) and (023) crystal planes of Cu_{1.81}Te. Fig. 6c indicates the selected area electron diffraction (SAED) image of the original material, which matches the (222) and (031) crystal planes of $Cu_{1,81}$ Te. The SAED image of the electrode in the discharged state is shown in Fig. 6f. Although the diffraction spot is relatively blurred, the (222) and (023) crystal planes can be distinguished. The result shows that after the intercalation of Al³⁺, the crystallinity of the material is reduced. As shown in Fig. 6i, after the charging process, the crystallinity of Cu_{1.81}Te is recovered, and the (023) crystal plane can be clearly distinguished. The SEM image of the Cu_{1.81}Te electrode after 50 cycles is shown in Fig. S10.† After 50 cycles, although most of the nanorod morphology could be maintained, it was noted that some of Cu1.81 Te was still destroyed, which was one of the main reasons for capacity decay. Fig. S11⁺ exhibits the mapping patterns of the fully discharged Cu_{1.81}Te electrode. It was clear that the Al element was evenly distributed in the rod-like structure of $Cu_{1.81}$ Te, indicating that the Al^{3+} ions were inserted in the entire Cu_{1.81}Te unit cell.

According to the above-mentioned results, the electrochemical mechanism during the discharging and charging of the $Al/Cu_{1.81}$ Te cell can be simplified to the following equation:

Cathode:

$$\operatorname{Cu}_{1.81}\operatorname{Te} + n \operatorname{Al}^{3+} + 3n \operatorname{e}^{-1} \rightleftharpoons \operatorname{Al}_n \operatorname{Cu}_{1.81}\operatorname{Te}$$

Anode:

 $7n \operatorname{Al}_2\operatorname{Cl}_7^- + n \operatorname{Al} \rightleftharpoons 4n \operatorname{AlCl}_4^- + 3n \operatorname{e}^{-1}$

Fig. 7 shows the schematic of the Al/Cu_{1.81}Te battery. During the discharging process, the Al³⁺ ions were inserted into Cu_{1.81}Te, and Al dissolved from the aluminum foil into the electrolyte. The opposite process occurred during the charging process.



Fig. 7 Schematic illustration of the Al/Cu_{1.81}Te battery.

Conclusions

To sum up, novel rod-like $Cu_{1.81}$ Te as a cathode material for AIBs was synthesized by the hydrothermal method. This is also the first time that telluride has been reported in AIBs. Similar to sulfides and selenides, $Cu_{1.81}$ Te exhibited an excellent electrochemical performance. At 40 mA g⁻¹, the initial capacities could reach 144 mA h g⁻¹, and the batteries could be cycled stably over 60 cycles. In addition, we also proved that the Al³⁺ intercalation and de-intercalation reactions occurred during the discharging and charging processes by using *ex situ* XPS, *ex situ* XRD, TEM and ²⁷Al NMR. We also suggested that the attenuation of the capacity may be due to the continuous destruction of the Cu_{1.81}Te structure. In a word, Cu_{1.81}Te is a potential cathode material for AIBs.

Conflicts of interest

There are no conflicts of interest to declare.

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