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Improved pseudocapacitive charge storage in highly ordered mesoporous TiO₂/carbon nanocomposites as high-performance Li-ion hybrid supercapacitor anodes†

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A Li-ion hybrid supercapacitor (Li-HSCs), an integrated system of a Li-ion battery and a supercapacitor, is an important energy-storage device because of its outstanding energy and power as well as long-term cycle life. In this work, we propose an attractive material (a mesoporous anatase titanium dioxide/carbon hybrid material, m-TiO₂-C) as a rapid and stable Li⁺ storage anode material for Li-HSCs. m-TiO₂-C exhibits high specific capacity (~198 mA h g⁻¹ at 0.05 A g⁻¹) and promising rate performance (~90 mA h g⁻¹ at 5 A g⁻¹) with stable cyclability, resulting from the well-designed porous structure with nanocrystalline anatase TiO₂ and conductive carbon. Thereby, it is demonstrated that a Li-HSC system using a m-TiO₂-C anode provides high energy and power (~63 W h kg⁻¹, and ~4044 W kg⁻¹).

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Introduction

High-power energy-storage devices have been regarded as indispensable systems for medium- and large-scale energy-storage applications such as electric vehicles (EVs) and smart grid technologies. This is because the medium- and large-scale energy storage applications require fast charging and discharging behaviors with long cycle lifetime.^{1,2} An electric double-layer capacitor (EDLC), which is one of the typical supercapacitors, is the representative energy-storage device that provides highly excellent power abilities (2–5 kW kg⁻¹) with stable cyclability.^{3–6} However, in spite of its superb abilities, it suffers from low energy performance (~10 W h kg⁻¹) due to its charge-storage mechanism being based on the physisorption of solvated ions at electrode/electrolyte interfaces.^{7–9}

As its alternative device, Li-ion hybrid supercapacitors (Li-HSCs) composed of Li-ion battery (LIB) electrode materials as anodes and EDLC electrode materials as cathodes in a non-aqueous electrolyte containing a Li salt have been intensively researched and developed in recent years.^{10–12} This is because that their electrochemical performance delivering both high

energy and power abilities with stable cyclability is highly attractive. Such unique characteristics result from asymmetric charge-storage mechanisms that Li⁺ from the electrolyte inserts to the anode materials (faradaic reaction) and anions such as PF₆⁻ and ClO₄⁻ are physically adsorbed on the surface of the cathode materials (non-faradaic reaction) during the charge process.¹⁰ However, because of the different charge-storage mechanisms between two electrodes, a kinetics imbalance issue is one of the major problems in Li-HSCs, which should be solved to develop high-performance Li-HSCs.^{13,14} In other words, the reaction mechanism of the anode materials using ionic diffusion in a crystal framework of electrode materials is much more sluggish than that of cathode materials.^{15,16}

To address this kinetics issue, two kinds of strategies have intensively been considered. One of the effective strategies is to reduce particle size of electrode materials in the nanoscale regime.¹⁵ Well-nanosized electrode materials have a variety of merits such as shortened lengths for Li⁺ diffusion/electron mobility (high-power ability) and enhanced electrode/electrolyte interface area (high capacity).^{17–19} In addition, reducing particle size of electrode materials in the nanoscale leads to improved pseudocapacitive reactions associated with surface-controlled reactions, which is kinetically not limited by diffusion-controlled reactions.²⁰ Another strategy is to use attractive carbonaceous and Ti-based anode materials such as graphite and titanium dioxide (TiO₂).^{21–24} Particularly, anatase TiO₂ is one of the promising anode materials for Li-HSCs, because of its beneficial Li⁺ insertion/extraction behaviors and a variety of merits. Firstly, it provides theoretical capacity of ~168 mA h g⁻¹ in the potential range of ~1.7 V (vs. Li/Li⁺) with

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small volume change (<4%) during cycling.^{21,25} Therefore, compared to the graphite anode working in the potential range of ~0.2 V (*vs.* Li/Li⁺), it is free from Li-plating problem and can provide highly stable cycle performance.^{26,27} Secondly, Li-HSCs using anatase TiO₂ anodes can use a cost-attractive and light-weight Al current collector in the anodic part instead of Cu current collector because alloying reaction between Li⁺ and Al does not take place in the potential range of 1.0–3.0 V (*vs.* Li/Li⁺) for anatase TiO₂. In addition, it is non-toxic and one of the highly abundant materials.^{28,29} However, the drawback of anatase TiO₂ is its poor electrical conductivity and relatively low Li⁺ diffusion rate.^{28,30,31}

Here, to rationally design the anatase TiO₂ as anode materials for Li-HSCs, we synthesized mesoporous anatase TiO₂/carbon nanocomposite (denote as m-TiO₂-C) by using block copolymer assisted simple synthesis method. It is clearly demonstrated in this work that design of the synthesized mesoporous electrode material comprising both nanocrystalline anatase TiO₂ and conductive carbon is highly appropriate to solve drawback of the anatase TiO₂ and to maximize its electrochemical performance (high capacity and rate capability), which result from improved Li⁺ diffusion kinetics with rapid electron transport. Furthermore, we proved that the well-designed m-TiO₂-C is highly attractive as anode materials for Li-HSCs delivering high energy and power.

Experimental

Synthesis of m-TiO₂-C

0.15 g of PEO-*b*-PS (poly(ethylene oxide)-*b*-poly(styrene), $M_n = 27\,466\text{ g mol}^{-1}$) was dissolved in 4 mL tetrahydrofuran (THF). 0.9 mL of titanium isopropoxide (TTIP) with 0.3 mL of 35–37% hydrochloric acid (concentrated HCl) was added dropwise to the block copolymer/THF mixture solution with continuous stirring. After 1 h of stirring, the mixture solution was poured to the glass Petri dish. The solvent was evaporated overnight at 40 °C, and then dried at 100 °C. The as-synthesized transparent film was obtained and then, the collected power was heat-treated in Ar atmosphere at 700 °C for 2 h. Commercial TiO₂ purchased from Sigma Aldrich was employed for control group (denoted as com-TiO₂).

Materials characterization

Structure and chemical characterization. The structure and morphology of prepared samples were investigated using scanning electron microscopy (SEM; S-4200 field-emission, Hitachi) and high resolution-transmission electron microscopy (HR-TEM; JEOL JEM-2010). Nitrogen adsorption-desorption analysis was conducted with 77 K with Micromeritics Tristar II 3020 system to estimate pore size and specific surface area. To confirm the specific pore morphology, small-angle X-ray scattering (SAXS) patterns was detected using 4C SAXS beamlines at the Pohang Light Source (PLS). To investigate crystalline phase, X-ray diffraction (XRD) pattern was identified by D/max-2500 a diffractometer (Rigaku, Cu-K α radiation). The carbon content of m-TiO₂-C was estimated using

thermogravimetric analysis (TGA; NETZSCH STA 449C). Electron energy loss spectroscopy (EELS) analysis was performed to identify containing elements using energy-filtering transmission electron microscopy (EF-TEM, JM-220FS).

Electrochemical measurements. For half and full-cell electrochemical tests, active materials including m-TiO₂-C and com-TiO₂ (80 wt%) were homogeneously mixed with super-P carbon (10 wt%) and polyvinylidene fluoride (PVDF, 10 wt%) in *N*-methyl-2-pyrrolidone (NMP). The slurries were pasted on current collector using doctor blade. The prepared electrodes were dried at 60 °C for 6 h and then, 110 °C for 12 h in vacuum oven. Subsequently, the electrodes were roll-pressed. For half-cell test, 2032-type coin cells were fabricated with lithium metal as both counter and reference electrodes in Ar-filled glovebox. The mass loading of active materials used as anode materials were carefully controlled around 1.0 mg cm⁻². The electrolyte was 1.0 M of LiPF₆ in mixture of ethylene carbonate/dimethyl carbonate (EC/DMC, 1 : 1 volume ratio, Panaxetec. Co., Korea). The activated carbon electrode used as cathode materials was prepared using commercially available MSP-20 (90 wt%), conductive carbon (5 wt%), and polytetrafluoroethylene (PTFE, 5 wt%). In the half-cell tests, the working voltages for anode and cathode materials were 1.0–3.0 and 3.0–4.5 V (*vs.* Li/Li⁺), respectively. In the case of full-cell, Li-HSC was assembled using m-TiO₂-C as an anode and MSP-20 as a cathode and the weight ratio of anode and cathode active materials was controlled to 1 : 3.5 in working voltages of 0–3.0 V. The energy and power of the Li-HSCs was calculated by numerically integrating the galvanostatic discharge profiles using eqn (1) and (2) as follows.³²

$$E = \int_{t_1}^{t_2} IV dt \quad (1)$$

where I is the constant current (A g⁻¹), V is the working voltage (V), t_1 and t_2 are the start/end of discharge time (s) of Li-HSCs, respectively, and

$$P = \frac{E}{t} \quad (2)$$

where t is the discharge time (s). All electrochemical tests were conducted using the WBCS-3000 battery cyclers (WonA Tech, Korea).

Results and discussion

Material characterization

The m-TiO₂-C was synthesized using a block copolymer-assisted synthesis method and schematic representation on the synthesis method is shown in Fig. 1. In brief, the Ti precursor (TTIP) and laboratory-made amphiphilic block copolymer (poly(ethylene oxide)-*b*-poly(styrene), PEO-*b*-PS) synthesized by atomic transfer radical polymerization (ATRP) were dissolved in tetrahydrofuran (THF). And then, to induce sol-gel reaction of TTIP, concentrated hydrochloric acid (HCl) was slowly added to the TTIP/block copolymer/THF mixture solution. The titanium oxide sol made by hydrolysis selectively interacts with the hydrophilic PEO part of PEO-*b*-PS *via* hydrogen bonds. During



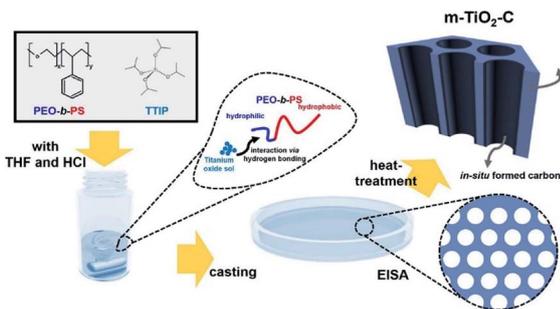


Fig. 1 Schematic illustration of the synthesis of m-TiO₂-C.

the evaporation of organic solvent at 40 °C (evaporation-induced self-assembly, ESIA), highly ordered mesostructure is formed by self-organization of titanium oxide sol/block copolymer mixture.^{33–36} After drying at 100 °C to induce the cross-linkage of titanium oxide sol, the as-synthesized TiO₂/block copolymer composite was heat-treated at 700 °C under inert atmosphere (Ar condition). During heat-treatment, as-synthesized TiO₂ is converted to crystalline TiO₂. In addition, PS part of PEO-*b*-PS is converted to mechanically stable and conductive carbon (*in situ* formed carbon in m-TiO₂-C). For comparison, commercially available TiO₂ (com-TiO₂) was employed.

Fig. 2a and b show that X-ray diffraction (XRD) patterns of m-TiO₂-C and com-TiO₂ are in good agreement with anatase TiO₂ (JCPDS no. 21-1272) with no noticeable impurities. Their average crystallite sizes calculated using the Debye–Scherrer equation were ~14 (m-TiO₂-C) and ~100 (com-TiO₂) nm, respectively.³⁷ In addition, high-resolution transmission electron microscopy (HR-TEM, Fig. 2c) image of m-TiO₂-C clearly shows the lattice fringes of anatase TiO₂ with (101) and (004) spacing of 0.353 and 0.238 nm, respectively, again indicating

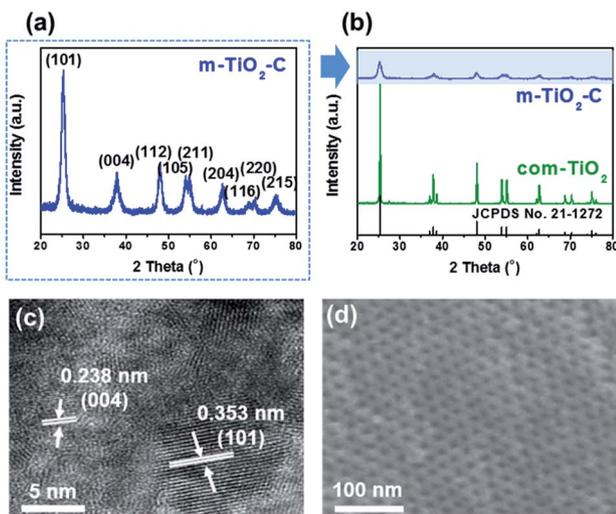


Fig. 2 (a) XRD pattern on m-TiO₂-C. (b) Comparison of XRD patterns on m-TiO₂-C and com-TiO₂. (c) HR-TEM image of m-TiO₂-C. (d) SEM image of m-TiO₂-C.

that the crystal structure of m-TiO₂-C well match the anatase TiO₂ phase. Mesoporous structure of m-TiO₂-C is identified through scanning electron microscopy (SEM) image, N₂ adsorption-desorption technique, and small-angle X-ray scattering (SAXS) patterns. SEM image (Fig. 2d) shows highly ordered mesoporous structure with uniform pore size. Compared to m-TiO₂-C, com-TiO₂ has irregular and larger particle shape/size without porosity (Fig. S1†). In addition, N₂ adsorption isotherm (Fig. 3a) of m-TiO₂-C corresponds to type IV curve with a sharp adsorption at ~0.9 *P*/*P*₀, indicating that uniform mesopores are predominant. The main pore size calculated using the Barret–Joyner–Halenda (BJH) method and specific surface area calculated using the Brunauer–Emmett–Teller (BET) of m-TiO₂-C were ~13 nm and ~123 m² g⁻¹ (Fig. 3b), respectively, which is significantly higher than that of com-TiO₂ (<2 m² g⁻¹). The mesoporous structural characterization of m-TiO₂-C is further demonstrated by SAXS pattern (Fig. 3c). Scattering peaks of m-TiO₂-C with a peak position ratio of 1 : 3^{1/2} : 4^{1/2} suggest that hexagonally ordered TiO₂ structure with a long-range order is well established.³⁸ To confirm the presence of *in situ* formed carbon in m-TiO₂-C, thermogravimetric analysis (TGA) and electron energy loss spectroscopy (EELS) analysis were employed. TGA result (Fig. 3d) proves that the *in situ* formed carbon content in the m-TiO₂-C is around 10 wt%. Furthermore, EELS analysis image (Fig. 4) directly shows the existence of *in situ* formed carbon in the m-TiO₂-C, representing that Ti, O, and C are uniformly dispersed.

Electrochemistry

Galvanostatic charge-discharge (de-lithiation and lithiation) test on m-TiO₂-C was conducted in the potential range of 1.0–3.0 V (*vs.* Li/Li⁺), showing that the m-TiO₂-C provides reversible charge capacity of ~198 mA h g⁻¹ at current of 0.05 A g⁻¹ (Fig. 5a). The typical plateau at a potential of ~1.7 V (*vs.* Li/Li⁺) demonstrates reversible Li⁺ intercalation into anatase TiO₂

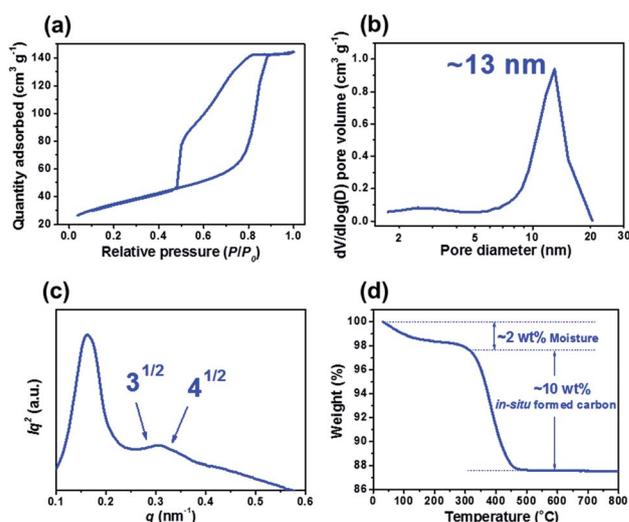


Fig. 3 (a) N₂ physisorption isotherm and (b) pore size distribution of m-TiO₂-C. (c) SAXS pattern of m-TiO₂-C. (d) TGA result of m-TiO₂-C.



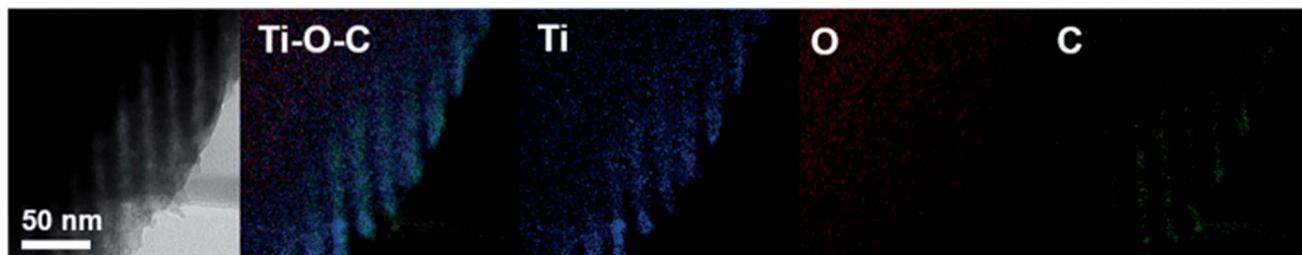


Fig. 4 EELS mapping images of m-TiO₂-C.

lattice, where $\text{TiO}_2 + x\text{Li}^+ + xe^- \leftrightarrow \text{Li}_x\text{TiO}_2$ ($0 \leq x \leq 0.5$).^{22,28,39} Fig. 5b shows the specific capacity of m-TiO₂-C is much higher ($\sim 198 \text{ mA h g}^{-1}$) than that of com-TiO₂ ($\sim 110 \text{ mA h g}^{-1}$) at current of 0.05 A g^{-1} . In addition, compared to com-TiO₂, the m-TiO₂-C showed much better rate capability with increasing currents from 0.05 to 5 A g^{-1} , also indicating that capacity retention of m-TiO₂-C with increasing currents is significantly outstanding than that of com-TiO₂ (Fig. S2a†). Fig. 5c and S2b† shows that the galvanostatic charge–discharge curves of m-TiO₂-C with increasing currents are well maintained with small overpotentials (reduced internal resistance) compared to those of com-TiO₂. It represents that well-ordered structure with conductive *in situ* formed carbon is highly beneficial to improve electrochemical performances of anatase TiO₂, mainly due to a variety of merits including (i) shortened diffusion lengths of Li⁺, (ii) superior electron mobility, (iii) easy penetration of electrolyte, (iv) plentiful charge-storage sites, and *etc.*^{34,40–42} It should be noted that the m-TiO₂-C exhibits better rate capability than other anatase TiO₂ anodes previously reported (Fig. S3†) and delivers highly stable cycle stability (capacity retention of $\sim 94\%$ with $\sim 100\%$ coulombic efficiency for ~ 350 cycles) at the current of 0.5 A g^{-1} (Fig. 5d and S4†).^{43–47} In addition, it shows

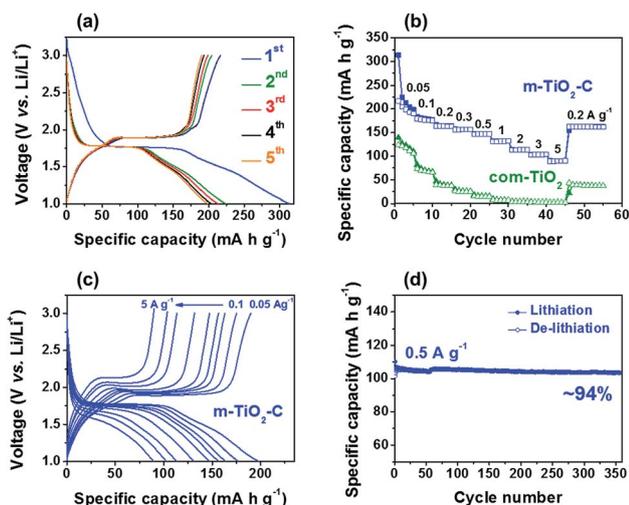


Fig. 5 (a) Galvanostatic charge–discharge profiles of m-TiO₂-C at 0.05 A g^{-1} . (b) Comparison of rate capability of TiO₂ electrodes at different currents from 0.05 to 5 A g^{-1} . (c) Galvanostatic charge–discharge profiles of m-TiO₂-C at various currents from 0.05 to 5 A g^{-1} . (d) Cycle performance of m-TiO₂-C at a current of 0.5 A g^{-1} .

stable long-term cyclability at high current of 3 A g^{-1} for 1000 cycles (Fig. S4,† capacity retention of $\sim 97\%$ with $\sim 100\%$ coulombic efficiency). Because the excellent rate capability and long-term cycle stability are main factors for application to anodes of Li-HSCs, m-TiO₂-C developed in this work could be the extremely potential Li-HSC anode material.

To further reveal the reason why m-TiO₂-C provides superior electrochemical behaviors, we performed cyclic voltammetry (CV) tests on m-TiO₂-C and com-TiO₂ conducted at sweep rates from 0.1 to 1.0 mV s^{-1} in the potential range of 1.0–3.0 V (*vs.* Li/Li⁺) as shown in Fig. 6a and S5.† The pair of redox peaks (1.7–2.0 V *vs.* Li/Li⁺) of m-TiO₂-C and com-TiO₂ at sweep rate of 0.1 mV s^{-1} well match the lithiation and de-lithiation reactions. In previous studies, it is well known that the anatase TiO₂ is influenced by the diffusion-controlled process with a two-phase process. Therefore, from $i = av^b$ equation (where a (mV s^{-1})) with CV tests, b value of anatase TiO₂ is close to 0.5, indicating that the diffusion-controlled reactions is more dominant than

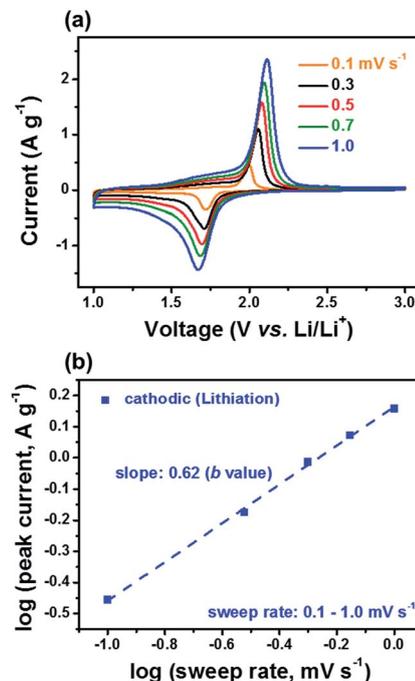


Fig. 6 (a) CV curves of m-TiO₂-C at different sweep rates of 0.1– 1.0 mV s^{-1} . (b) $\log(i)$ vs. $\log(v)$ plot of cathodic peak current on m-TiO₂-C electrode.



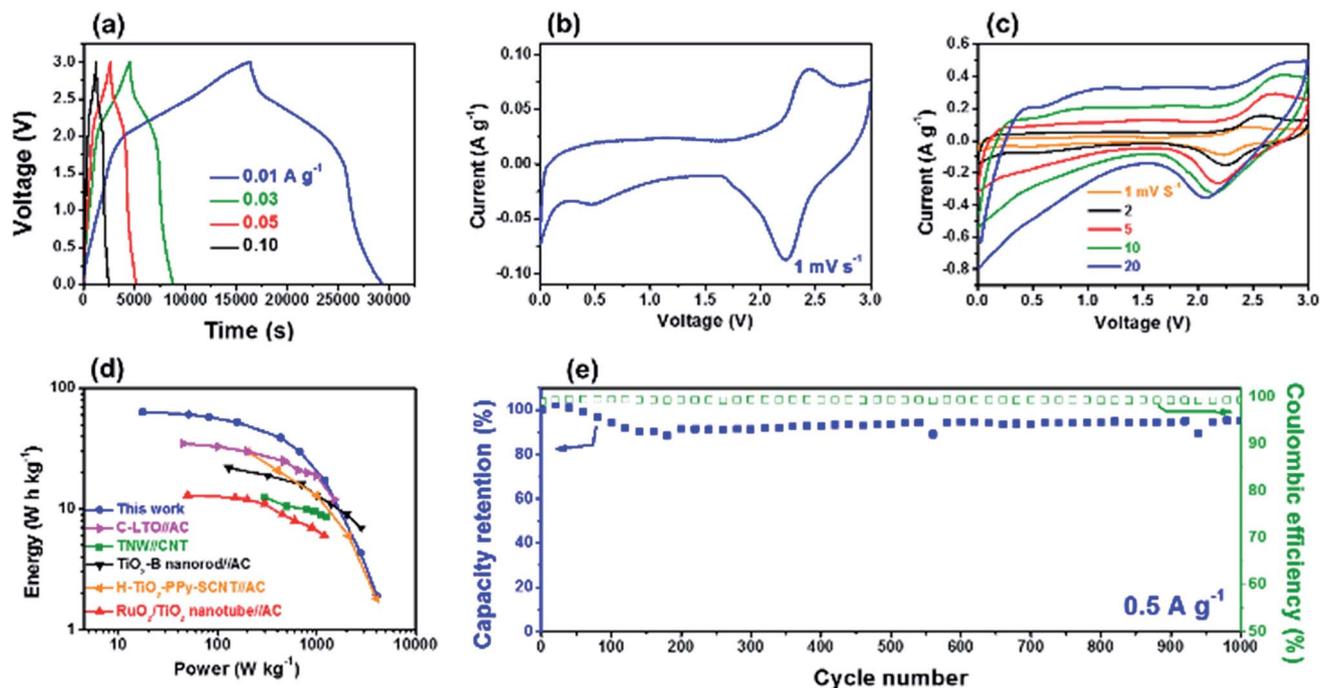


Fig. 7 (a) Galvanostatic charge–discharge profiles of the Li-HSC at different currents from 0.01 to 0.1 A g⁻¹. (b and c) CV curves of the Li-HSCs at different sweep rates of 1–20 mV s⁻¹. (d) Ragone plots compared with previously reported results. (e) Cycle performance of the Li-HSC at a current of 0.5 A g⁻¹.

the surface-controlled reaction ($b = 1.0$).^{20,48,49} As shown in Fig. 6b and S6†, b value of m-TiO₂-C obtained from peak currents of the lithiation process is around 0.62, indicative of that the charge-storage reaction mechanism in m-TiO₂-C is influenced by both the surface- and diffusion-controlled reactions (improved pseudocapacitive reaction).⁴⁹ It is considerably contrast to that of com-TiO₂ showing severe deviation with increasing sweep rates from 0.1 to 1.0 mV s⁻¹ because of huge internal resistance. In addition, b values of m-TiO₂-C obtained in potential range of 1.2–2.1 V (vs. Li/Li⁺) are also higher than 0.5 (Fig. S6†).

Before building Li-HSC system using the m-TiO₂-C anode, electrochemical behavior of MSP-20 (commercially available activated carbon) as a cathode material for the Li-HSC was investigated by galvanostatic charge–discharge half-cell test. Fig. S7† shows that reversible specific capacity of the MSP-20 is ~60 mA h g⁻¹ at a current of 0.05 A g⁻¹ in the voltage range of 3.0–4.5 V (vs. Li/Li⁺). Considering the specific capacities and working voltages between m-TiO₂-C anode and MSP-20 cathode, the Li-HSC system using m-TiO₂-C anode and MSP-20 cathode was carefully assembled and its galvanostatic charge–discharge tests were conducted at various currents in the potential range of 0.0–3.0 V (Fig. 7a and S8†). Galvanostatic charge–discharge curves of Li-HSC using m-TiO₂-C as anode and MSP-20 as cathode at current rates from 0.01 to 5 A g⁻¹ do not exhibit typical triangular shape, dissimilar to those of conventional symmetric SCs.⁵⁰ It is again confirmed from CV data at sweep rates of 1–20 mV s⁻¹ (Fig. 7b and c) that the CV profile does not follows a typical rectangular shape of conventional symmetric SCs. The galvanostatic charge–discharge and CV shapes of the

Li-HSC are mainly due to combination of the faradaic reaction at the m-TiO₂-C anode and the non-faradaic reaction at the MSP-20 cathode.²² The maximum energy and power of the Li-HSC calculated using eqn (1) and (2) with galvanostatic charge–discharge curves were ~63 W h kg⁻¹ and ~4044 W kg⁻¹, respectively. The Ragone plot on the trade of relationship between energy and combination of the faradaic reaction at the m-TiO₂-C anode and the non-faradaic reaction at the MSP-20 cathode.²² The maximum energy and power of the Li-HSC calculated using eqn (1) and (2) with galvanostatic charge–discharge curves were ~63 W h kg⁻¹ and ~4044 W kg⁻¹, respectively. The Ragone plot on the trade of relationship between energy and power shows that its energy and power is much better than that of other results previously reported (Fig. 7d).^{43,51–54} Finally, long-term cycle stability of the Li-HSC was investigated. Fig. 7e shows that the cycling stability of the Li-HSC at a current of 0.5 A g⁻¹ is well maintained with ~100% coulombic efficiency for 1000 cycles. The Ragone plot and cycle performance data imply that m-TiO₂-C is highly suitable as the anode material for Li-HSC system.

Conclusions

In summary, we reported the block copolymer assisted one-pot synthesis method of m-TiO₂-C and its application for high-power anode materials of Li-HSC. The m-TiO₂-C delivered high specific capacity (~198 mA h g⁻¹ at 0.05 A g⁻¹) and rate capability (~90 mA h g⁻¹ at 5 A g⁻¹) with stable cycle performance. Its electrochemical performance was superior compared to com-TiO₂, mainly due to synergistic effects of



unique mesostructure and hybridization between anatase TiO₂ and *in situ* formed carbon. Thereby, the Li-HSC system using the m-TiO₂-C anode possessed high energy and power abilities ($\sim 63 \text{ W h kg}^{-1}$ and $\sim 4044 \text{ W kg}^{-1}$) in the potential range of 0.0–3.0 V, implying that the energy-storage device (Li-HSC) can be a promising alternative to conventional symmetric SCs.

Conflicts of interest

There are no conflicts to declare.

Acknowledgements

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