

Understanding carbon-based electrode architectures for all-solid-state lithium–sulfur batteries

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Research Article

Understanding carbon-based electrode architectures for all-solid-state lithium–sulfur batteries

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Abstract

All-solid-state lithium–sulfur (Li-S) batteries offer improved safety, energy density, and cost efficiency. However, the electronically and ionically insulating nature of S necessitates substantial incorporation of conductive carbon and solid electrolyte. To enhance S utilization, it is essential to understand how effective phase boundaries govern the electrochemical behavior of S composite electrodes. In this study, the relationship between carbon surface area and its spatial distribution within the S composite electrode is systematically examined. S electrodes incorporating Ketjen Black exhibit largely improved utilization of $\text{Li}_6\text{PS}_5\text{Cl}$ and S, along with enhanced electron and ion transport at higher active-material ratios compared with those using acetylene black and SuperP. Galvanostatic intermittent titration technique and electrochemical impedance spectroscopy analyses further confirm that S redox kinetics are facilitated in composite electrodes containing Ketjen Black. As a result, optimized electrodes with 40 wt. % S demonstrate improved rate performance even at low stack pressure (700 mAh g_s^{-1} at 1C under both 75 MPa and 10 MPa) and stable cycling (1000 cycles at 1C under 75 MPa and 350 cycles at 0.2C under 10 MPa). This study highlights the critical role of carbon architecture in forming effective interphases that enable high S utilization and improved cycling and rate performances in all-solid-state Li–S batteries.

Introduction

All-solid-state Li-S batteries (ASSLSBs) are considered a promising next-generation battery platform due to high specific energy (up to 600 Wh kg⁻¹) and cost efficiency. [1, 2]. Sulfur (S) offers a superior theoretical specific capacity of 1672 mAh g⁻¹ compared with conventional oxide-based positive electrode materials (*e.g.*, LiNi_xCo_yMn_(1-x-y)O₂), < 278 mAh g⁻¹), and it is an earth-abundant by-product of fossil-fuel processing with an extremely low elemental cost (\$0.25 kg⁻¹ as of May 2026). [3, 4] Furthermore, pairing S with a solid electrolyte (SE) has additional merits to eliminate polysulfide dissolution and shuttle effects, which remain major challenges in liquid-electrolyte Li-S batteries.[5–7]

Despite these advantages, realizing high S utilization in ASSLSBs requires careful electrode design. S is electronically insulated (σ_e : $\sim 10^{-30}$ S cm⁻¹) which necessitates a substantial amount of conducting agents, often carbon-based materials, to establish effective electronic percolation networks. [8, 9] In addition, ~ 80 wt. % volume change between S₈ and Li₂S induces chemo-mechanical degradation that disrupts interfacial contacts. To improve the electronic conductivity along with consistently maintained contacts, S composite design has been explored by incorporating iodide into S (S_{9.3}I, $\sim 5.9 \times 10^{-7}$ S cm⁻¹) [10], integrating carbon-based conductive frameworks [11, 12], and employing transition metal sulfides (FeS₂, MoS₂, WS₂). [13–15] Furthermore, the intrinsically low Li⁺ diffusion coefficient (D_{Li^+} : 10^{-31} - 10^{-110} m² s⁻¹) of S requires to be supported by the use of solid electrolytes with higher ionic conductivity. [16] Among inorganic SEs, sulfide-based electrolytes such as β -Li₃PS₄ (β -LPS), Li₇P₃S₁₁ (LPS), Li₆PS₅Cl (LPSCl), Li₁₀GeP₂S₁₂ (LGPS) provide high ionic conductivity (> 3 mS cm⁻¹) and low bulk densities (~ 2 g cm⁻³), and good chemical and electrochemical stability with S. [6, 17] Although their narrow electrochemical windows lead to electrolyte decomposition, the resulting products (intermediate: Li₃PS₄, reduced species: Li₂S, Li₃P and oxidized species: elemental S and P₂S₅) are redox-active and can contribute additional capacity. [18, 19] Particularly, LPSCl has shown highly reversible redox behavior with S cathode, thereby motivating efforts to leverage its reactivity to further enhance capacity. [20–22]

Given that electrochemical reactions require the concurrent transport of electron and Li⁺ at the active material interface, effective utilization of S and SEs critically depends on the formation of stable and conductive triple-phase boundaries. Accordingly, prior studies have investigated various material modifications and compositional strategies. Approaches to improve performance include tailoring S particle size to the micron scale for enhanced rate capability and long-term performance [22], engineering sulfide-based solid electrolytes through titanium doping to enhance SE electronic conductivity [23], and employing low-density SE (< 1.5 g cm⁻³) to increase volumetric coverage. [24] In parallel, conductive-additive strategies have been investigated to tailor the architecture of S composite electrodes by integrating indium metal to reinforce interfacial contacts [25], studying the aggregation sensitivity of carbon to understand S-C network formation [26], and blending different carbon species to enhance the capacity utilization from both S and SE. [27] However, a systematic understanding of how carbon surface area and its spatial distribution influence the binary carbon–LPSCl interphase and the resulting triple-phase interactions remains limited. It is essential to identify how electronic and ionic pathways develop within the composite and to establish rational design principles for high-performance S electrodes.

In this work, we present a comprehensive study of the influence of carbon architecture on interphase formation with LPSCl and S using representative nanocarbon materials, Ketjen Black (KB), acetylene black (AB) and Super P (SP), which exhibit markedly different specific surface areas and aggregation properties. Based on the investigation of electronic and ionic conductivities across varying composite compositions and stack pressures, we demonstrated that the KB-based composite electrode with 40 wt. wt. % S achieved highly stable performance even at low stack pressure. The improved rate capability leads to capacity of 700 mAh g_s⁻¹ at 1C under both 75 MPa and 10 MPa, and the enhanced capacity utilization at 1C extended the stable cycle life to 1000 cycles for 75 MPa and 350 cycles for 10 MPa. To clarify the underlying mechanisms, galvanostatic intermittent titration (GITT) and in situ electrochemical impedance spectroscopic (EIS) analyses were conducted, revealing improved interfacial stability and enhanced charge-transfer and ion-diffusion kinetics in KB-based S electrodes. Surface characterizations further corroborated the higher S utilization with KB, as evidenced by the evolution of S and Li₂S signals. Overall, this study elucidated how carbon architecture governs effective triple-phase formation in S composite electrodes and provided insights into the electrode design for high-performance solid-state Li-S batteries.

Results

Carbon Architecture–Dependent LPSCI utilization

The structural differences between Ketjen Black (KB), acetylene black (AB) and Super P (SP) were characterized by Brunauer–Emmett–Teller (BET) analysis (**Fig. 1a, b and Supplementary Fig. 1**). KB exhibited significantly higher N_2 uptake across the entire pressure range compared to AB and SP (**Fig. 1a and Supplementary Fig. 1a and c**). Consistently, KB showed a remarkably larger mesoporous surface area of $1474.9 \text{ m}^2 \text{ g}^{-1}$, whereas AB and SP exhibited much lower values of $76.3 \text{ m}^2 \text{ g}^{-1}$ and $87.1 \text{ m}^2 \text{ g}^{-1}$, respectively (**Fig. 1b and Supplementary Fig. 1b and d**). In addition, Raman spectroscopy and X-ray photoelectron spectroscopy (XPS) were conducted to examine structural defects and surface chemistry, respectively (**Supplementary Figs. 2 and 3**). Raman spectra showed no substantial differences in the I_D/I_G ratios (1.05 for KB, 1.02 for AB, and 0.98 for SP), while XPS analysis revealed similar surface chemistry among the three carbons.

To investigate the carbon architecture-dependent interphase development with LPSCI, composites were prepared with KB, AB and SP and electrochemically evaluated across different compositions (**Fig. 1c-e and Supplementary Fig. 4 and 5**). KB-based composites exhibited higher LPSCI utilization across all compositions with lower voltage hysteresis compared to AB- and SP-based composites. Notably, KB showed a non-linear dependence on carbon content, whereas AB and SP exhibited rapid capacity fading with decreasing carbon ratio (**Fig. 1f**). Electronic and ionic conductivity measurements revealed that electronic conductivity strongly depends on carbon content, whereas ionic conductivity remains within a similar range (**Fig. 1g, Supplementary Fig. 6 and 7, Table S1 and S2**). This trend accounts for the rapid decrease in LPSCI utilization observed in AB- and SP-based composites. However, the KB-based composites deviate from this trend, as their LPSCI utilization does not follow the changes of electronic conductivity. For comparison, the electronic conductivities of the individual carbon materials were also measured at 75 and 10 MPa (**Supplementary Fig. 8**). KB, AB, and SP exhibited comparable electronic conductivities at both pressures. This discrepancy indicates that the observed LPSCI utilization behavior cannot be explained by bulk electronic transport alone but rather arises from the formation of electronically effective interphases, as shown in the **Supplementary Fig. 9**. Due to the high surface area of KB, excess KB at a low LPSCI/C ratio (2.5/1) leads to inactive carbon domains and a reduced effective interphase. As the LPSCI/C ratio increases, the electronically effective interphase reaches its maximum at 4/1 and decreases at 5/1, although it remains larger than that at 2.5/1 (**Supplementary Fig. 9a**). To quantitatively evaluate the electronically effective interphase, the electrochemical impedance spectroscopy (EIS) data were analyzed using the extracted cathode charge-transfer resistance (R_c) and effective interphase capacitance (C_{eff}) obtained from the equivalent-circuit fitting (**Supplementary Fig. 10a, b, g, and h**). [28, 29] Consistent with the proposed interphase evolution, R_c decreased from (2.5/1) to (4/1) and slightly increased at (5/1), while C_{eff} followed the order of (4/1) > (5/1) > (2.5/1) (**Supplementary Fig. 10a and b**). The minimum in R_c and maximum in C_{eff} at (4/1) indicate the most efficient interphase formation and utilization at this composition. Although additional LPSCI was introduced at (5/1), the slight decrease in C_{eff} suggests that the extent of the electronically effective interphase was partially reduced while maintaining substantial interfacial contact.

In contrast, the effective interphase in AB- and SP-based composites progressively decreases with increasing LPSCI fraction compared to KB (**Supplementary Fig. 9b and c**). As observed in **Supplementary Fig. 11 and 12**, AB- and SP-based composites exhibited agglomerated carbon domains and higher porosity, with more severe agglomeration and inhomogeneous carbon distribution in SP. The EIS analysis further clarifies the distinct interphase and transport evolution in AB- and SP-based composites (**Supplementary Fig. 10a–c, e, f, and h**). For AB composites, C_{eff} continuously decreased from (2.5/1) to (5/1), indicating a progressive reduction in the electronically effective interphase with increasing LPSCI fraction (**Supplementary Fig. 10b**). Notably, AB (2.5/1) exhibited a larger C_{eff} and lower R_c than KB (2.5/1), representing more effective interphase formation when sufficient carbon was available. This observation is consistent with the capacity-utilization trend shown in Fig. 1f, where AB (2.5/1) exhibited higher LPSCI utilization than KB (2.5/1). However, unlike KB, the electronically effective interphase in AB continuously diminished as the LPSCI fraction increased, indicating limited tolerance to carbon depletion. R_c followed the order of (4/1) > (2.5/1) > (5/1) (**Supplementary Fig. 10a**). Although R_c decreased substantially at (5/1), this should not be interpreted as improved interphase utilization because finite-length transport processes emerged at this composition, as evidenced by the appearance of R_{ws} and T_{ws} (**Supplementary Fig. 10c, e, and h**). [30, 31] Here, R_{ws} represents the resistance associated with finite-length transport within the composite network, while T_{ws} reflects the characteristic relaxation time of this transport process. The emergence of these parameters suggests that the electronically effective interphase becomes spatially disconnected at AB (5/1), resulting in fragmented transport pathways and reduced transport length scales, as illustrated in **Supplementary Fig. 9b**. In this regime, charge carriers can no longer efficiently access the entire interphase network and instead experience finite-length transport through isolated electrochemically active domains. For SP, C_{eff} continuously decreased while R_c increased with increasing LPSCI fraction, indicating progressive loss of the electronically effective interphase accompanied by increasingly hindered charge-transfer processes (**Supplementary Fig. 10a and b**). Moreover, finite-length transport features emerged already at SP (4/1), indicating an earlier disruption of the transport network compared to AB (**Supplementary Fig. 10c, f, and h**). Furthermore, increasing R_{ws} together with decreasing T_{ws} from SP (4/1) to SP (5/1) indicate progressively hindered transport through increasingly fragmented pathways with reduced effective transport length scales, as illustrated in **Supplementary Fig. 9c**.

On the other hand, ionic conductivity of the composite follows a different trend (KB > SP > AB), with relatively small differences among the three composites. This behavior can be elucidated by considering both the ionic conductivity, surface area, and microstructure of the carbons. Among the carbons, SP exhibited the highest ionic conductivity (SP > AB > KB, **Supplementary Fig. 13**). However, ionic transport in the composites is additionally influenced by the LPSCI-carbon interphase, including surface-mediated Li^+ transport associated with carbon surface area (1474.9 $\text{m}^2 \text{g}^{-1}$ for KB, 87.1 $\text{m}^2 \text{g}^{-1}$ for SP, 76.3 $\text{m}^2 \text{g}^{-1}$ for AB) [32] and carbon distribution within the composite (**Supplementary Fig. 11**). These combined contributions result in the relatively small differences in ionic conductivity among the LPSCI-carbon composites. Long-term cyclability at 0.2C (**Fig. 1h**) further highlighted the effect of carbon architecture. The capacity utilization of AB and SP composites was significantly reduced with increasing LPSCI fraction from (2.5/1) to (4/1). In contrast, the

KB-based composites at (4/1) and (5/1) maintained high capacities of $\sim 135 \text{ mAh g}^{-1}$ and $\sim 130 \text{ mAh g}^{-1}$ with stable capacity retention. Notably, the KB composites at (2.5/1) delivered lower capacity than AB and SP at (2.5/1), further indicating that optimal performance requires a balanced LPSCI-to-carbon ratio.

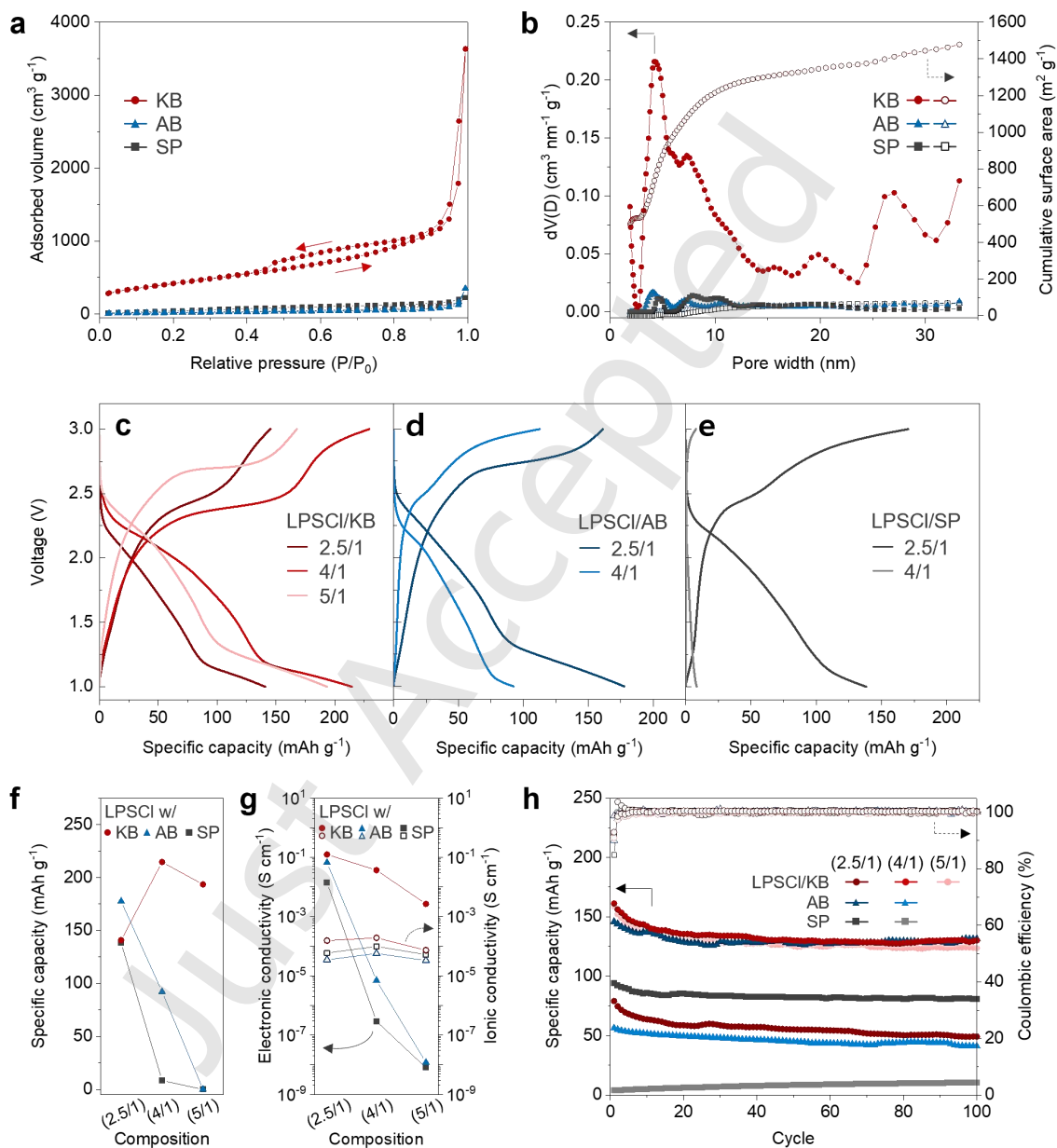


Fig. 1. **a** N₂ adsorption and desorption isotherm, and **b** Pore size distribution and cumulative surface area of KB, AB and SP. Voltage profiles of LPSCI composite with different ratios of KB, AB and SP at 0.05C ($1C = 2.1 \text{ mA cm}^{-2}$): **c** KB. **d** AB. **e** SP. **f** Specific discharge capacities comparison at each composition. **g** Electronic and ionic conductivities of LPSCI composite at each composition. **h** Long-term cyclability at 0.2C ($1C = 2.1 \text{ mA cm}^{-2}$). Stack pressure: 75 MPa; testing temperature: $25 \pm 1^\circ \text{C}$.

Carbon Architecture–Dependent Electronic and Ionic Transport in S Composite Electrodes

To extend the understanding of carbon architecture–dependent interphase behavior to actual S composite electrodes, the charge transport characteristics through the triple-phase boundary were evaluated by measuring electronic and ionic conductivities at different S ratios under stack pressures ranging from 75 to 10 MPa (**Fig. 2a and b, Supplementary Fig. 14 and 15, Table S3 and S4**). The S composites with commonly utilized compositions were prepared by fixing the LPSCl ratio at 50 wt. % and varying the S ratio from 30 wt. % to 40 wt. % (corresponding to carbon ratios of 20 wt. % and 10 wt. %). [21, 27, 33–36] In the electronic conductivity measurements, all KB-, AB- and SP-based S composites exhibited comparable values in order of $KB > AB > SP$ at the 30 wt. % of S. However, when the S ratio increased to 40 wt. %, AB and SP showed significantly larger degradation of 1.5 and 7 orders of magnitude respectively, whereas KB showed only a half-order decrease (**Fig. 2a**). Meanwhile, the stack-pressure-dependent decrease in electronic conductivity was approximately half an order of magnitude across all S composites.

For ionic conductivity, all S composites at each ratio remained within a comparable order of magnitude, showing much smaller degradation with decreasing the stack pressure (**Fig. 2b**). Moreover, the ionic conductivity of S/C/SE composites followed the same trend observed in LPSCl/C systems ($KB > SP > AB$). Given the comparable ionic conductivity at each S ratio, it could be inferred that the electronic network predominantly influences the capacity utilization of S composite electrodes. These carbon ratio-dependent transport behaviors are further rationalized in terms of effective electronic interphases (**Fig. 2c**). At high carbon ratio, intimate triple-phase boundaries among S, carbon and LPSCl are maintained, forming electronically effective interphases that follow the order of $KB > AB > SP$. However, at low carbon ratio, the composite becomes highly sensitive to the integrity of the carbon network. KB sustains significantly higher electronic conductivity due to its larger surface area and well-distributed particle network, which supports a highly interconnected electronic percolation. In contrast, AB and SP lose not only carbon-carbon interconnectivity but also electronic interphase contact with S, with this effect being even severe in SP due to its more agglomerated carbon domains.

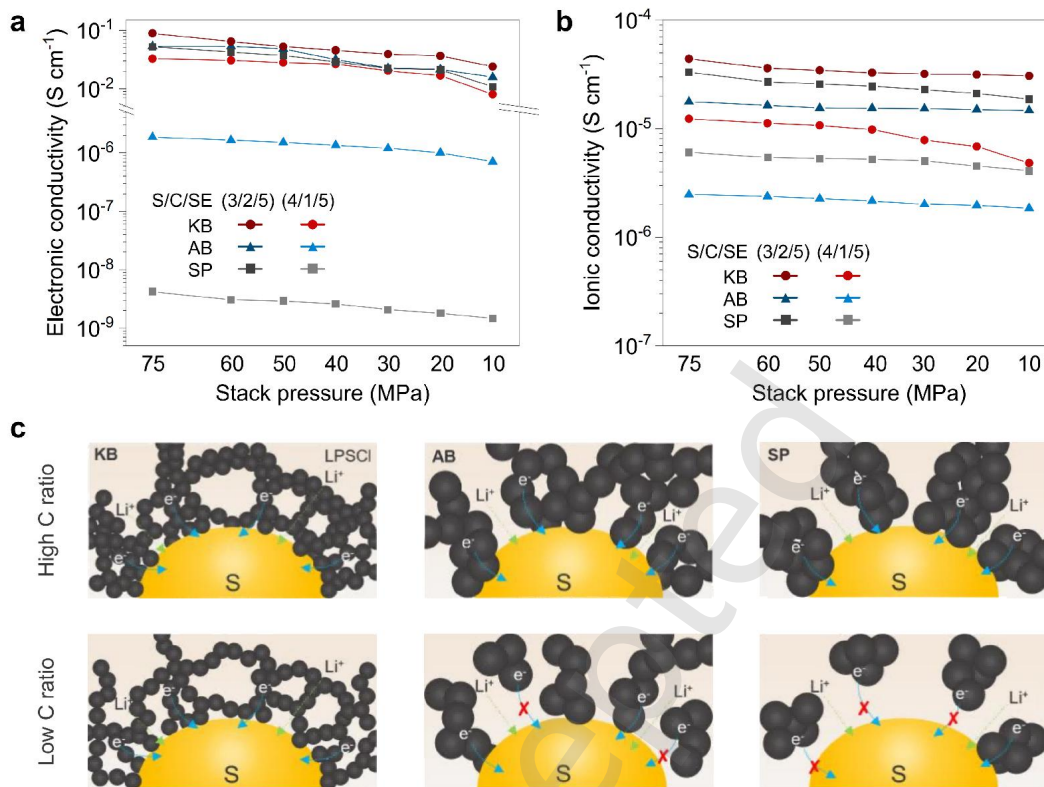


Fig. 2. Electronic and ionic conductivity measurement of S composite electrode with S ratios of 30 wt. % and 40 wt. % under stack pressures ranging from 75 to 10 MPa. **a** Electronic conductivity. **b** Ionic conductivity. Testing temperature: $25 \pm 1\ ^\circ C$. **c** Schematic illustration of the carbon percolation network at different carbon ratios.

Rate Capability and Long-term Cycling performance

The influence of carbon architecture on S utilization was further validated through the cell tests under dynamic cycling conditions. In the rate capability tests, KB-based electrodes delivered higher capacities with suppressed voltage hysteresis across all C rates, whereas AB- and SP-based electrodes exhibited substantially lower capacities with higher voltage hysteresis (**Fig. 3a-d and Supplementary Fig. 16**). Within the KB-based electrodes, the capacity retention increased from 60 % (30 wt. % S) to 86 % (40 wt. % S), indicating that excessive carbon at 30 wt. % S leads to inefficient electron-transport network, whereas at 40 wt. % S a well-maintained carbon network is preserved even at high C rates. To further investigate the stability of the corresponding carbon networks, long-term cycling at high-rate of 1C was conducted (**Fig. 3e**). KB (S 30 wt. %) electrode showed rapid capacity loss at the early cycles and gradual recovery to 1065 mAh g⁻¹ (737th cycle) followed by capacity fading to 763 mAh g⁻¹ (1000th cycle), indicating dynamic interphase evolution associated with the redistribution and agglomeration of carbon aggregates.[26] Meanwhile, the KB (S 40 wt. %) electrode exhibited markedly stable capacity utilization of ~750 mAh g⁻¹ over 1000 cycles, representing durable interphases during repeated conversion reactions.

To further investigate the origin of the distinct cycling behaviors, EIS and distribution of relaxation times (DRT) analysis were performed before and after cycling (**Supplementary Fig. 17**). Both pristine and cycled KB (30 wt. % S) electrodes exhibited larger C_{eff} but substantially higher R_c than KB (40 wt. % S) that maintained substantially lower R_c (**Supplementary Fig. 17a-c**). The corresponding DRT spectra showed that the charge-transfer-related peak of KB (30 wt. % S) was shifted toward longer relaxation time with higher resistance compared with KB (40 wt. % S), indicating slower charge-transfer kinetics (**Supplementary Fig. 17d and e**). [37, 38] In addition, it decreased in half after cycling. DRT analysis during the early stages of cycling further revealed that the charge-transfer resistance markedly decreased within the first 200 cycles and subsequently increased toward the end of cycling (**Supplementary Fig. 18b**). Consistent with this dynamic evolution, the agglomerated carbon domains observed in the pristine electrode were redistributed after 50 cycles (**Supplementary Fig. 19**). These electrochemical and microstructural changes support that the capacity decay and subsequent recovery of KB (30 wt.% S) originate from the dynamic redistribution of the carbon network and corresponding interphase evolution during cycling. In contrast, KB (40 wt. % S) maintained relatively stable charge-transfer resistance during the initial 100 cycles, followed by the gradual increase upon further cycling (**Supplementary Fig. 18c**). Cross-sectional analysis also showed that the carbon distribution was maintained after 50 cycles (**Supplementary Fig. 20**). These results indicate the more stable carbon network and interphase, consistent with the stable long-term cycling. These results are also consistent with the superior rate retention of KB (40 wt. % S) observed in **Fig. 3a and b**. In contrast, AB- and SP-based electrodes exhibited limited capacity utilization, reflecting insufficient electronic percolation and poor interfacial contact.

At the reduced stack pressure of 10 MPa, the effect of carbon architecture became more pronounced. The rate capability tests showed that the KB-based electrodes delivered higher capacities with smaller voltage hysteresis than the AB- and SP-based electrodes at both compositions (**Supplementary Fig. 21a-f and 22**). Within the KB-based electrodes, the KB (40 wt. % S) electrode maintained both capacity and small voltage hysteresis across C rates,

resulting in higher utilization than that of KB (30 wt. % S) particularly above 0.4C. Furthermore, in terms of capacity retention, the KB (40 wt. % S) electrode was superior to KB (30 wt. % S) which showed rapid capacity decaying within the very initial cycles (Supplementary Fig. 21g). This again highlights the suppressed utilization arising from excessive carbon, which disrupts effective triple-phase formation under low-pressure conditions.

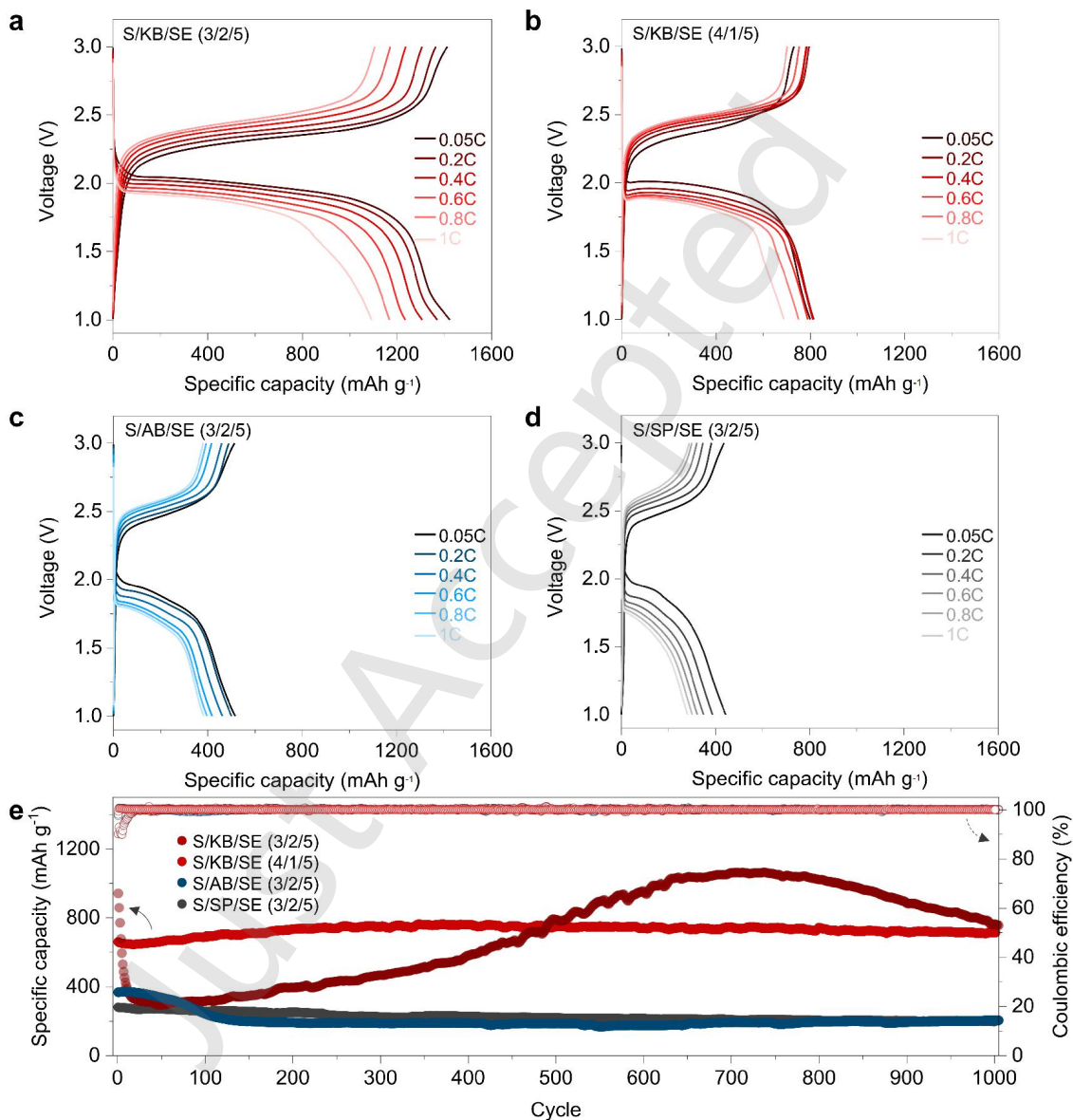


Fig. 3. LiIn||S cell performance. **a-d** Rate capability tests of S composite electrodes with KB, AB and SP at S ratios of **a, c and d:** 30 wt. % and **b:** 40 wt. %. **e** Long-term capacity retention at 1C (1C = 1.5 mA cm⁻²). Stack pressure: 75 MPa; Testing temperature: 25 ± 1 °C.

GITT and in situ EIS Characterization of Electron/Ion Transport

In addition to evaluating the interphase kinetics and stability during prolonged cycling, galvanostatic intermittent titration technique (GITT) immediately followed by in situ EIS analyses was conducted at a substantially lower current to elucidate the transport and interfacial characteristics during the conversion reaction (**Fig. 4 and Supplementary Fig. 23**). The GITT measurement showed a similar trend across all carbon species, with polarization gradually increasing as the conversion reaction progressed during both charge and discharge (**Fig. 4 a, c, e and g**). However, substantial differences in the polarization as well as S utilization between different S composite cathodes was observed. The KB-based electrode showed largely reduced voltage polarization at 30 wt. % S and comparable voltage polarization even at 40 wt. % S during each relaxation interval compared to AB- and SP-based electrode, indicating facilitated electron-ion transport through the well-developed triple-phase boundaries.

The DRT analysis derived from in situ EIS measurements provided further insight into the interfacial and transport characteristics (**Fig. 4b, d, f and h**). A common feature observed across all electrodes was that higher resistance values were obtained during charge than during discharge. In particular, the resistance increased during charge as the conversion reaction proceeded from Li_2S to S-rich states, while it decreased during discharge. This behavior was consistent with the lower electronic and ionic conductivity of S compared to Li_2S . [39] Additionally, the KB-based electrodes showed a gradual increase in both R_{ct} and R_{d} during cycling. Meanwhile, AB- and SP-based electrodes exhibited relatively high resistance values from the initial state, with less variation during cycling, indicating that less effective triple-phase interphases were formed from the beginning in these electrodes. Notably, the resistance values of the KB (40 wt. % S) electrode were comparable to those of the AB- and SP-based electrodes (30 wt. % S) over most of the conversion process. This demonstrates that the optimized carbon architecture in the KB-based composite effectively sustained coupled electron-ion transport during the reactions.

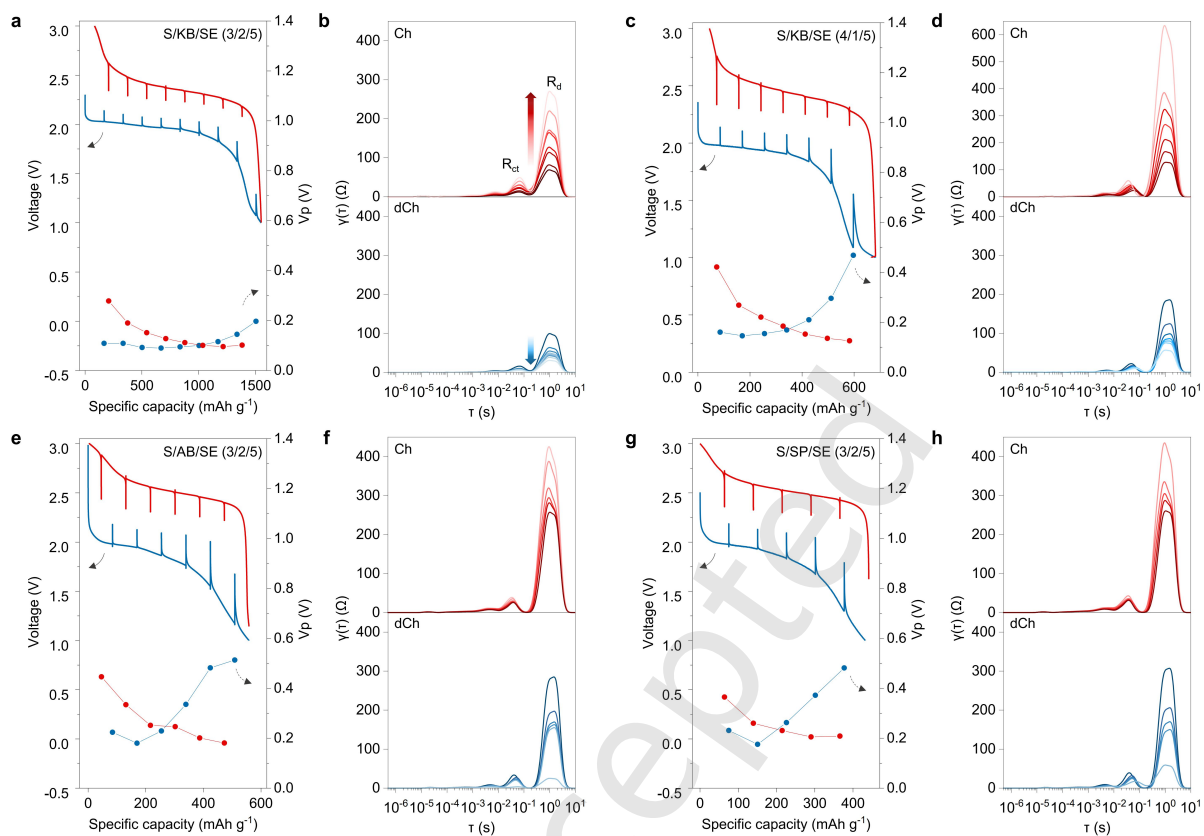


Fig. 4. Analysis of GITT and DRT on discharge and charge of S composite with KB, AB and SP. **a, c, e and g** Voltage profiles of galvanostatic pulses and open-circuited state and corresponding voltage polarization. **a:** KB (30 wt. % S). **c:** KB (40 wt. % S). **e:** AB (30 wt. % S). **g:** SP (30 wt. % S). **b, d, f and h** DRT spectra. **b:** KB (30 wt. % S). **d:** KB (40 wt. % S). **f:** AB (30 wt. % S). **h:** SP (30 wt. % S). Galvanostatic titration at 0.06 mA cm^{-2} for 2 hours; Relaxation at open circuit for 1 hours; Stack pressure: 75 MPa; Testing temperature: $25 \pm 1^\circ \text{C}$.

Structural and Chemical State Characterization of Discharged S Electrodes

The structural and chemical states of the discharged S composite electrodes at 40 wt. % S were examined to further clarify the architecture-dependent electrochemical behavior (**Fig. 5**). The 30 wt. % S electrodes were not used due to the dominant carbon signal in the characterization. Cross-sectional images and corresponding carbon mappings revealed clear structural differences (**Fig. 5a-c**). The KB-based electrode displayed uniform surface morphologies and well distributed S and KB particles, suggesting that the interconnected KB framework was largely preserved after the volume change associated with Li_2S formation. In contrast, the AB- and SP- based electrodes exhibited spatially heterogeneous surface, with aggregated carbon particles, indicating the inefficient triple-phase boundaries. As shown in **Fig. 5d**, X-ray diffraction patterns of the discharged electrodes revealed the clear disappearance of crystalline S peaks for the KB-based electrode, showing that S was largely consumed during discharge at the 40 wt. % S. However, the AB- and SP-based electrode retained distinct S peaks after discharge, demonstrating that a considerable fraction of S remained unreacted. The S 2p XPS spectra further provided further insight into the chemical state of the discharged products (**Fig. 5e**). [23, 40] The KB-based electrode showed clearly developed Li_2S peaks (160.2 eV and 161.3 eV) along with residual elemental S signals (163.5 eV and 164.4 eV), while the AB- and SP-based electrode exhibited the absence of Li_2S formation. Overall, these structural and chemical analyses demonstrated that the KB-based electrode at 40 wt. % S promoted more uniform Li_2S formation and maintained the electrode integrity necessary for efficient S utilization, whereas the AB- and SP-based electrode suffered from severe structural heterogeneity and incomplete conversion, leading to substantially limited electrochemical performance.

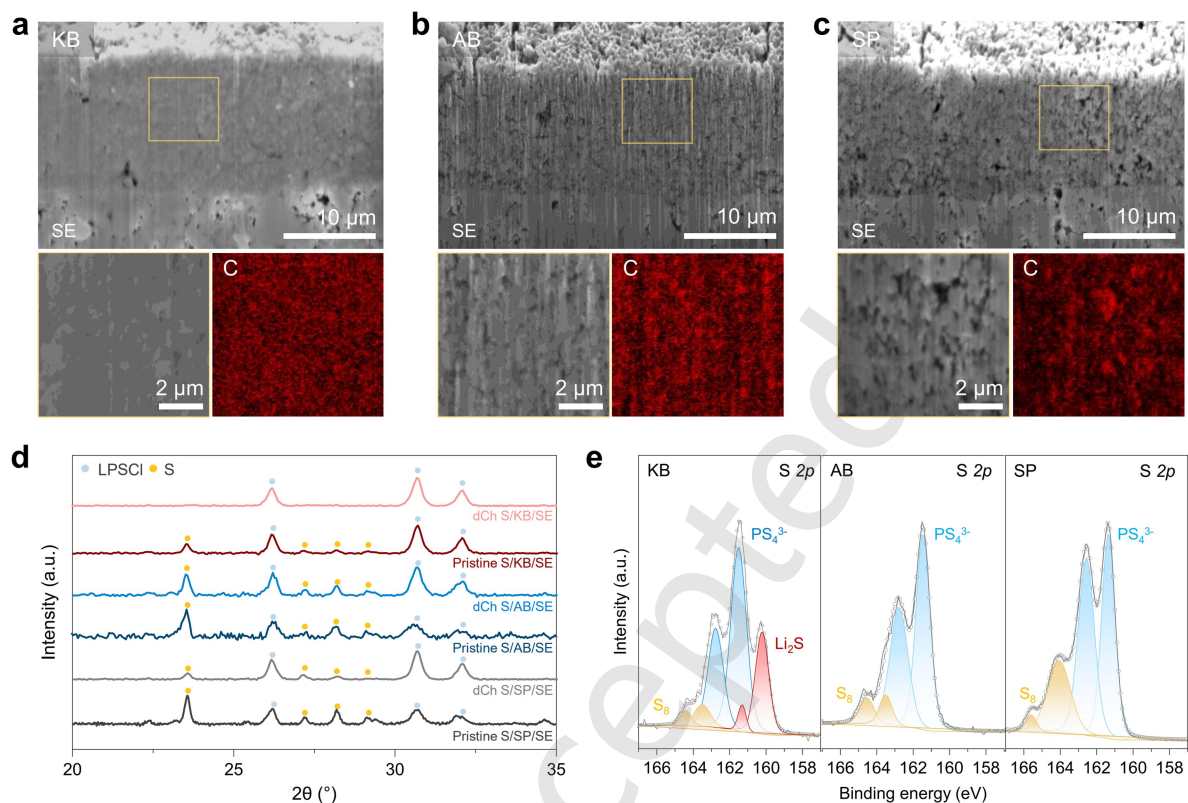


Fig. 5. **a-c** Cross-sectional morphology and elemental mapping images of discharged S composite electrodes at 40 wt. % S. **a:** KB. **b:** AB. **c:** SP. **d** X-ray diffraction pattern of pristine and discharged S composite with KB, AB and SP at 40 wt. % S. **e** S 2p X-ray photoelectron spectra of discharged S composite at 40 wt. % S. (Left) KB. (Middle) AB. (Right) SP.

Conclusion

In summary, this work showed that KB, with its larger surface area, enabled well-developed triple-phase interphases and enhanced utilization of both LPSCl and S at the high S ratio of 40 wt. % compared with AB and SP. KB-based electrodes at 40 wt. % S maintained superior rate capability even under low stack pressure (700 mAh g_s^{-1} at 1C under both 75 and 10 MPa), along with substantially reduced voltage hysteresis, and showed stable cycling over 1000 cycles at 1C under 75 MPa. GITT and DRT analyses further confirmed that the KB architecture suppressed polarization by facilitating charge-transfer and diffusion processes. Structural and chemical characterizations additionally exhibited preserved electrode integrity with uniformly distributed KB, as well as more Li_2S formation in the KB-based electrodes at 40 wt. % S. Overall, this study elucidates that engineering carbon architecture is a key factor in constructing electronically and ionically effective interphases and provides electrode design guidance for achieving high S utilization at elevated S ratios in all-solid-state Li-S batteries.

Methods

Materials preparation

S (100 mesh, Sigma-Aldrich), Li powder (SLMP[®], ≥ 97 wt. %, Livent), In powder (< 45 μm , 99.99 wt. %, MSE supplies), $\text{Li}_6\text{PS}_5\text{Cl}$ (LPSCl; Mitsui Kinzoku), Ketjen Back (EC-600JD, Nouryon), acetylene black (≥ 99.9 wt. %, Thermo Scientific) and SuperP (Timcal) were used as received. S was further ball-milled at 500 rpm for 10 hours with 1 minute rest time every 1 hour, using high-energy ball-milling machine (PBM-0.4A, Xiamen Tmax Battery Equipments Ltd.) and 50 mL air-tight zirconia ball-milling jars for the positive electrode composite. The ball-to-LPSCl mass ratio was 36:1, using 5 mm zirconia balls. All the ball-milling preparation was progressed in the glovebox (Vigor; $\text{O}_2 \leq 0.05$ ppm, $\text{H}_2\text{O} \leq 0.02$ ppm, Argon atmosphere).

Electrodes preparation

Li-In powder composite. The Li-In composite was prepared by mixing Li and In powder at the atomic ratio of 1 to 1 for 5 min at 1800 rpm using a vortex mixer (MI0101002D, Four E's Scientific) in the Ar-filled glovebox.

S positive electrode powder composite. The S powder composite was prepared by ball milling ball-milled S, LPSCl and carbons (KB, AB or SP) at 500 rpm for 10 hours with 1 minute rest time every 1 hour, using high-energy ball-milling machine (PBM-0.4A, Xiamen Tmax Battery Equipments Ltd.) and 50 mL air-tight zirconia ball-milling jars. The ball-to-LPSCl mass ratio was 36:1, using 5 mm zirconia balls. All the ball-milling preparation was progressed in the glovebox (Vigor; $\text{O}_2 \leq 0.05$ ppm, $\text{H}_2\text{O} \leq 0.02$ ppm, Argon atmosphere).

Solid-state cell assembly

The solid-state cells were assembled with two titanium plungers and one polyetheretherketone (PEEK) holder with diameter of 10 mm in the glove box. A SE layer was first prepared by pressing 70 mg of LPSCl at 310 MPa for 10 s. LPSCl composite (0.9 mg_{LPSCl} for LPSCl ratios of 71.4 wt. %, 80 wt. % and 83.3 wt. %) or S powder composite (0.9 mg_{S} cm^{-2} for both S ratios of 30 wt. % and 40 wt. %) was placed on one side of the layer and pressed at 375 MPa for 3 min, followed by LiIn negative electrode preparation (70 mg) on the opposite side by pressing at 125 MPa for 3 min. The assembled cells were rested under a stack pressure of 75 MPa or 10 MPa for 1 h before testing. Conductivity measurements were conducted using an ion-blocking cell for electron conductivity, an electron-blocking cell for ion conductivity, and a blank cell to determine the background resistance excluding the sample contribution (**Supplementary Fig. 24**). For electronic conductivity measurements, the corresponding powder was pressed at 3 tons for 3 min. For ionic conductivity measurements, LPSCl (70 mg) was first pressed at 3 tons for 3 min, followed by addition of the corresponding sample powder and pressing at 3 tons for 3 min. A second LPSCl layer (70 mg) was then added and pressed at 3 tons for 3 min. LiIn powder (30 mg) was placed at both ends and pressed at 1 ton for 10 s. For the blank cell, LPSCl (140 mg) was pressed at 3 tons for 3 min, followed by addition of LiIn powder (30 mg) at both ends and pressing at 1 ton for 10 s.

Electrochemical measurements

All battery cycling tests were performed on a potentiostat (VSP-300, BioLogic) and a galvanostat (BTS4000, Neware) inside the glove box at 25 ± 1 °C. Electronic and ionic conductivities were measured using DC polarization. For electronic conductivity measurements, a constant DC voltage of 50 mV was applied for 1 min, and the electronic conductivity was calculated according to

$$\sigma_e = \frac{L}{R_e A} = \frac{IL}{VA}$$

where L is the pellet thickness, A is the electrode area (0.785 cm^2), V is the applied voltage, I is the measured current, and R_e is the corresponding resistance. For ionic conductivity measurements, a constant DC voltage of 100 mV was applied for 1 h, and the steady-state current was used to determine the total cell resistance ($R_{\text{total}} = V/I_{\text{ss}}$). The blank-cell resistance (R_{blank}) was subtracted from the total cell resistance, and the ionic conductivity was calculated according to

$$\sigma_{\text{ion}} = \frac{L}{A(R_{\text{total}} - R_{\text{blank}})}$$

where L is the sample thickness and A is the electrode area (0.785 cm^2).

GITT was used to measure the voltage polarization development in open circuit for 1 h after galvanostatic titration at 0.06 mA cm^{-2} for 2 h during discharge to $1 \text{ V}_{\text{Li/Li}^+}$ and charge to $3 \text{ V}_{\text{Li/Li}^+}$ of S electrode with 30 wt. % S and 40 wt. % S ratios at stack pressure of 75 MPa. The potentiostatic electrochemical impedance spectra (PEIS) was measured immediately after each galvanostatic titration. PEIS circuits were obtained under the following conditions: at open-circuit voltage; frequency range from 7 MHz to 0.1 Hz; sinusoidal amplitude of 10 mV; and 6 data points per decade in Logarithmic spacing. DRT spectra of symmetric cells was converted from PEIS circuit using MATLAB. S cells with LiIn were galvanostatically charged up to $3 \text{ V}_{\text{Li/Li}^+}$ and then discharged to $1 \text{ V}_{\text{Li/Li}^+}$. Constant-voltage step at $3 \text{ V}_{\text{Li/Li}^+}$ was not included in the charging. For the rate capability test, the current density was ramped up from 0.05C to 1C at the stack pressure of 75 MPa and 10 MPa. Used conditions of the stack pressure were noted in corresponding figures. The applied stack pressure was measured and monitored by load sensor calibrated with Instron Loadframe.

Material characterization

The X-ray diffraction (XRD) patterns of the S composite electrodes were obtained at 2° min^{-1} (XRD; D2 Phaser, Bruker with Cu-Ka1 radiation). All electrode samples were prepared in heat-sealed bags inside the Ar-filled glovebox prior to transfer to the XRD chamber. The discharged state of S electrode with KB, AB and SP at 40 wt. % S ratio was investigated in S 2p by X-ray photoelectron spectroscopy (XPS; AXIS Supra, Kratos Analytical) with an Al monochromatic anode source at 15 kV with a 5.0×10^{-8} Torr vacuum level. The spot area for XPS measurements is $300 \mu\text{m} \times 700 \mu\text{m}$. All electrode samples were transferred in a heat-sealed bag inside the Ar-filled glovebox before they were transferred to the XPS glovebox. The XPS sample transfer is conducted under N_2 environment to avoid air-exposure. Survey scans were collected with a 1.0 eV step size, followed by high resolution scans with a step

size of 0.1 eV for S 2p. Etching using Argon1000+ gas cluster ion source (GCIS) accelerated at 10 keV was carried out to clean the surface and mill the electrode over a raster area of 2 mm $\times$ 2 mm with energy of 1000 keV for 15 s. The surface morphology of electrode was visualized by scanning electron microscopy (SEM; FEI Apreo, ThermoFisher Scientific). All electrode samples were prepared in heat-sealed bags inside the Ar-filled glovebox and immediately transferred to the SEM chamber after opening.

Just Accepted

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Author contributions

Y.J. and Z.C. conceived the idea and conceptualization. Y.J. contributed experiments, methodology, data curation, data analysis and investigation. J.W. and A.P. performed XPS characterization. R.E. performed Raman characterization. F.L., D.J.L., G.K., S.H.H. and J.B.L. contributed data analysis and discussion. Y.J. and Z.C. wrote the manuscript. Z.C. supervised the project. All authors discussed the results and commented on the paper.

Competing interests

Z.C., Y.J., D.J.L., G.K., S.H.H. and J.B.L. declare that a patent application on this work has been filed through UC San Diego's Office of Innovation and Commercialization and LG Energy Solution, Ltd (US 63/927,155). The remaining authors declare no competing interests.

Data availability

Source Data are provided with this paper.

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Understanding carbon-based electrode architectures for all-solid-state lithium–sulfur batteries

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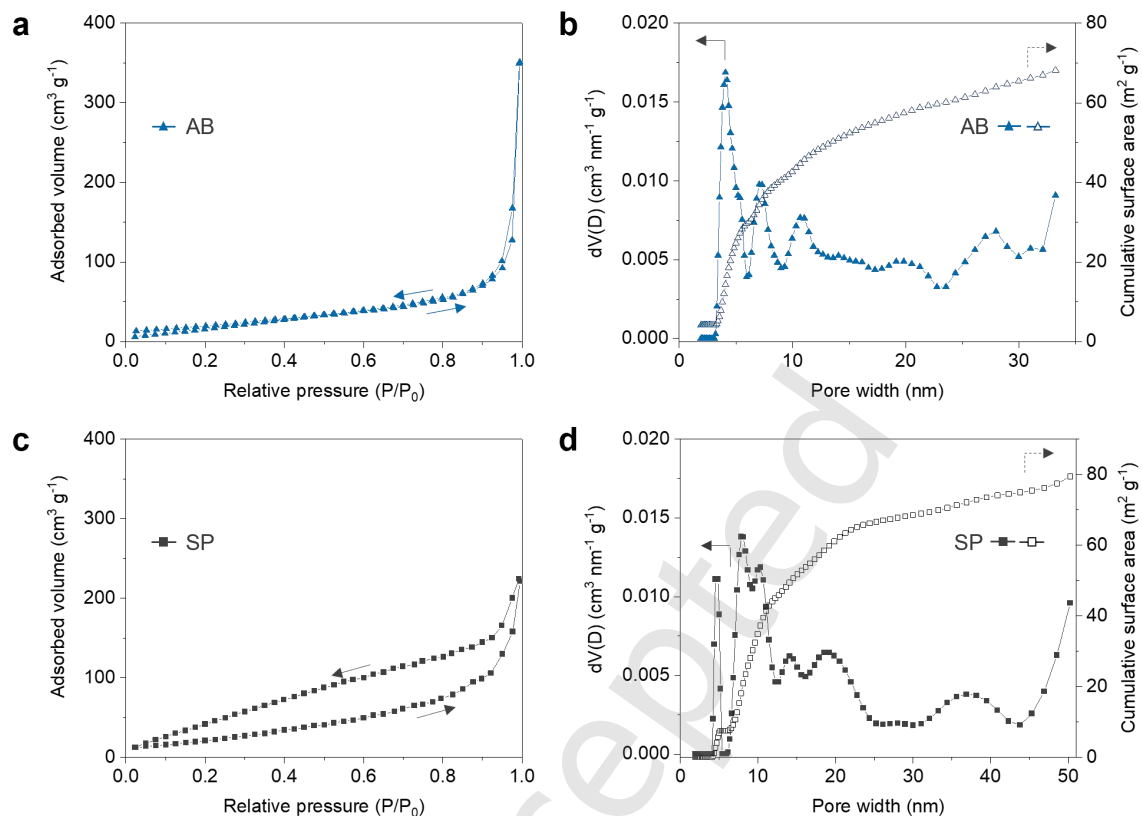
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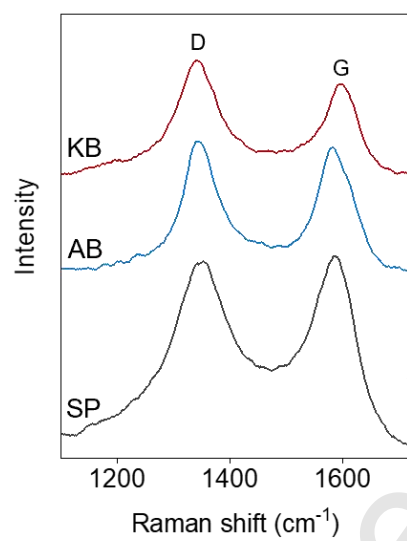
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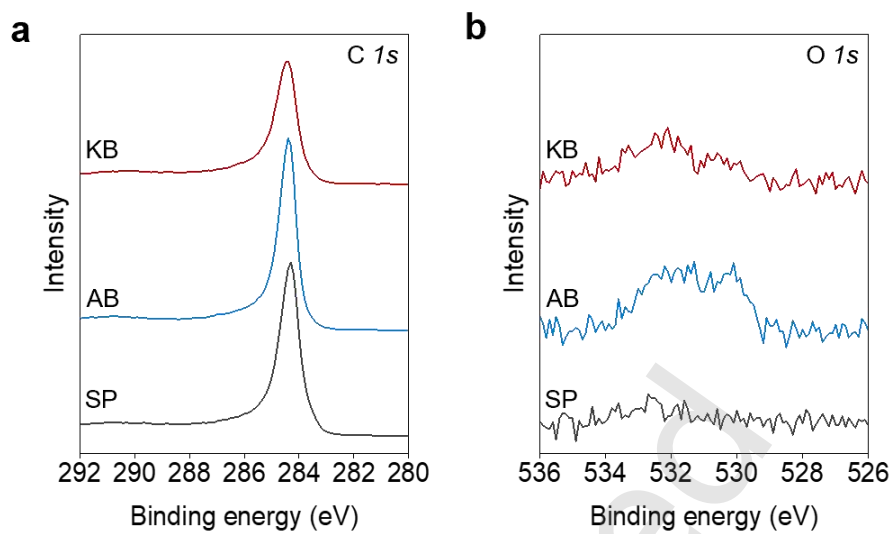
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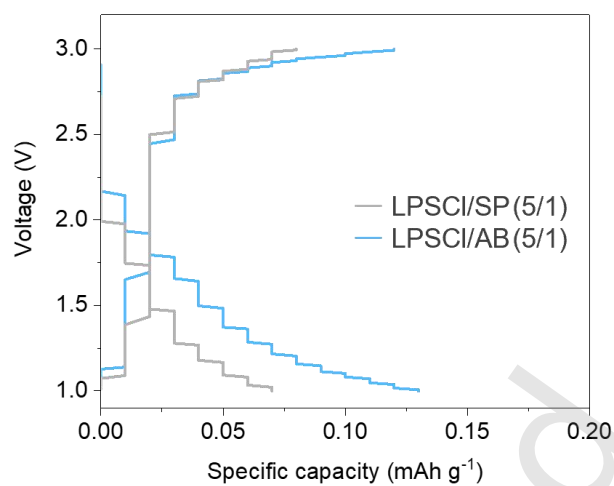
Supplementary Figure 1. Enlarged BET results of AB and SP. **a** and **c**: N₂ adsorption and desorption isotherm. **b** and **d**: Pore size distribution and cumulative surface area.



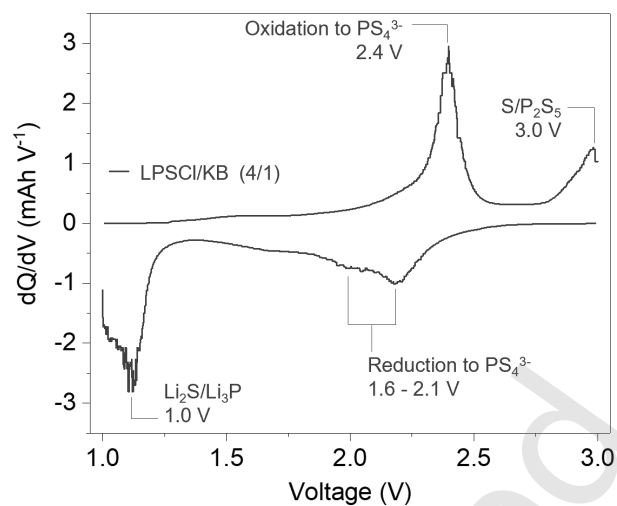
Supplementary Figure 2. Raman spectra of KB, AB and SP.



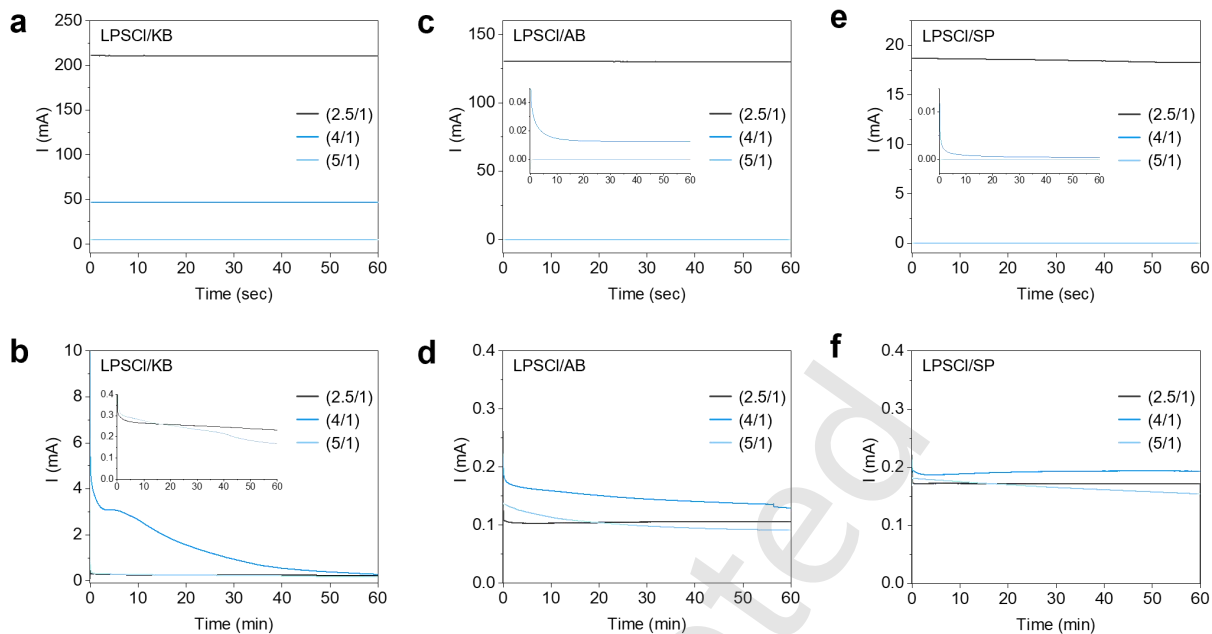
Supplementary Figure 3. X-ray photoelectron spectra of KB, AB and SP. **a** C 1s. **b** O 1s.



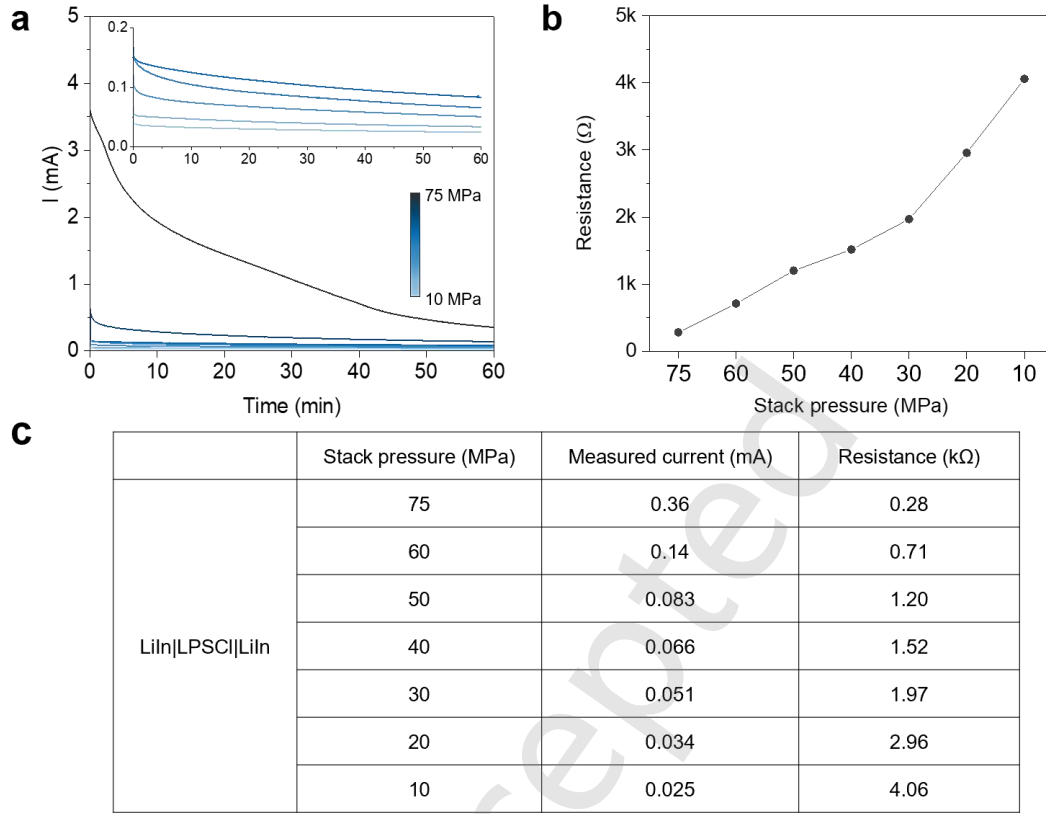
Supplementary Figure 4. Enlarged voltage profiles of LPSCI/AB and LPSCI/SP at (5/1). Stack pressure: 75 MPa; Testing temperature: 25 ± 1 °C.



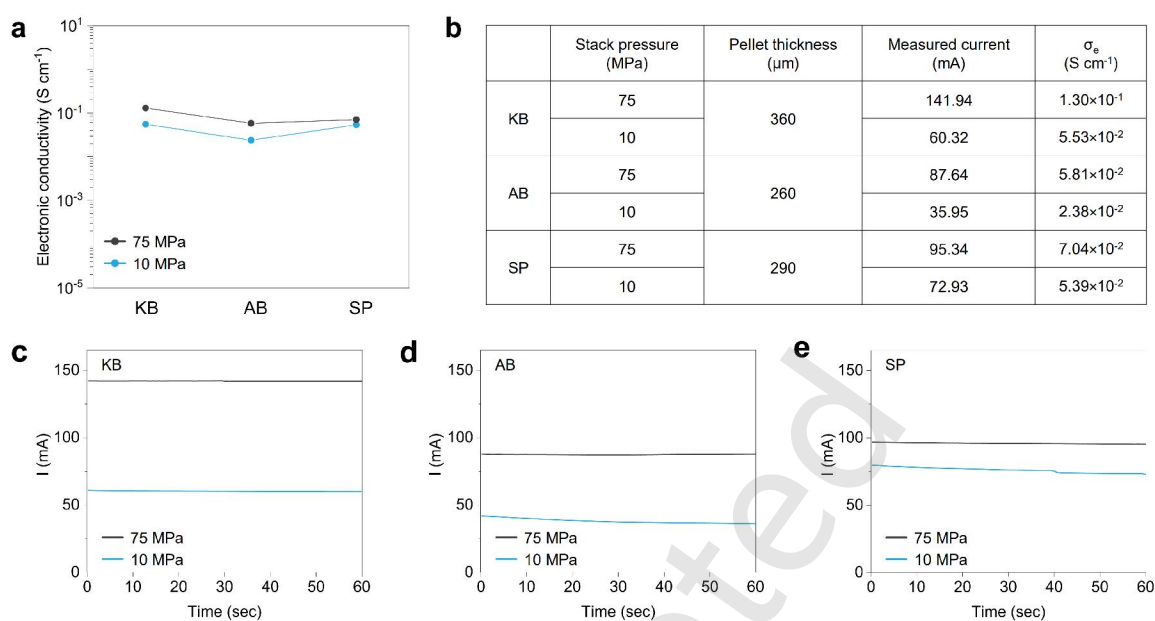
Supplementary Figure 5. dQ/dV profile of LPSCI/KB (4/1). Stack pressure: 75 MPa; Testing temperature: 25 ± 1 °C.



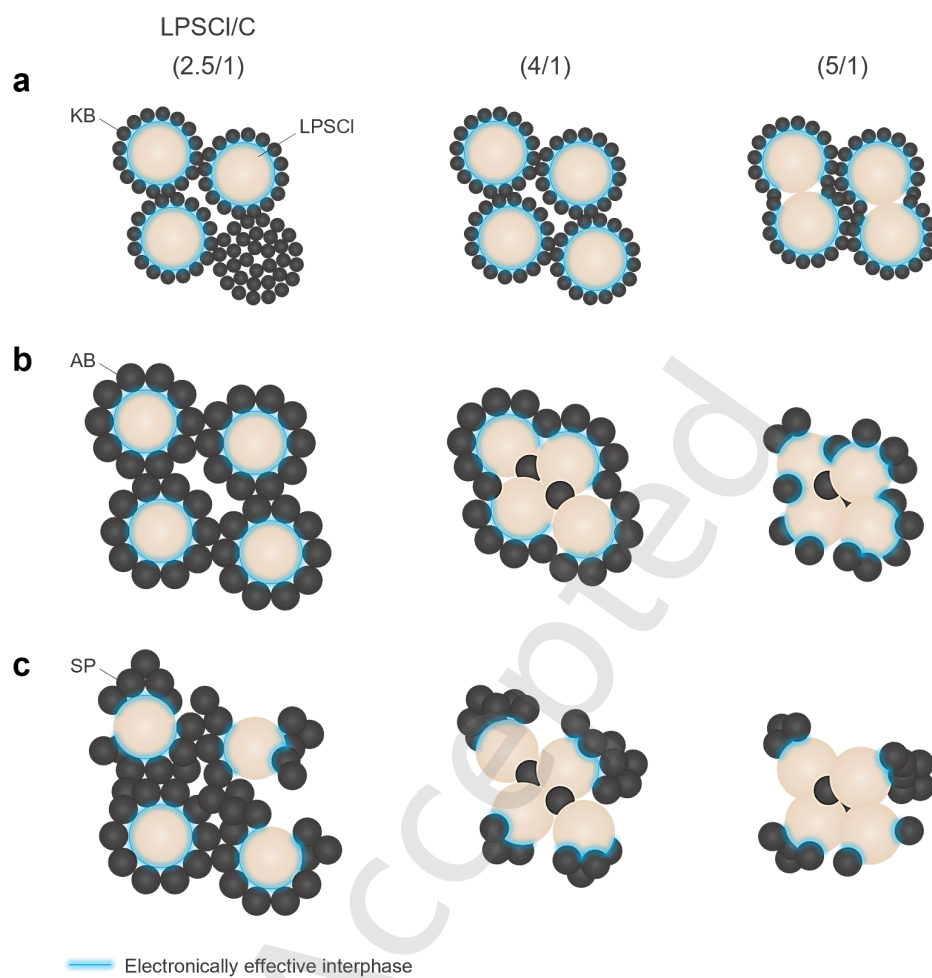
Supplementary Figure 6. DC polarization curves for electronic (top: a, c, e) and ionic conductivity (bottom: b, d, f) measurement of LPSCI composite. **a and b:** KB; **c and d:** AB; **e and f:** SP. DC polarization was conducted using an applied voltage of 50 mV for electronic conductivity measurements and 100 mV for ionic conductivity measurements at a temperature of 25 ± 1 °C.



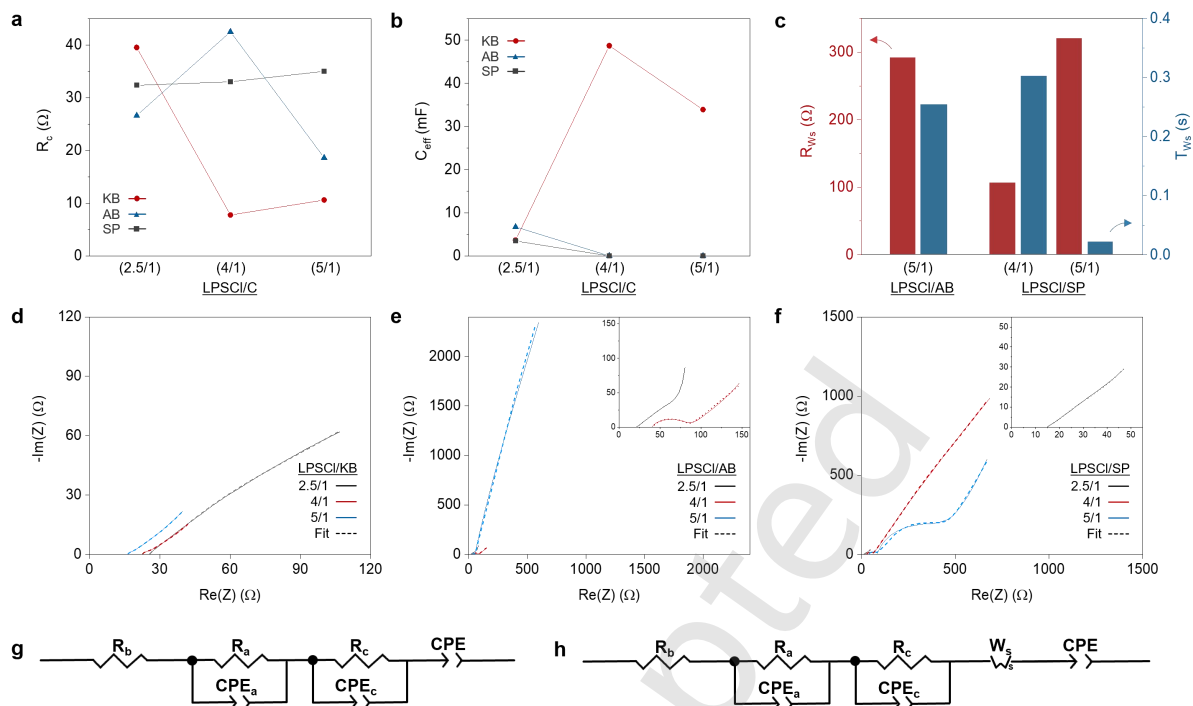
Supplementary Figure 7. **a** DC polarization curves of LiIn|LPSCl|LiIn cell measured under different stack pressures. **b** Resistance obtained from **a**. **c** Parameters used for ionic conductivity measurement. DC polarization measurements were conducted at an applied voltage of 100 mV and a temperature of 25 ± 1 °C. The obtained resistance was used to calculate the ionic conductivity of the composite electrodes.



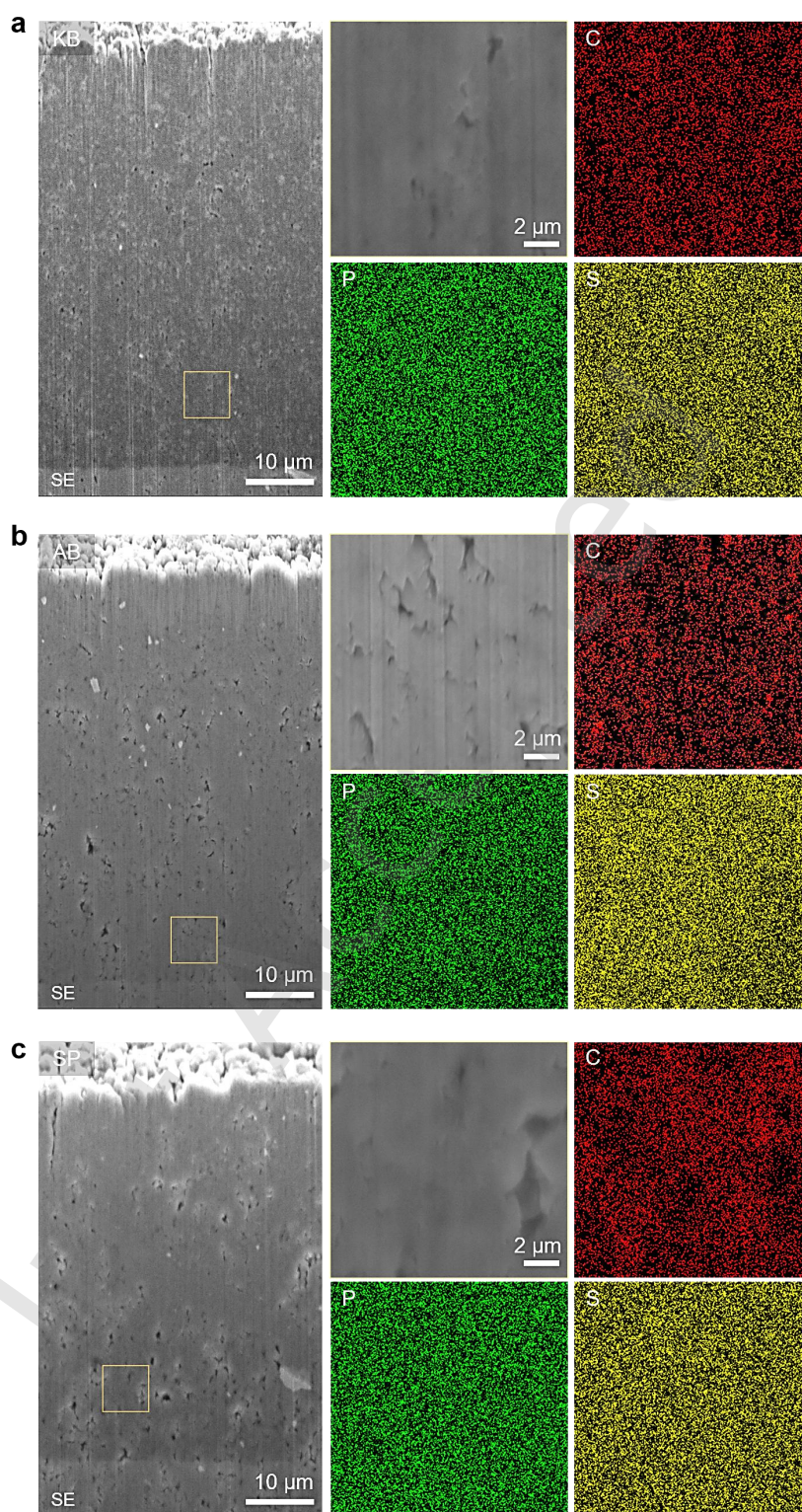
Supplementary Figure 8. **a** Electronic conductivity measurement of KB, AB and SP. **b** Parameters used for electronic conductivity measurement. DC polarization curves. **c** KB. **d** AB. **e** SP. DC polarization was conducted at an applied voltage of 50 mV and a temperature of $25 \pm 1^\circ\text{C}$.



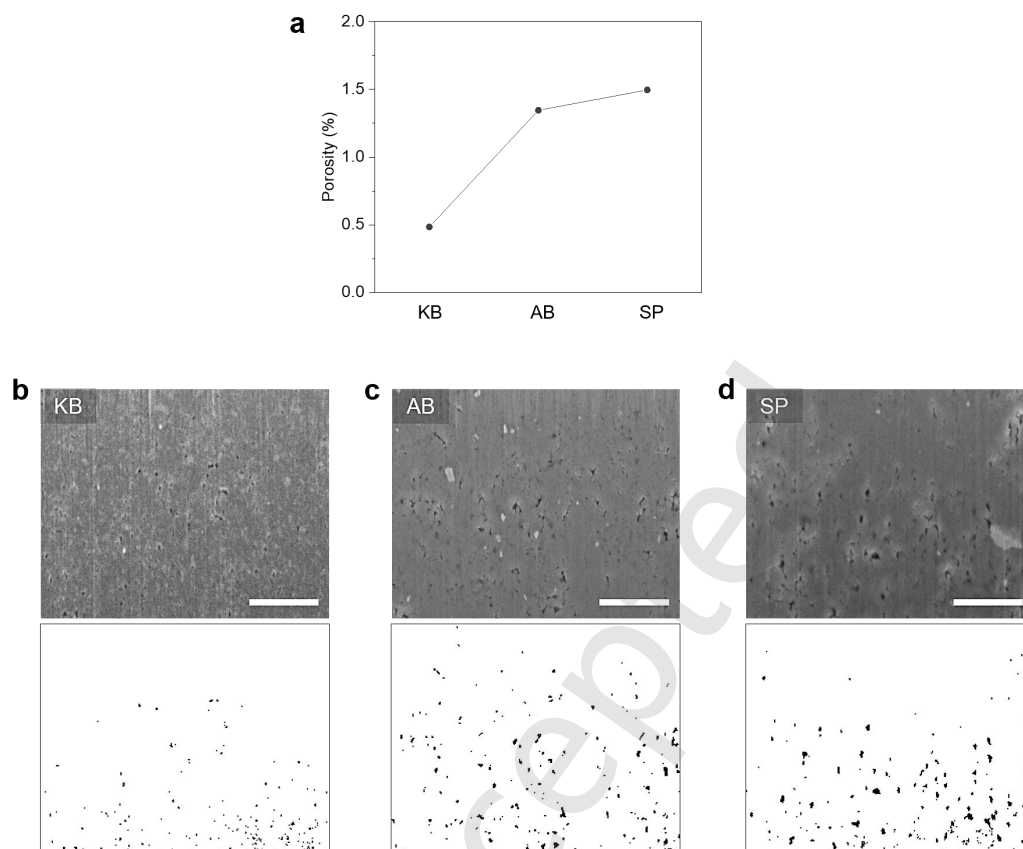
Supplementary Figure 9. Schematics of carbon distribution depending on LPSCl to carbon ratios of (2.5/1), (4/1) and (5/1). **a** KB. **b** AB. **c** SP.



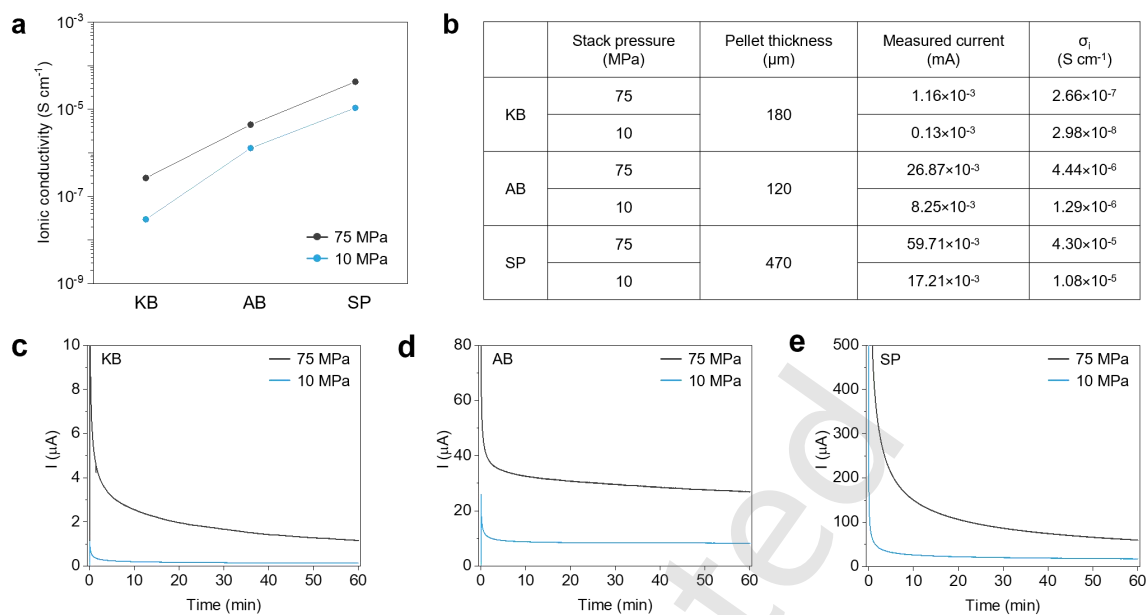
Supplementary Fig. 10. EIS analysis of LPSCI/C composites with different carbon architectures and compositions. **a** Cathode charge-transfer resistance (R_c) and **b** effective interphase capacitance (C_{eff}) of LPSCI/C composites with different carbon architectures (KB, AB, and SP) and LPSCI/C ratios. **(c)** Finite-length transport resistance (R_{Ws}) and relaxation time (T_{Ws}) of AB (5/1), SP (4/1), and SP (5/1). Representative EIS spectra and fitted curves for **d** KB, **e** AB, and **f** SP composites. Equivalent circuits used for fitting the impedance spectra of composites **g** without and **h** with finite-length transport behavior. R_b : bulk resistance, R_a : anode charge-transfer resistance, CPE_a : anode constant-phase element associated, W_s : short finite-length Warburg element.



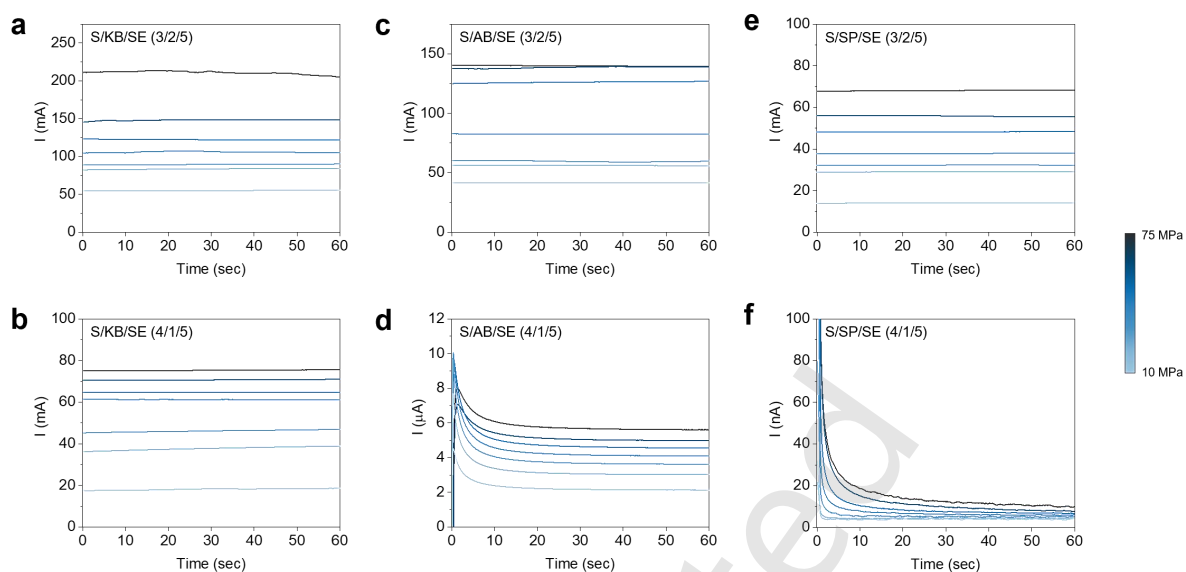
Supplementary Figure 11. Cross-sectional morphology and elemental mapping images of pristine LPSCl composite electrodes at 80 wt. % LPSCl. **a** KB. **b** AB. **c** SP. Middle and right columns show enlarged regions of the yellow-boxed regions from the low-magnification images in the left column and corresponding elemental mapping.



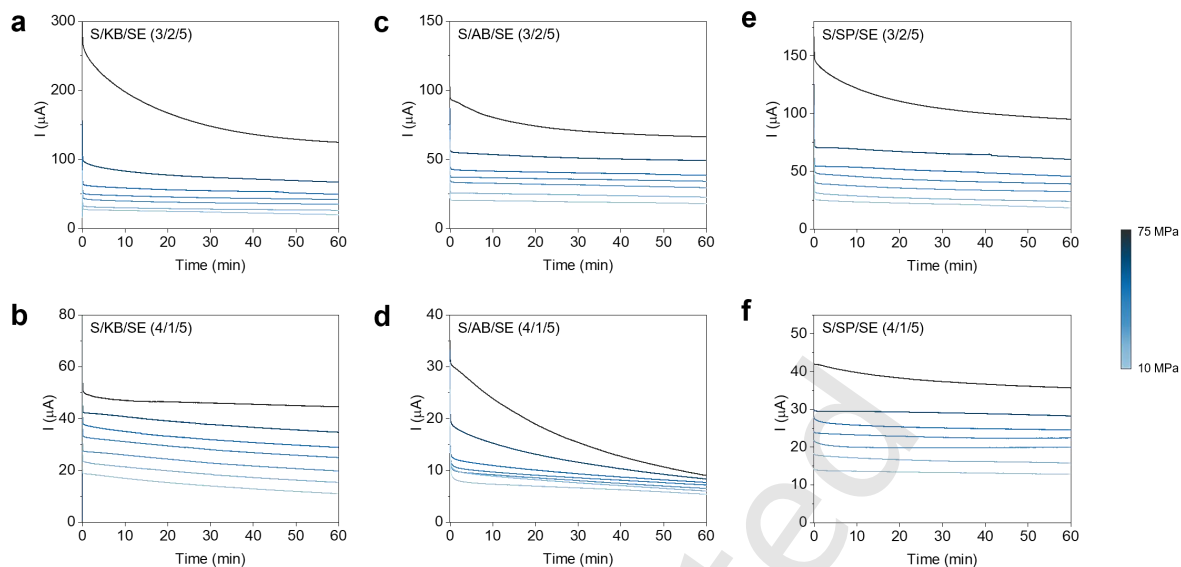
Supplementary Figure 12. **a** Porosity of pristine LPSCl composite electrodes at 80 wt. % LPSCl. Cropped regions used for porosity measurements and corresponding binary images from the low-magnification image in Supplementary Fig. 11. **b** KB. **c** AB. **d** SP. Scale bars: 10 μm.



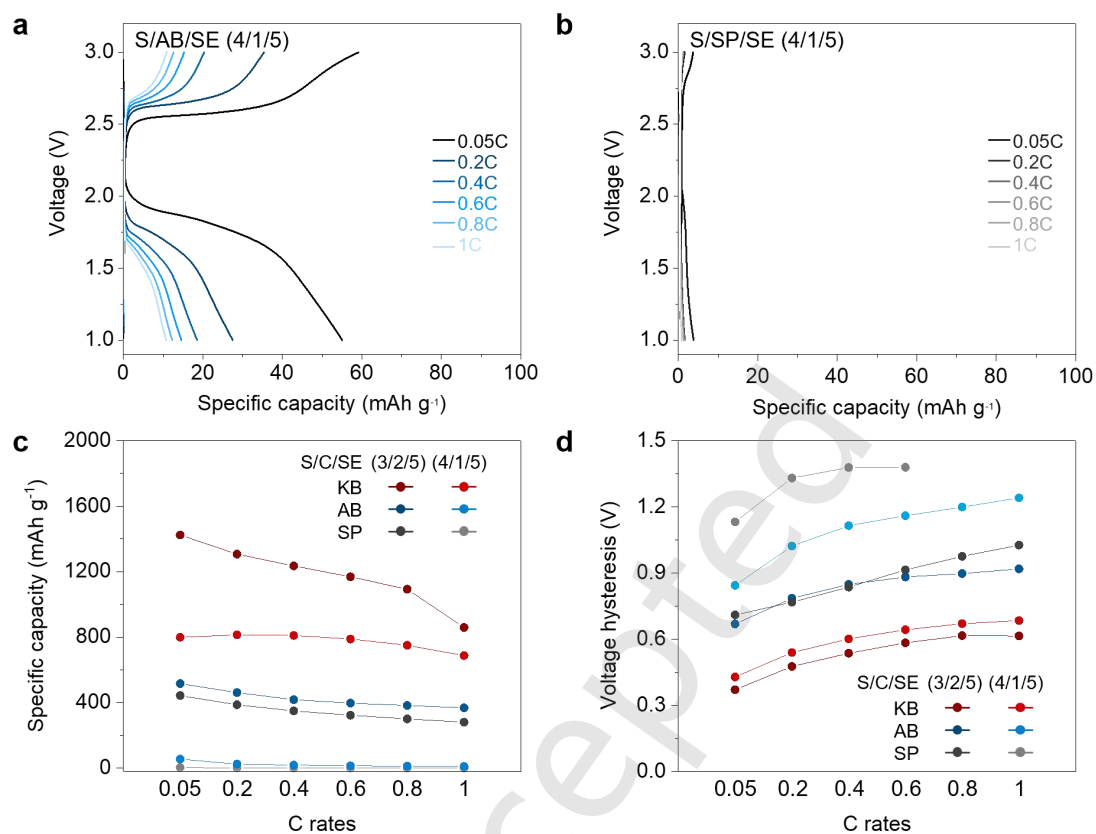
Supplementary Figure 13. **a** Ionic conductivity measurement of KB, AB and SP. **b** Parameters used for ionic conductivity measurement. **c-d** DC polarization curves. **c:** KB; **d:** AB; **e:** SP. DC polarization was conducted at an applied voltage of 100 mV and a temperature of 25 ± 1 °C.



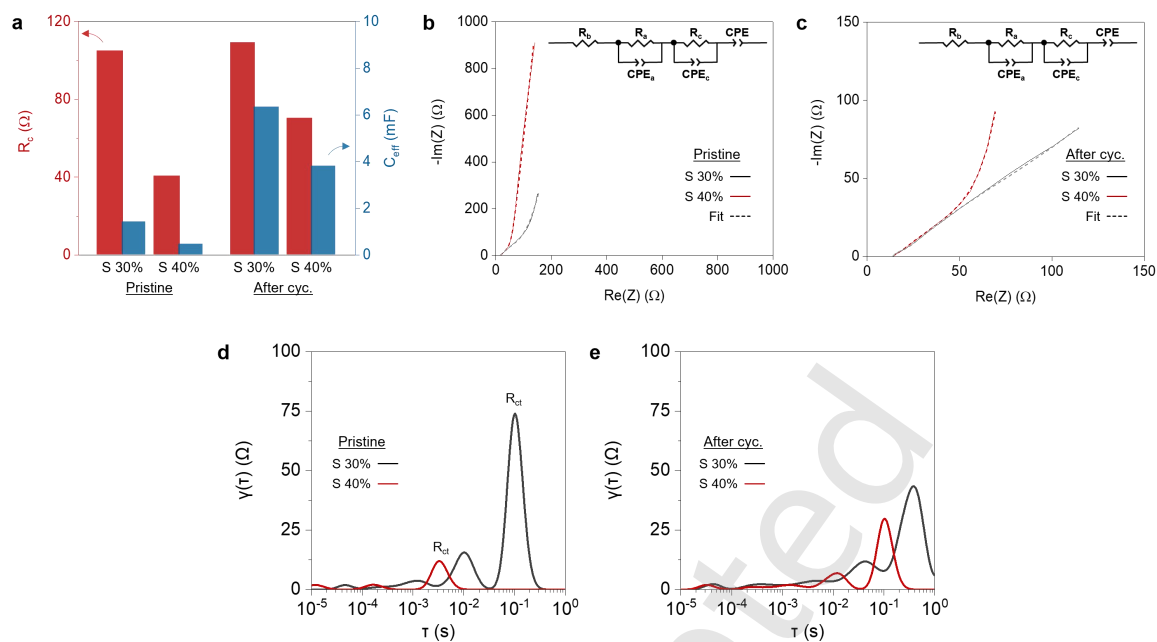
Supplementary Figure 14. DC polarization curves for electronic conductivity measurement of S composite with S ratios of 30 wt. % and 40 wt. % under stack pressures ranging from 75 to 10 MPa. **a and b:** KB; **c and d:** AB; **e and f:** SP. DC polarization was conducted at an applied voltage of 50 mV and a temperature of 25 ± 1 °C.



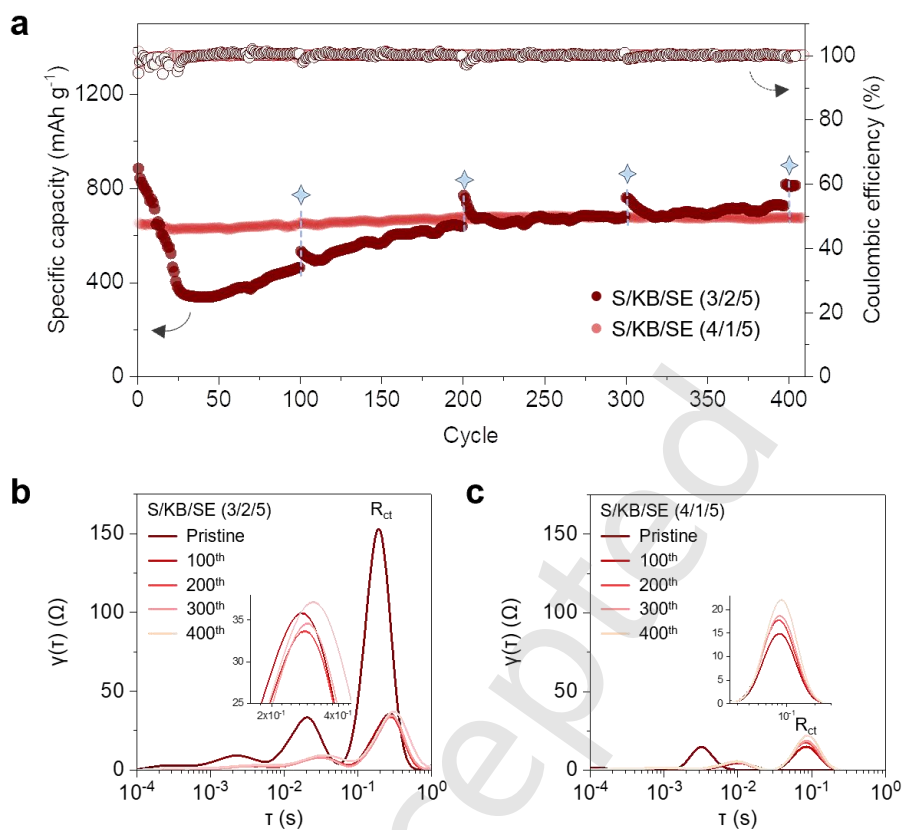
Supplementary Figure 15. DC polarization curves for ionic conductivity measurement of S composite with S ratios of 30 wt. % and 40 wt. % under stack pressures ranging from 75 to 10 MPa. **a and b:** KB; **c and d:** AB; **e and f:** SP. DC polarization was conducted at an applied voltage of 100 mV and a temperature of 25 ± 1 °C.



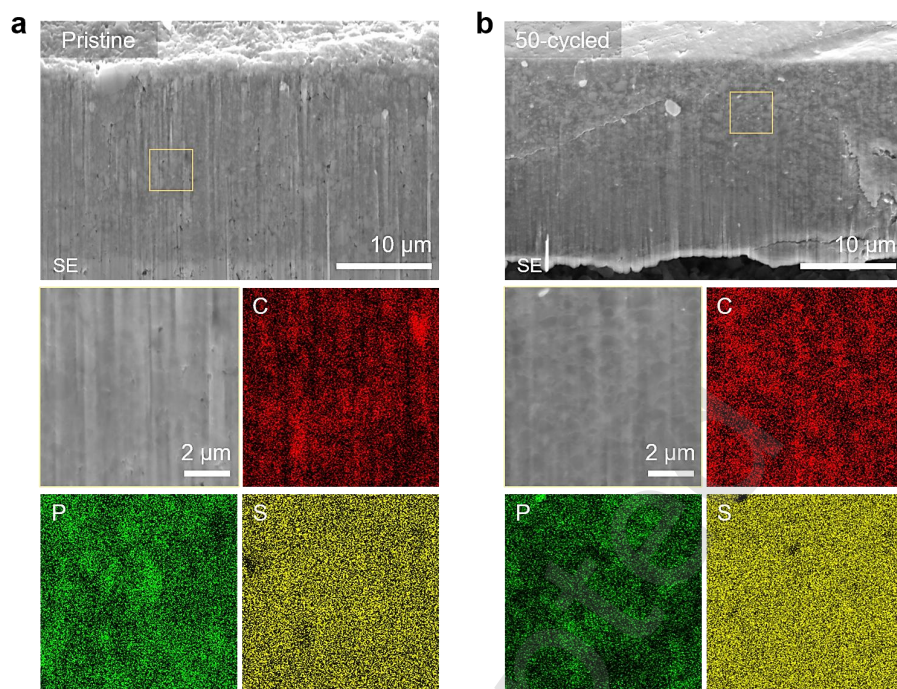
Supplementary Figure 16. a and b: Rate capability tests of sulfur composite with AB and SP at sulfur ratios of 40%. **c and d:** Rate capability comparison of sulfur composite with KB, AB and SP at sulfur ratios of 30% and 40% under stack pressure of 75 MPa. **c** Specific discharge capacities. **d** Voltage hysteresis.



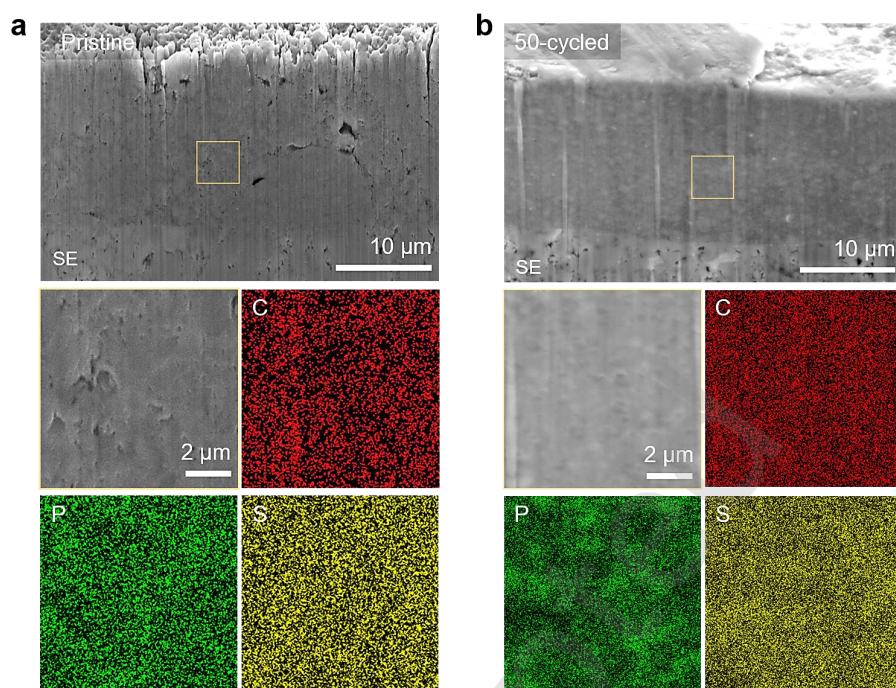
Supplementary Fig. 17. **a** R_c and C_{eff} of KB electrodes with 30 wt. % and 40 wt. % S in pristine and cycled states. EIS spectra and fitted curves for **b** pristine and **c** cycled electrodes. Insets in **b** and **c** show the equivalent circuit used for fitting the impedance spectra. DRT spectra for **d** pristine and **e** cycled electrodes.



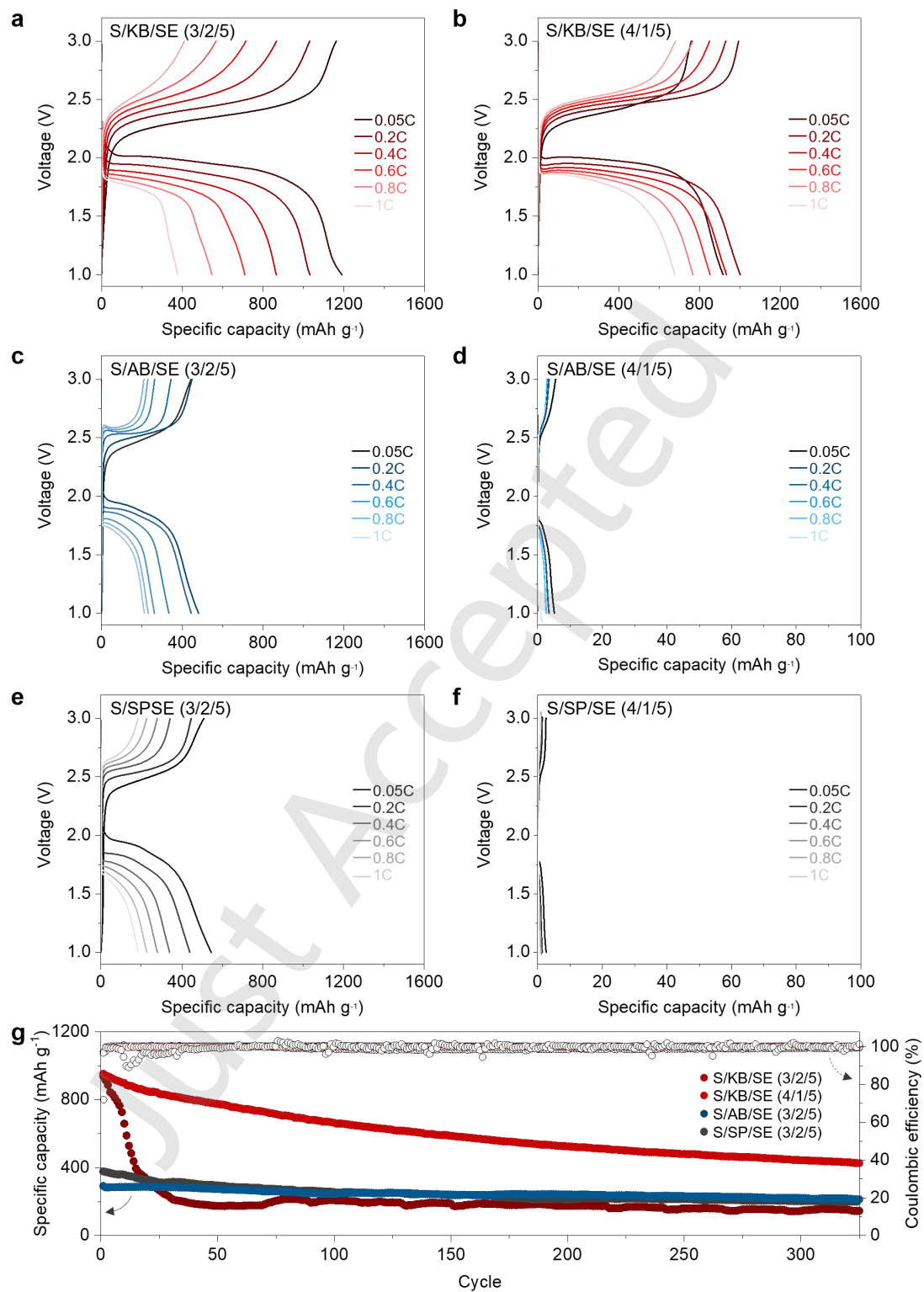
Supplementary Figure 18. **a** Long-term cycling performance of S composite electrodes with KB at S ratios of 30 wt. % and 40 wt. % under 1C (1C = 1.5 mA cm⁻²). DRT spectra in discharged state of S composite electrodes with **b** 30 wt. % **c** 40 wt. %. DRT analysis was conducted in pristine state and every 100 cycles as indicated by blue symbols. Stack pressure: 75 MPa; Testing temperature: 25 ± 1 °C.



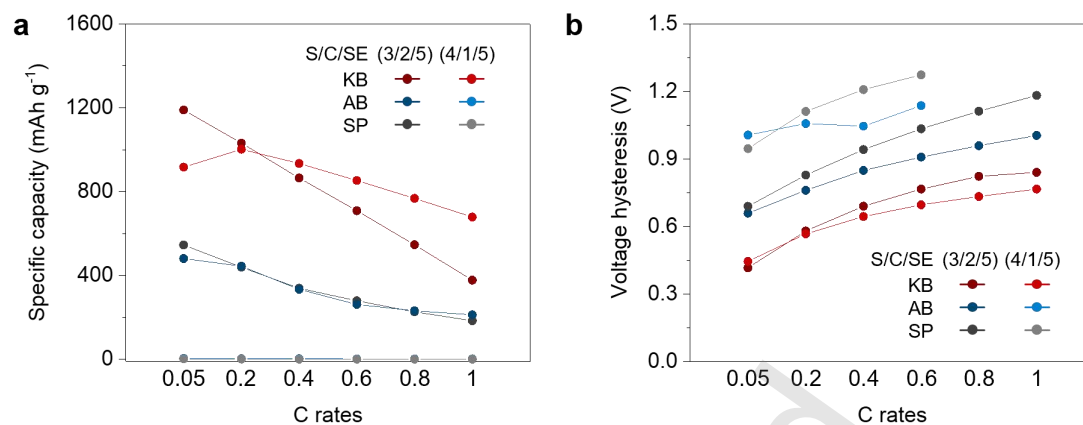
Supplementary Figure 19. Cross-sectional morphology and elemental mapping images of S composite electrodes with KB at 30 wt. % S. **a** Pristine. **b** After 50 cycles. Middle and bottom rows show enlarged images of the yellow-boxed regions from the low-magnification images in the top row and corresponding elemental mapping.



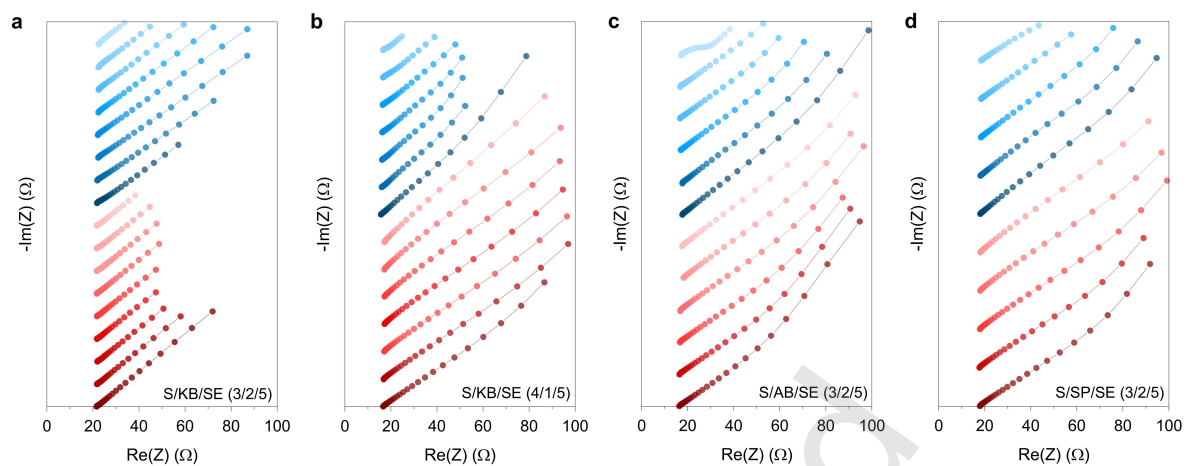
Supplementary Figure 20. Cross-sectional morphology and elemental mapping images of S composite electrodes with KB at 40 wt. % S. **a** Pristine. **b** After 50 cycles. Middle and bottom rows show enlarged images of the yellow-boxed regions from the low-magnification images in the top row and corresponding elemental mapping.



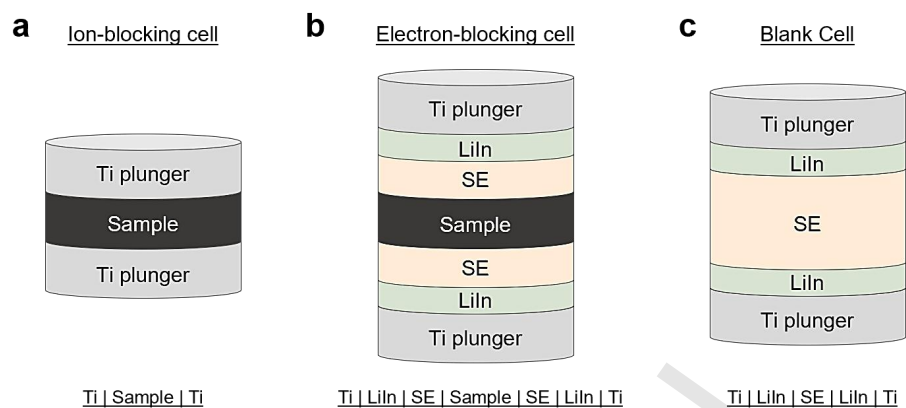
Supplementary Figure 21. LiIn||S cell performance at stack pressure of 10 MPa. **a-f:** Rate capability tests of sulfur composite with KB, AB and SP at sulfur ratios of **a, c and e:** 30% and **b, d and f:** 40%. **g** Long-term capacity retention at 0.2C (1C = 2 mA cm⁻²). Testing temperature: 25 ± 1 °C.



Supplementary Figure 22. Rate capability comparison of sulfur composite with KB, AB and SP at sulfur ratios of 30% and 40% under stack pressure of 10 MPa. **a** Specific discharge capacities. **b** Voltage hysteresis.



Supplementary Figure 23. EIS spectra of KB, AB and SP-based electrodes. **a** KB (30 wt. % S) **b** KB (40 wt. % S). **c** AB (30 wt. % S). **d** SP (30 wt. % S).



Supplementary Figure 24. Schematics of cell configuration for conductivity measurement. **a** Ion-blocking cell. **b** Electron-blocking cell. **c** Blank cell.

Table S1. Parameters used for electronic conductivity measurement of LPSCI composite at each composition.

		Stack pressure (MPa)	Pellet thickness (μm)	Measured current (mA)	σ_e (S cm^{-1})
LPSCI/KB	(2.5/1)	75	230	210.87	1.24×10^{-1}
	(4/1)		310	46.71	3.69×10^{-2}
	(5/1)		210	4.99	2.67×10^{-3}
LPSCI/AB	2.5/1		200	130.15	6.63×10^{-2}
	(4/1)		210	1.24×10^{-2}	6.65×10^{-6}
	(5/1)		210	2.23×10^{-5}	1.19×10^{-8}
LPSCI/SP	2.5/1		0.03	18.27	1.40×10^{-2}
	(4/1)		0.031	3.68×10^{-4}	2.90×10^{-7}
	(5/1)		0.028	1.12×10^{-5}	7.98×10^{-9}

Table S2. Parameters used for ionic conductivity measurement of LPSCI composite at each composition.

		Stack pressure (MPa)	Pellet thickness (μm)	Measured current (mA)	σ_i (S cm^{-1})
LPSCI/KB	(2.5/1)	75	180	2.57×10^{-1}	1.55×10^{-4}
	(4/1)		100	2.87×10^{-1}	1.91×10^{-4}
	(5/1)		180	1.68×10^{-1}	7.31×10^{-5}
LPSCI/AB	2.5/1		180	1.06×10^{-1}	3.48×10^{-5}
	(4/1)		220	1.29×10^{-1}	5.69×10^{-5}
	(5/1)		210	9.11×10^{-2}	3.28×10^{-5}
LPSCI/SP	2.5/1		140	1.71×10^{-1}	5.91×10^{-5}
	(4/1)		180	1.94×10^{-1}	9.82×10^{-5}
	(5/1)		150	1.54×10^{-1}	5.20×10^{-5}

Table S3. Parameters used for electronic conductivity measurement of S composite at each composition.

		Stack pressure (MPa)	Pellet thickness (μm)	Measured current (mA)	σ_e (S cm^{-1})
S/KB/SE	(3/2/5)	75	170	2.05×10^2	8.88×10^{-2}
		60		1.49×10^2	6.44×10^{-2}
		50		1.22×10^2	5.27×10^{-2}
		40		1.05×10^2	4.55×10^{-2}
		30		9.02×10^1	3.91×10^{-2}
		20		8.45×10^1	3.66×10^{-2}
		10		5.53×10^1	2.40×10^{-2}
	(4/1/5)	75	170	7.56×10^1	3.27×10^{-2}
		60		7.09×10^1	3.07×10^{-2}
		50		6.47×10^1	2.80×10^{-2}
		40		6.10×10^1	2.64×10^{-2}
		30		4.69×10^1	2.03×10^{-2}
		20		3.89×10^1	1.68×10^{-2}
		10		1.86×10^1	8.05×10^{-3}
S/AB/SE	(3/2/5)	75	150	1.39×10^2	5.33×10^{-2}
		60		1.38×10^2	5.29×10^{-2}
		50		1.27×10^2	4.85×10^{-2}
		40		8.24×10^1	3.15×10^{-2}
		30		5.97×10^1	2.28×10^{-2}
		20		5.59×10^1	2.15×10^{-2}
		10		4.14×10^1	1.58×10^{-2}
	(4/1/5)	75	130	5.61×10^{-3}	1.86×10^{-6}
		60		4.97×10^{-3}	1.65×10^{-6}
		50		4.55×10^{-3}	1.51×10^{-6}
		40		4.09×10^{-3}	1.35×10^{-6}
		30		3.62×10^{-3}	1.20×10^{-6}
		20		3.02×10^{-3}	1.00×10^{-6}
		10		2.13×10^{-3}	7.07×10^{-7}
S/SP/SE	(3/2/5)	75	300	6.83×10^1	5.22×10^{-2}
		60		5.55×10^1	4.25×10^{-2}
		50		4.84×10^1	3.70×10^{-2}
		40		3.79×10^1	2.89×10^{-2}
		30		2.92×10^1	2.23×10^{-2}
		20		2.77×10^1	2.12×10^{-2}
		10		1.41×10^1	1.08×10^{-2}
	(4/1/5)	75	230	1.03×10^{-5}	4.19×10^{-9}
		60		7.51×10^{-6}	3.06×10^{-9}
		50		7.14×10^{-6}	2.91×10^{-9}
		40		6.37×10^{-6}	2.60×10^{-9}
		30		5.10×10^{-6}	2.08×10^{-9}
		20		4.42×10^{-6}	1.80×10^{-9}
		10		3.60×10^{-6}	1.47×10^{-9}

Table S4. Parameters used for ionic conductivity measurement of S composite at each composition.

		Stack pressure (MPa)	Pellet thickness (μm)	Measured current (mA)	σ_i (S cm^{-1})
S/KB/SE	(3/2/5)	75	220	1.09×10^{-4}	4.40×10^{-5}
		60		6.71×10^{-5}	3.60×10^{-5}
		50		4.96×10^{-5}	3.44×10^{-5}
		40		4.21×10^{-5}	3.27×10^{-5}
		30		3.51×10^{-5}	3.19×10^{-5}
		20		2.60×10^{-5}	3.15×10^{-5}
		10		2.01×10^{-5}	3.06×10^{-5}
	(4/1/5)	75	190	4.45×10^{-5}	1.23×10^{-5}
		60		3.47×10^{-5}	1.12×10^{-5}
		50		2.88×10^{-5}	1.07×10^{-5}
		40		2.50×10^{-5}	9.78×10^{-6}
		30		1.98×10^{-5}	7.84×10^{-6}
		20		1.54×10^{-5}	6.84×10^{-6}
		10		1.10×10^{-5}	4.81×10^{-6}
S/AB/SE	(3/2/5)	75	170	6.64×10^{-5}	1.77×10^{-5}
		60		4.92×10^{-5}	1.64×10^{-5}
		50		3.85×10^{-5}	1.55×10^{-5}
		40		3.42×10^{-5}	1.54×10^{-5}
		30		2.95×10^{-5}	1.53×10^{-5}
		20		2.27×10^{-5}	1.50×10^{-5}
		10		1.81×10^{-5}	1.47×10^{-5}
	(4/1/5)	75	170	9.00×10^{-6}	2.47×10^{-6}
		60		8.33×10^{-6}	2.37×10^{-6}
		50		7.70×10^{-6}	2.27×10^{-6}
		40		7.16×10^{-6}	2.15×10^{-6}
		30		6.55×10^{-6}	2.01×10^{-6}
		20		6.02×10^{-6}	1.96×10^{-6}
		10		5.40×10^{-6}	1.85×10^{-6}
S/SP/SE	(3/2/5)	75	200	9.51×10^{-5}	3.31×10^{-5}
		60		6.03×10^{-5}	2.69×10^{-5}
		50		4.57×10^{-5}	2.58×10^{-5}
		40		3.92×10^{-5}	2.46×10^{-5}
		30		3.25×10^{-5}	2.29×10^{-5}
		20		2.40×10^{-5}	2.11×10^{-5}
		10		1.84×10^{-5}	1.87×10^{-5}
	(4/1/5)	75	100	3.57×10^{-5}	6.07×10^{-6}
		60		2.83×10^{-5}	5.42×10^{-6}
		50		2.45×10^{-5}	5.30×10^{-6}
		40		2.25×10^{-5}	5.22×10^{-6}
		30		2.00×10^{-5}	5.05×10^{-6}
		20		1.58×10^{-5}	4.53×10^{-6}
		10		1.28×10^{-5}	4.09×10^{-6}