

Article

Silicon/Biomass Carbon Composite as a Low-Cost Anode for Lithium-Ion Batteries

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Abstract: Various biomass materials have been developed as precursor materials to fabricate carbon-based anodes for Li-ion battery (LIB) applications due to their inherent sustainability and low cost. However, the low theoretical specific capacity of carbon materials (273 mAh g^{−1}) restricts their further application as an anode for LIBs. Herein, silicon/reed catkin composites (Si/RC) and silicon/apricot shell-derived carbon (Si/AC) are successfully fabricated, and their performances are evaluated as anode materials for LIBs. The Si/RC anode displays a reversible capacity of 318.4 mAh g^{−1} after 100 cycles at 200 mA g^{−1} and remains 229.3 mAh g^{−1} after 1000 cycles at 1 A g^{−1}. This work discloses the feasibility and promising prospects of utilizing biomass reed catkins to prepare low-cost, porous, carbon-based materials for energy storage applications.

Keywords: reed catkins; apricot shells; biomass carbon; anode; lithium-ion battery



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1. Introduction

Lithium-ion batteries (LIBs) are regarded as one of the most hopeful energy storage devices due to their excellent energy density [1–3]. They have been found to have extensive applications in electric vehicles and portable electronic devices [4–6]. Recently, to respond to the energy crisis caused by rapid economic development, the research and production of novel electrode materials with outstanding performance, low cost, and renewability have become hot topics [7,8]. Biomass carbon (BC) is considered a promising electrode material due to its abundance, cost-effectiveness, environmental friendliness, and renewability [9–12]. Compared with other materials, the use of BC as battery electrodes not only reduces the cost of batteries and creates economic benefits but also alleviates the severe energy crisis and environmental pollution problems [13–17].

Biomass has recently attracted growing attention as one of the most abundant sustainable resources on the earth. Some typical biomass materials, such as reed [18–20], apricot shell [21], cherry pit [22], corn stalk [23], wheat straw [24,25], rice straw and husk [25–27], banana peel [28], grass [29], etc., have been used to synthesize carbon-based anodes with various morphologies and structures for LIB applications. Compared to other biomass, reed catkins (RCs) are easily available, which reduces the cost, and the natural fluffy structure is favorable for the fabrication of porous carbons with high specific surface area [18]. Apricot shell-derived carbon (AC), an inexpensive and relatively abundant agricultural by-product,

is also an inedible residue of lignocellulosic nature from which many studies have been conducted to obtain activated carbon using various methods [19]. Both RC and AC have the characteristics of relatively wide source and low cost and are suitable for being used as biomass carbon in the anode of LIBs. However, the low theoretical capacity density of carbon materials limits the further development of battery performance [30]. Silicon, as one of the common anode materials for LIBs, has an extremely high theoretical capacity. The reversible capacity of Si electrodes is dramatically reduced due to severe volume change during the charging/discharging process and the poor conductivity [31,32]. Recently, the combination of silicon and carbon to produce Si/C composites was reported to improve the capacity of the carbon-based materials effectively [33–35]. Liu et al. have reported a new preparation way to synthesize three-dimensional porous Si/C nanoarchitectures by extending the famous Mg thermal method [36]. Lucia et al. have reported a Si/C composite with a silicon content of 30 wt.% with a specific capacity of up to 917 mAh g^{−1} after 200 cycles and good long-cycling stability at high current densities [37]. Chen et al. prepared a silicon/graphite composite by optimizing the ratio of silicon to graphite, showing a capacity of 522.4 mAh g^{−1} at 1 A and good cycling stability with 92.9% capacity retention after 400 cycles [38]. The above studies prove that the Mg thermal reduction method can effectively create Si to produce Si/C composites. In order to meet the market demand, more low-cost Si/C composites with various structures are worth developing.

In this paper, reed catkins and apricot shells are selected for biomass carbon sources, which is for the following reasons: Firstly, an excellent carbon source for biomass should be characterized by a low price and a wide range of sources, and we selected the current materials for the study with the above factors in mind. Secondly, in terms of material properties, the commonly used biomass carbon is often taken from plants (non-edible), edible plants or their fruits, agricultural by-products, and so on. In order to avoid wasting food, we chose inedible reed catkins and agricultural by-products, almond shells, as carbon sources. Thirdly, from the geographical aspect, coconut shell, as a typical biomass material in southern China, has been widely studied and applied, while representative biomass carbon sources are still lacking and should be developed in northern China. Therefore, reed flowers and almond shells, which exist in large areas and in large quantities in northern China, are selected as representative research objects.

In this study, we report the synthesis of Si/RC and Si/AC composites through an inexpensive Mg thermal reduction process [39,40] using RC and AC as biomass sources. The as-obtained Si/RC and Si/AC present porous structures, in which Si spheres are attached to pores or the surface of carbon materials. Thus, good electrical conductivity, high specific capacity, and excellent electrochemical properties can be obtained. The work can provide ideas for the development of low-cost biomass carbons and has the potential to facilitate the application of biomass carbons in energy storage fields.

2. Materials and Methods

The schematic diagram for synthesizing the Si/RC and Si/AC is shown in Figure 1. RC and AC were firstly crushed and cleaned with anhydrous ethanol. The RC and AC were heated to 300 °C in a tube furnace in an Ar atmosphere at a heating rate of 3 °C/min for 2 h and then directly heated to 750 °C at a heating rate of 3 °C/min for 1 h and then cooled in the furnace to obtain the porous carbons. The porous carbons were mixed with KOH in a mass ratio of 1:4 and heated to 750 °C in a tube furnace in an Ar atmosphere at a heating rate of 3 °C/min for 1 h. They were then immersed in 1 M HCl for 12 h to obtain the RC and AC precursors [41].

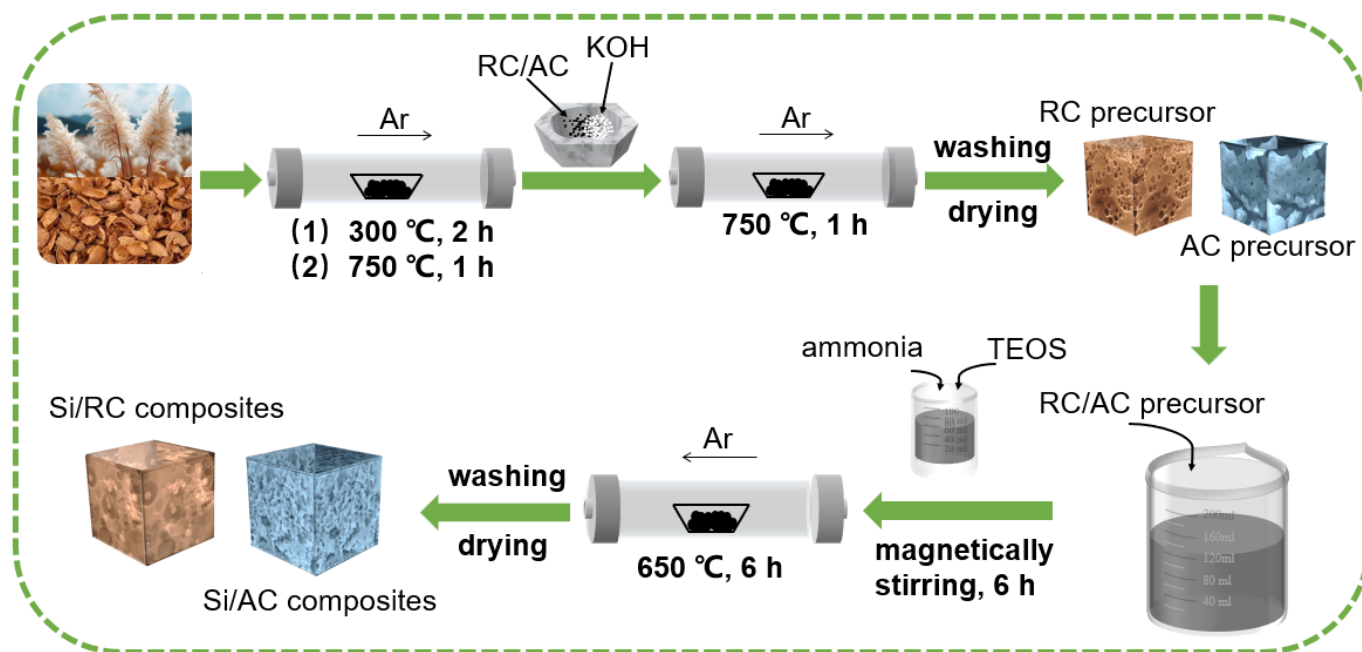


Figure 1. Schematic diagram showing the fabrication process of Si/RC and Si/AC composites.

The precursors were dispersed in the solution consisting of 40 mL of anhydrous ethanol and 5 mL of deionized water and mixed with ultrasonication. After stirring well, 10 mL of 25% ammonia and 0.7 mL of 98% tetraethyl orthosilicate (TEOS) were added to form SiO_2 , and the solutions were magnetically stirred for 6 h [36]. The intermediate products were then obtained by drying at 60 °C for 12 h. Then the products were mixed with NaCl and Mg powder (in a mass ratio of 1:10:1) and were further heated to 650 °C in a tube furnace in an Ar atmosphere at a heating rate of 3 °C/min for 6 h to reduce SiO_2 to Si [39,40]. After cooling, the samples were immersed in 1 M HCl for 6 h to remove MgO and Mg_2Si that may be formed in the reaction and immersed in 0.5 M HF for 30 min to remove the remaining SiO_2 . Finally, the Si/RC and Si/AC composites were obtained after washing with anhydrous ethanol and drying at 60 °C for 12 h. Alcohol is used to clean the surface of the sample from residues and to prevent oxidation of the material in the work.

The preparation route involved in the study has a good scale-up feasibility. Firstly, the sintering process and magnesium thermal reduction method we used are methods that have been proven to be feasible by commercial production. Secondly, the biomass raw materials we used are abundant and easy to collect, which greatly reduces the production cost and improves the economic efficiency. In addition, this work introduces Si using biomass as a precursor, and this scheme can also be used to introduce other elements to form other carbon-based composites with better synthesis scalability.

The phase structure of the Si/RC and Si/AC was analyzed by X-ray diffraction (XRD, Rigaku D8 Discover, Karlsruhe, Germany) using $\text{Cu-K}\alpha$ radiation. Scanning electron microscopy (SEM, Quanta 450 FEG, Hillsboro, OR, USA) and transmission electron microscopy (TEM, JEM-2010, Tokyo, Japan) were utilized to reveal the microstructure and morphology of the sample. The Raman spectra were identified using a Laser Raman Spectrometer (Raman, LabRAM HR Evolution, Tokyo, Japan) with a 632 nm laser. The valence states of elements were examined by X-ray photoelectron spectra (XPS, ESCALAB 250Xi, Waltham, MA, USA). Specific surface area and porosity were analyzed by the nitrogen low-temperature adsorption method, according to BET (Brunauer, Emmertt, and Teller) and BJH (Barrett, Joyner, and Halenda) equations.

To assemble half cells for electrochemical testing, the slurry was formed by mixing 70% Si/RC or Si/AC composite, 20% Ketjen black, and 10% carboxymethylcellulose (CMC) binder in deionized water. The anodes were then prepared by coating the slurry on Cu foil and drying at 60 °C for 10 h. The Si/RC or Si/AC anodes, lithium cathodes, Celgard 2400 membranes, and electrolyte (1 M LiPF₆ dissolved in a mixture of ethylene carbonate (EC) and diethyl carbonate (DEC), EC/DEC = 1:1 by volume) were encapsulated in CR2025 coin cells in a glove box containing high purity Ar (H₂O < 0.01 ppm and O₂ < 0.01 ppm). Cyclic voltammetry (CV) and electrochemical impedance spectroscopy (EIS) of the cells were tested at 0.1 mV s^{−1} from 0.01 to 3 V on an electrochemical workstation (Princeton, Versa STAT 4). Galvanostatic charge–discharge (GCD) was tested with a NEWARE battery tester in the voltage window of 0.01–3 V (vs. Li/Li⁺).

3. Results

Figure 2a,d shows the SEM images of the RC and AC carbon materials, revealing plentiful and well-distributed pores. The morphology of the Si/RC is displayed in Figure 2b,c, where porous Si spheres can be observed in the pores of RC. Figure 2e,f uncovers the morphology of the Si/AC, where Si can be seen as a clustering of peanut-shaped particles attached to the surface of AC. Obviously, Si in the Si/RC can adhere to the pores of RC more probably because the pores of RC are larger and richer than AC, while Si in the Si/AC is only distributed on the surface of AC. Therefore, compared to the Si/AC, the Si/RC exhibits a more favorable structure to obtain good electrochemical properties. The structural similarity between AC and RC is that they both have a certain amount of mesopores and macropores, thus forming two scales of porous structure. The differences between AC and RC are the pore size and volume ratio of macropores. The macropore diameter of RC is about ~3 µm, and the percentage of macropore is 40–60%, while the macropore diameter of AC is only ~1 µm, and the percentage of macropore is only 3–10%.

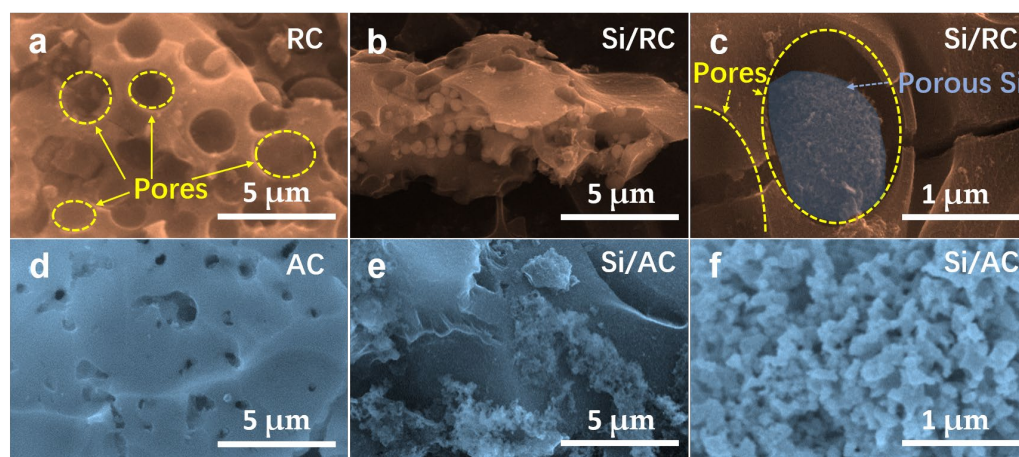


Figure 2. SEM images of RC (a), Si/RC (b,c), AC (d) and Si/AC (e,f).

To further reveal the microstructure of the Si/RC, TEM tests were conducted. Figure 3a,b show that the RC in Si/RC is present in the form of thin films, while most of the Si is attached to the RC film in the form of porous spheres, which is consistent with the SEM result. Furthermore, there are many nanoscale pores distributed in the RC films (Figure 3b,c). Figure 3d presents the boundary between Si and C, where the arc-shape boundary further confirms that Si spheres are attached to the surface or pores of RC. The elemental mapping of the Si/RC is shown in Figure 3e–h, which show the distribution law of different elements. The atomic percentages of C, Si, and O are 55.9%, 27.3%, and 16.8%, respectively. It can be seen that the spherical particles are mainly composed of silicon, while

the substrate is mainly composed of carbon elements. Oxygen elements can be seen to be evenly distributed in the material and may be attributed to slightly surface oxidation of the sample in the air, or incomplete reduction and removal of SiO_2 .

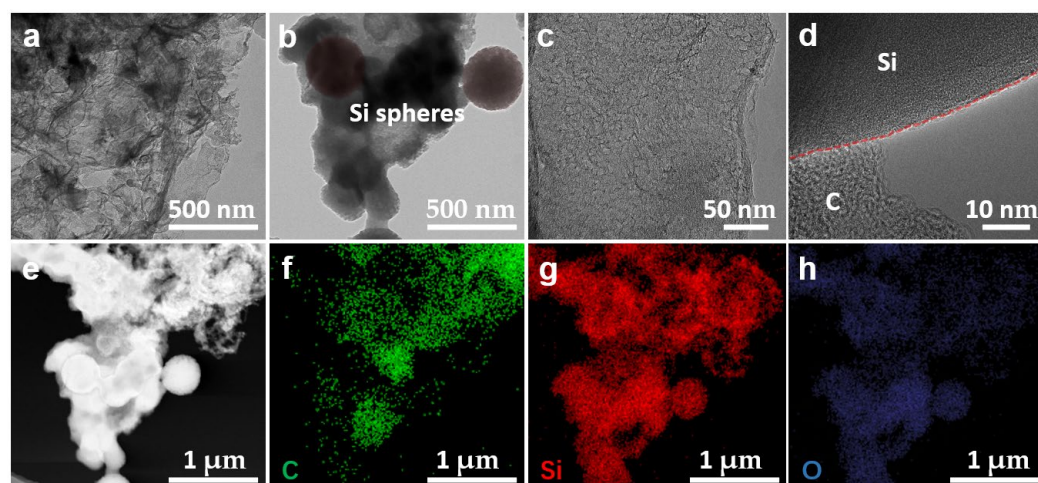


Figure 3. TEM images of Si/RC (a–d) and EDS mapping images of C, Si, and O elements of Si/RC (e–h).

To disclose the pore characteristics of the Si/RC, nitrogen adsorption–desorption experiments were tested (Figure 4). Figure 4a presents the N_2 adsorption–desorption isotherms, showing a type II form with a type H4 hysteresis loop, which reveals the presence of mesopores. The specific surface area of the composite is shown to be $791.89 \text{ m}^2/\text{g}$. Moreover, Figure 4b exhibits the pore size distribution curve of the Si/RC. The pore size is basically in the scope of 2–5 nm. The as-prepared material uncovers mesoporous characteristics, which is beneficial for the insertion/exsertion of Li^+ and relieves the volume expansion of Si, thus presenting the potential to improve cycle performance.

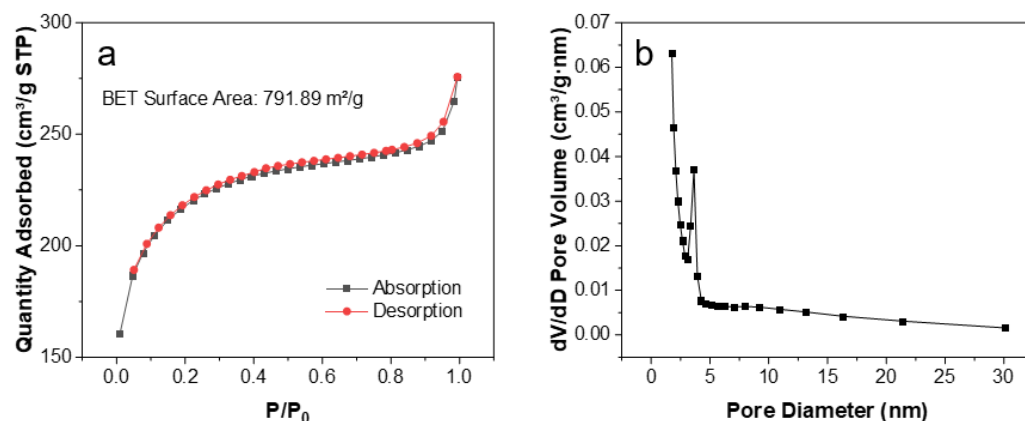


Figure 4. (a) Nitrogen adsorption–desorption isotherms of Si/RC. (b) Pore size distributions of Si/RC.

The XRD patterns present the crystal structure of the Si/RC and Si/AC. As displayed in Figure 5a, the broad peaks centered at 2θ around 22° and 44° correspond to the (002) and (101) planes of amorphous carbon structure generated by the RC and AC [42], while the sharp peaks at 2θ around 28° , 47° , 56° , 69° , 76° , and 88° correspond to the (111), (220), (311), (400), (331), and (422) planes of Si (JCPDS No. 75-0589), which indicates that SiO_2 is successfully converted to Si. Moreover, the peak intensity of Si in the Si/RC is stronger than that in the Si/AC, indicating that the Si/RC has a higher content of Si than the Si/AC. The macropores in RC are richer and more regular than those in AC, which makes it

easier for the Si spheres to be retained within the carbon matrix instead of sticking to the carbon surface.

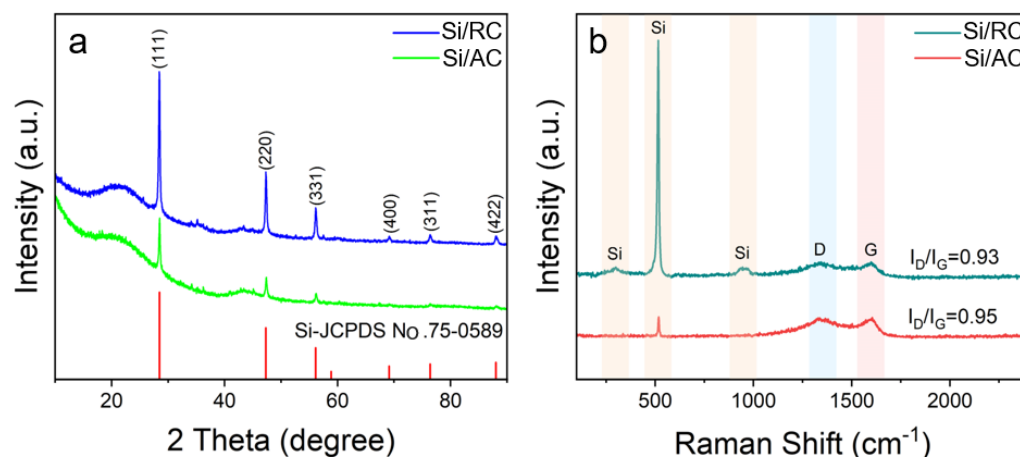


Figure 5. XRD pattern (a) and Raman spectra (b) of Si/RC and Si/AC.

Figure 5b shows Raman spectra of the Si/RC and Si/AC, further confirming their composition and the degree of carbon graphitization. It can be observed that the Raman spectra of both the Si/RC and Si/AC contain the distinct peaks of C and Si, which proves that the Si/RC and Si/AC have been successfully synthesized in this work. Two detached characteristic peaks at 1353 cm^{-1} and 1597 cm^{-1} can be assigned to the D band and G band, respectively [43]. And the D band to G band ratio (I_D/I_G) of the Si/RC and Si/AC are 0.93 and 0.95, respectively, which indicates that the degree of graphitization of the RC precursor is slightly higher. The sharp peaks of the Si/RC at around 300 cm^{-1} , 514 cm^{-1} , and 949 cm^{-1} correspond to the characteristic peaks of crystalline Si, whereas the peaks of the Si/AC only show a distinct characteristic peak of crystalline Si at around 514 cm^{-1} [44]. Obviously, the Si/RC has a higher content of Si because the characteristic peaks of Si in the Si/RC are sharper than those of the Si/AC, which is in accordance with the XRD results.

Figure 6 discloses the valence states of the elements in Si/RC by XPS. The full XPS survey confirms the existence of C, Si, O, N, and F elements (Figure 6a). The characteristic peaks at around 102, 153, 285, 400, 534, 688, 980, and 1225 eV are in accordance with Si 2p, Si 2s, C 1s, N 1s, O 1s, F 1s, O KLL, and C KLL, respectively. In addition, the high-resolution XPS spectra of C 1s is fitted to four characteristic peaks located at 284.8, 285.7, 287.5, and 290.3 eV, which relate to the C-C, C-O, semi-ionic C-F, and C=O bonds (Figure 6b), respectively [45]. The appearance of semi-ionic C-F bonds may be caused by residual traces of HF [46]. The Si 2p spectrum is fitted to three decomposed peaks located at 100.9, 102.0, and 103.9 eV, which can be ascribed to the Si-Si, Si-C, and Si-O bonds, respectively (Figure 6c). The Si-C bonds are attributed to some Si doping into carbon. The O 1s spectrum is fitted to three decomposed peaks located at 532.3, 533.7, and 534.9 eV, which are in accord with the Si-O, C-O, and O-C=O bonds, respectively (Figure 6d), further confirming the trace existence of silicon oxides.

The CV curves of the Si/RC are displayed in Figure 7a. The Si/RC reveals clear reduction peaks at 0.67 and 0.25 V during the first cycle. The reduction peak at 0.25 no longer appears in subsequent cycles, associating with the formation of an irreversible SEI film during the discharge process [47]. While the reduction peak at 0.67 still presents, corresponding to the reduction of Si to Li_xSi . The oxidation peak at 0.54 V during the charging corresponds to the process of Li^+ removing from the Li_xSi alloy and the formation of amorphous Si [48]. The oxidation peak observed at 0.98 V relates to the extraction of Li from the carbon [49]. Another weak peak at 2.11 V may be related to the presence of

N, which comes from BC materials [50]. Since Si is gradually activated during the charge–discharge process, the intensity of the oxidation peak at 0.54 V gradually enhances as the cycle number increases. As is shown in Figure 7c, the trend and the shape of the CV curves of the Si/AC are essentially the same as those of the Si/RC, except for a peak shift and slightly different peak intensities. The potential at which the reduction peak appears in Si/RC (0.67 V) is slightly lower than that in Si/AC (0.71 V), which may be attributed to the fact that the content of silicon in Si/RC is higher than that in Si/AC (conformed by XRD and Raman).

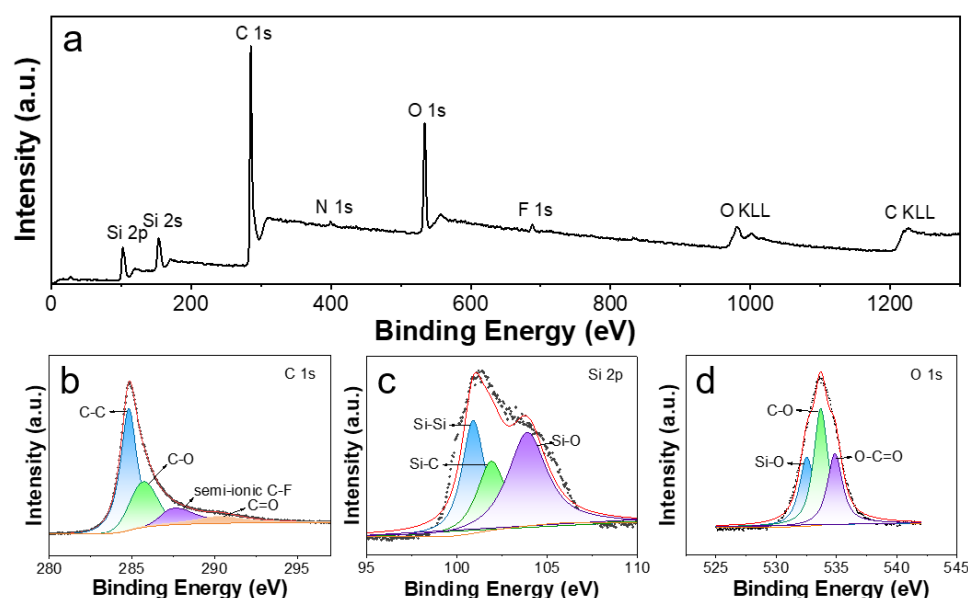


Figure 6. XPS spectra of Si/RC (a) and high-resolution: C 1s (b), Si 2p (c), and O 1s (d).

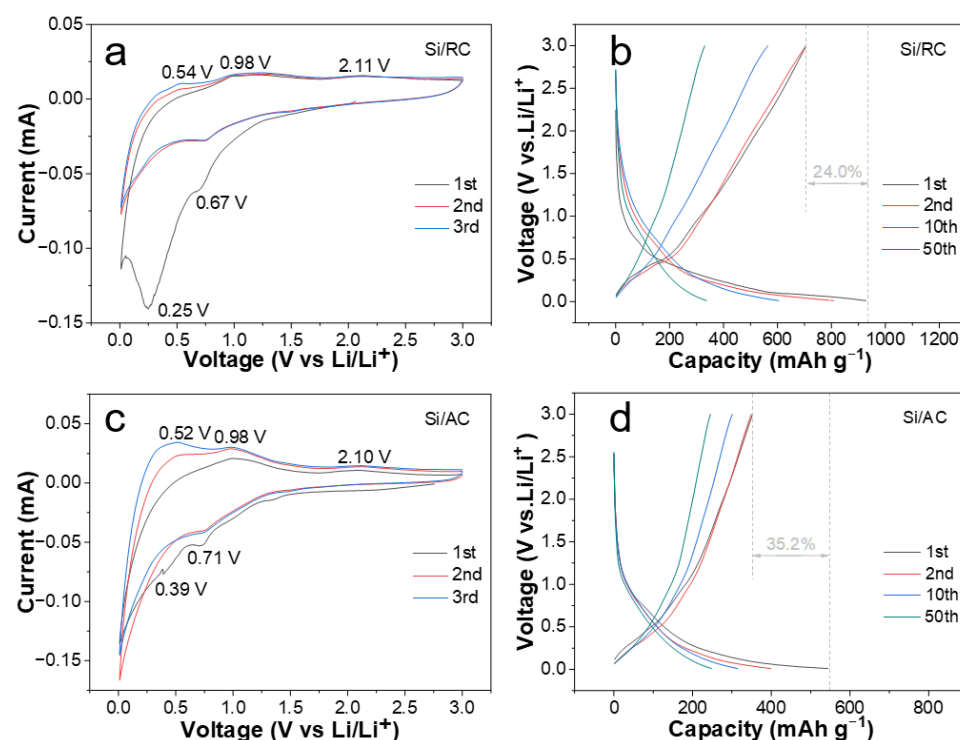


Figure 7. CV curves measured at 0.2 mV s^{-1} between 0.01 and 3 V: (a) Si/RC and (c) Si/AC; GCD curves measured at 200 mA g^{-1} for the LIBs with the anode material of Si/RC (b) and Si/AC (d).

Figure 7b,d presents the GCD curves of the Si/RC and Si/AC electrodes at 200 mA g^{-1} for different cycle numbers. During the first discharge process, a slight discharge plateau is observed in the voltage ranges of 0.6–0.7 V and 0.25–0.3 V for the Si/RC while 0.7–0.8 V for the Si/AC, which matches the peak potentials of the CV curves. The initial discharge/charge specific capacity of the Si/RC electrode is $928.3/705.9 \text{ mAh g}^{-1}$ with a capacity loss efficiency of 24.0%, and that of the Si/AC electrode is $544.0/352.6 \text{ mAh g}^{-1}$ with a capacity loss efficiency of 35.2%. In the following cycles, the capacity of the anodes decays rapidly, and the coulombic efficiency (CE) for the Si/RC and Si/AC anodes gradually increases to 97.9% and 98.8%, respectively, after 50 cycles, which indicates that they have relatively good stability.

The cyclic performance of the Si/RC and Si/AC anodes cycling at 200 mA g^{-1} is displayed in Figure 8a. In the first cycle, the Si/RC anode delivers a higher discharge-charge capacity of $928.3/705.9 \text{ mAh g}^{-1}$ than the Si/AC, corresponding to a CE of 76%. This is because during the first charge/discharge process, the electrolyte decomposes irreversibly to form the SEI film, and this process expends a mass of Li ions, resulting in a low CE for the first cycle. The capacity of the Si/RC anode begins to decay rapidly until it drops to 337.4 mAh g^{-1} after 50 cycles. This value is higher than that of the Si/AC at 248.1 mAh g^{-1} . After cycling for 100 cycles, the Si/RC anode delivers a reversible capacity of 318.4 mAh g^{-1} , while the Si/AC anode only remains 225.2 mAh g^{-1} . With the increase in cycle numbers, the CE of the Si/RC and Si/AC are in the range of 98.5%~99.2%, which indicates that the cycling stability of them is relatively good. The cycling performance of the Si/RC and Si/AC anodes cycling at 1 A g^{-1} is presented in Figure 8b. After 1000 cycles, the Si/RC anode shows a remaining capacity of 229.3 mAh g^{-1} , while the Si/AC anode has only 93.9 mAh g^{-1} left, indicating that the Si/RC has a better long-cycling property at a relatively high current intensity.

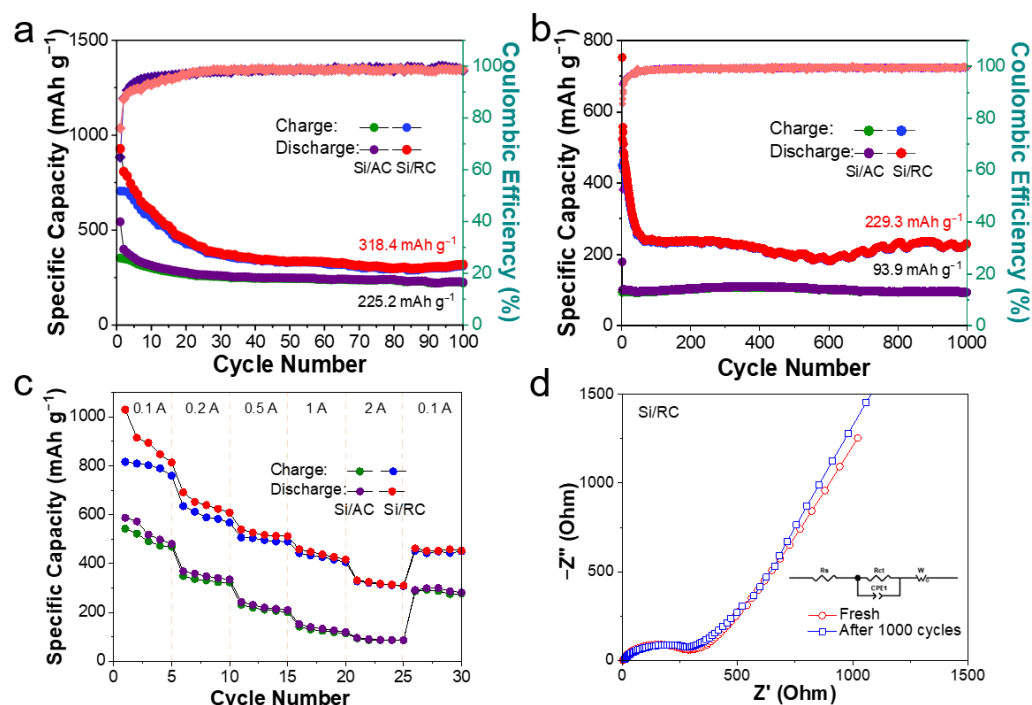


Figure 8. Cyclic performance of Si/RC and Si/AC anodes at 200 mA g^{-1} (a) and 1 A g^{-1} (b). (c) Rate performance of Si/RC and Si/AC anodes at different current densities. (d) Nyquist plots for Si/RC anode before and after 50 cycles.

The rate performance of the Si/RC and Si/AC is presented in Figure 8c, where the current density increases from 100 mA g^{-1} to 2 A g^{-1} and then back to 100 mA g^{-1} . The

initial capacity of the Si/RC and Si/AC anodes at 100 mA g^{-1} are slightly higher than that at 200 mA g^{-1} . A decreasing capacity is observed in the initial cycles at 100 mA g^{-1} , which is also found in the initial cycles at 200 mA g^{-1} (Figure 8a). It is generally accepted that the capacity decreases with increasing current density due to the kinetic limitations of the cell. When the current density reaches 2 A g^{-1} , the capacity of the Si/RC and Si/AC anodes presents 331.2 and 95.8 mAh g^{-1} , respectively. And when the current density is returned to 100 mA g^{-1} again, the reversible capacity of the Si/RC and Si/AC is 462.0 and 291.0 mAh g^{-1} , respectively.

EIS measurements were conducted to investigate the impedance of the Si/RC electrode in the frequency range of $0.01\text{--}10^5 \text{ Hz}$. The Nyquist plots before and after the Si/RC electrode cycling are shown in Figure 8d. These curves consist of semicircles in the high frequencies and slanted straight lines in the low frequencies, which correspond to the charge transfer resistance (R_{ct}) at the electrode/electrolyte interface and Li^+ diffusion in the electrode, respectively [50]. As shown in Figure 8d, there is no significant difference between the semicircle in the high-frequency region and the slanted straight line in the low-frequency region of the curves before and after cycling of Si/RC, which indicates that the stability of the Si/RC electrode is good during the cycling process.

After the addition of Si, the two materials show an obvious difference in electrochemical properties, which is mainly from two aspects. The first aspect, in terms of carbon structure, is that RC has a higher specific surface area than AC, and macropores and mesopores are uniformly distributed in large areas, forming better electron transport paths and ion transfer channels than AC, thus promoting the formation of better electrochemical kinetics. The second aspect is that the distribution and morphology of Si is quite different in the two carbon materials. Since RC has richer and larger-sized macropores, Si spheres can be embedded in these pores after formation. In contrast, AC has smaller macropores, and the Si spheres are difficult to embed in the macropores and thus accumulate on the surface of the carbon material. This large area of silicon accumulation causes drastic volume changes during the electrochemical cycling, thus deteriorating the electrochemical properties of the material. In summary, the obvious difference between the two materials in the porous structure of carbon and the distribution of silicon results in significant differences in electrochemical properties.

Table 1 reveals the electrochemical performance of the biomass-derived carbon materials used as anodes for LIBs in this work and some of the previously published ones. It can be seen from the table that the electrochemical property of the currently developed materials is at the upstream level in the field, showing a good competitive advantage. Firstly, the presence of many mesopores and channels on the RC surface can increase the specific surface area and effectively improve the electron transport capacity and ionic transportability of carbon materials. Secondly, the macropores in the RC matrix facilitate the preservation of Si within the carbon matrix instead of stacking on the carbon surface. Thirdly, Si/RC retains the porous structure of the precursor, which can effectively alleviate the volume expansion of Si during the charging/discharging process and improve the cycling stability. Last but not least, the introduction of an appropriate proportion of Si effectively improves the specific capacity of the composite. In summary, the bimodal porous carbon structure (both macropores and mesopores) and the appropriate loading of Si spheres inside the macropores of carbon have become the main reasons for the current material to obtain good lithium storage performance.

Table 1. Comparison of Li storage property of the as-prepared material and the previously reported studies.

Materials	Source	Current Density (mA/g)	Cycle Number	Capacity (mAh/g)	Ref.
LiFeO ₄ @C	Potato peel	0.1C	100	150.4	[51]
N doping carbon	Corn starch	0.2C	100	273	[52]
Porous carbon	Banana peel	0.2C	200	272	[53]
LFP@NFPC-2	Bacterial residues	1C	200	161.2	[54]
Honeycomb-AC	Argan shells	50	200	220	[55]
Activated carbon	Cotton stalks	100	100	271.7	[56]
Activated carbon	Avocado seeds	100	100	400	[57]
Spherical mesoporous carbon	Waste green tea	100	100	498	[58]
Porous carbon	Bagasse	2000	200	198.7	[59]
Si/C composites	Reed catkins	0.2C 1C	100 1000	318.4 229.3	This work

The further optimization of Si/C composite can be studied in two stages. In the first stage, the current Si/C composite preparation process can be optimized. In terms of carbon, by changing the pretreatment conditions, the structure of the pores can be adjusted to increase the size and/or the volume ratio of the macropores and improve the connectivity between these pores to attach as much Si as possible. In terms of Si, the craft conditions can be adjusted to further reduce the size of the Si spheres. This allows as much Si as possible to be attached into the macropores of carbon rather than on the carbon surface. At the same time, reducing the size of the Si spheres is more conducive to alleviating the volume change of the material. In the second stage, based on the results of the first stage, C and Si can be further optimized, respectively. For the Si spheres, a proper process can be chosen to encapsulate the Si with a C layer or doping Si with conductive elements. For the C skeleton, N, P, S, and other elements can be doped to further enhance the conductivity of the material and improve the electrochemical reaction kinetics.

4. Conclusions

In this work, the Si/RC and Si/AC composites are prepared by using RC and AC as the precursors, respectively, and utilizing the magnesium thermal reduction method. The experimental results show that the Si/RC exhibits relatively better electrochemical performance as an anode for LIBs compared with Si/AC. After cycling for 100 cycles, the Si/RC electrode delivers a reversible capacity of 318.4 mAh g^{−1} at 200 mA g^{−1}. When cycling at 1 A g^{−1}, the Si/RC electrode reveals a reversible capacity of 229.3 mAh g^{−1} after 1000 cycles. The enhanced cycling stability of Si/RC can be attributed to its special porous skeleton structure and more reasonable distribution of Si spheres. It is found that the structural features of BC precursors are important factors affecting the electrochemical properties of Si/C composites. Optimizing the structure of BC precursors is important for the development of high-performance BC-based composites. This work provides a reference for the application of low-cost biomass materials in the field of energy storage, which is conducive to promoting the popularization and application of BC-based composites as anode for LIBs.

Author Contributions: Conceptualization, H.X. and Z.W.; data curation, Z.M. and H.L.; formal analysis, Z.M. and Z.X.; funding acquisition, H.X. and Z.W.; investigation, Z.M. and Z.X.; methodology, C.Q.; project administration, C.Q. and Z.W.; resources, X.L. and C.Q.; supervision, X.L.; validation, H.L. and H.X.; visualization, Z.X. and H.L.; writing—original draft, Z.M.; writing—review and editing, X.L. and Z.W. All authors have read and agreed to the published version of the manuscript.

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Data Availability Statement: The original contributions presented in this study.

Conflicts of Interest: The authors declare no conflicts of interest.

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