Solid–electrolyte-interphase design in constrained ensemble for solid-state batteries†

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Solid-state-batteries (SSBs) represent one of the most promising directions in the energy-storage field. The development of SSBs, however, is currently limited by the complex [electro-]chemical reactions that inevitably occur at the interface of solid-state electrolyte (SSE) particles. Moreover, given the material complexity of such systems, there is no straightforward methodology for addressing these interface instabilities. In this work, a combined high-throughput ab initio computation and machine learning approach is used to study and design solid-state solid–electrolyte-interphase (SEI) with tunable electrochemical stabilities using our unique constrained ensemble description. 

Solid-state batteries are widely seen as the next great step in energy storage technology to displace lithium-ion batteries as the gold standard of secondary batteries. However, the development of solid state batteries has remained slow, largely due to complicated [electro-]chemical reactions that occur at interfaces between solid phases. Thus, there is an urgent need to develop a more effective approach to handle this major design hurdle. In this work, we introduce a comprehensive method to systematically study and, ultimately, design solid–electrolyte-interphases for such an all-solid electrochemical system. Within the general framework of mechanical constriction design, a new concept of critical effective modulus of interphase is proposed and quantified. Any local mechanical constriction level below this critical value will allow decompositions to happen. The decomposition, however, will fill local voids and cracks, and eventually increase the local mechanical constriction beyond the critical value to suppress further decompositions through metastability. Therefore, from this unique perspective of interphase electrochemical evolution, designing the interphase with a reduced critical effective modulus becomes an effective approach to increase dynamic interface stabilities. The method can be applied to the interface design of broad types of material systems, including glass, ceramic, polymer and their composites.

The rapid charging, high energy density, and non-flammability of SSBs1–8 are critical for the ongoing push to achieve 100% vehicle electrification – a necessity for meaningful progress on climate change. Despite this promise, there remain many technical barriers in the way of broadly adoptable SSBs. One of the greatest of which is the electrochemical instability often found at the various interfaces in SSB cells.9–13

In our previous work, we have detailed how the fundamental thermodynamics of materials deviate in SSBs as compared to conventional liquid-cells.11–13 In short, liquid-cells operate in accordance with the well-studied state variables of pressure (p), temperature (T), and voltage (ϕ). In other words, the operator of a conventional cell sets these state variables experimentally and the material phases equilibrate (by, for example, [de]lithiation). Thus, the conventional liquid-cell can be well described by the \( pT\phi \) thermodynamic ensemble, which we term the “unconstrained” ensemble for contrast with the “constrained” ensemble for all-solid-state systems, described below.

Our previous work has shown that, unlike liquid systems, pressure is not a valid state variable in SSBs. In other words, the experimenter does not have direct control of the pressure within a working cell. When chemical decompositions occur within SSBs, strong local pressure gradients can form as the pressure is a function of the phases which are present locally. In liquid-cells, the liquid–electrolyte would flow to equilibrate this local pressure in response to any external pressure applied.
at the cell level. This cannot occur in an SSB, where the SSE maintains a pressure gradient. Thus, an operator has no direct control of the local pressure within an SSB. As a result, the electrochemical stability of solid-electrolytes varies considerably depending on cell and material design.

To describe these profound thermodynamic differences in SSBs, we introduce the concept of “constrained” ensemble, where the pressure is no-longer a state variable but is a function of the mole-fraction of local phases ($\{x_i\}$). There are two critical insights for understanding this change of state variables. The first is the observation that many solid-electrolytes tend to decompose to products that are larger in total volume than the original electrolyte. The second is that, unlike liquid batteries, this volume expansion induced by decomposition has a profound impact on the underlying thermodynamics of the system.

With regards to the volume expansion, consider the electrolyte Li$_{10}$GeP$_2$S$_{12}$ (LGPS). When LGPS chemically decays, it forms the products Li$_4$GeS$_4$ (LGS) and Li$_3$PS$_4$ (LPS). These products maintain a volume that is 2% larger than the original LGPS. Moreover, the volume expansion from electrochemical decompositions of LGPS when charged vs. lithium metal is even larger. In fact, the volume increases can reach values in excess of 30% at the complete oxidation (at high voltage) and reduction (at low voltage) limits.$^{11,13}$ This increase in volume is referred to as the reaction strain (or reaction dilation) and is denoted $\varepsilon_{\text{RXN}}$. It is important to note that the reaction strain is a stress-free strain, meaning that unlike conventional strain, it does not result from an applied stress – it is solely the result of the [electro]-chemical decomposition that forms products having a different volume than the initial reactant (i.e. the electrolyte).

However, under an applied mechanical constriction, like the typical conditions of SSBs, the reaction strain can result in local stresses. That is to say, although the reaction strain is not caused by mechanical stress, it can cause new local mechanical stresses to form within the battery. These local stresses are the result of trying to fit a larger phase (the decomposed products) into the space that was previously occupied by a smaller phase (the electrolyte). This phenomenon is generalized by the formula $p = K_{\text{eff}}\varepsilon_{\text{RXN}}$, where the pressure ($p$) is the average compressive stress on a solid volume unit and $K_{\text{eff}}$ is the local effective mechanical modulus. This formula quantifies that for a solid system, characterized by the modulus $K_{\text{eff}}$, to accommodate some expanded decomposition products, a local pressure of $p$ must be generated. The reaction strain is directly a function of the mole fraction of the present phases. For example, when the respective mole fractions of LGPS, LGS, and LPS are 1, 0, 0, then the reaction strain is 0%. Conversely, when the respective mole fractions are 0, 1/3, 2/3, the reaction strain is 2%. As a result of pressure being a function of $\varepsilon_{\text{RXN}}$ and $\varepsilon_{\text{RXN}}$ being a function of the present mole fractions, it is seen that pressure is now no-longer a state variable but instead a function of the present mole fractions.

This dependency of the mechanical stress distribution on the local mole fractions is what causes such strong changes in the thermodynamics of solid-state vs. liquid batteries. To illustrate this, consider a thought experiment in which a small volume fraction of the total electrolyte, located at $\bar{r}_0$, has decomposed. The decomposed [volume] fraction ($x_0$) is now a piecewise function of the form:

$$x_0(\bar{r}) = \begin{cases} 1 & \text{when } \bar{r} = \bar{r}_0 \\ 0 & \text{when } \bar{r} \neq \bar{r}_0 \end{cases}$$

In this case, the reaction strain will be non-zero at $\bar{r}_0$ and zero everywhere else. As a result, the pressure caused by this reaction strain will also be non-zero only in the neighborhood of $\bar{r}_0$. In a liquid battery, such a pressure gradient would not be in equilibrium as fluids must have equal pressures at all point. As a result, the decomposed products would flow within the liquid electrolytes until the pressure comes to equilibrium with the environment (i.e. 1 atm for non-pressurized battery cells). In a solid-state system, however, this flow of the decomposed products will not occur because solids can maintain pressure gradients. The result is that a fluid battery will maintain a constant pressure equal to the environment regardless of the decomposed fractions whereas a solid-state battery can potentially have significant localized pressure that is not controlled by the external environment. That is, pressure is a valid state variable of liquid batteries but not one of solid-state batteries.

Consideration of the functional form of $p(\{x_i\})$ has shown to have profound impact on the intrinsic electrochemical stability of ceramic-sulfide SSEs. When the functional form is taken to be constant (i.e. isobaric conditions), the traditional $pT\phi$ (grand canonical) ensemble is recovered, and the electrochemical stability is seen to be on the order of 1.7–2.1 V vs. lithium for LGPS.$^{14,15}$ In contrast, when the pressure is taken to be proportional to the local strain, the (meta)stability window can greatly expand to approximately 1–4 V,$^{1,12,13}$ and kinetic stability can reach down to 0 V and up to 10 V.$^{1,14}$

We then for the first time apply the constrained ensemble to explicitly treat the calculation of interface stability, and articulate quantitatively one most critical aspect regarding interface stabilities in SSBs in electrochemical cycling, through the implementation of high-throughput $ab$ $initio$ computations and machine learning. We find that when the effective local modulus at the interface ($K_{\text{eff}}$) is above some critical threshold decided by interface materials (denoted $K_{\text{crit}}$), the interface will remain stable throughout cycling. When the effective modulus is lower than this threshold, the interface is unstable and will grow with cycling and thus limit the battery performance. Finally, we utilize this computational approach to develop a method for engineering solid–electrolyte-interphases (SEI’s) which are stabilized through electrochemical evolution within a prescribed voltage window.

**Modelling the interfaces within the constrained ensemble**

The free energy picture of interface decomposition is depicted in Fig. 1(a). The interface is initially composed of two phases

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(\(x\) and \(\beta\)) and is seen to decompose with a reaction energy magnitude of \(G_{\text{hull}}\) via some reaction:

\[
(1 - x_{\beta})x + x_{\beta}\beta \rightarrow \sum_i d_i D_i
\]  

(1)

Where \(d_i\) and \(D_i\) are the decomposed fractions and phases, respectively. The value of \(x\) that maximizes \(G_{\text{hull}}\) is denoted \(x_m\) and defines the “pseudo-phase” as \((1 - x_m)x + x_m\beta\). This pseudo-phase is the most kinetically driven decomposition by definition. The value of the pseudo-phase for modelling the interface is that we can computationally represent the interface as a single phase with composition \((1 - x_m)x + x_m\beta\) and energy \((1 - x_m)G_x + x_mG_\beta\) where \(G_i\) is the free energy of phase \(i\). Thus, \(ab\ initio\) computations can be used to predict interface phenomena. Detailed discussion of pseudo-phase methods can be found in ref. 16.

In liquid-cells (\(pT\phi\) ensemble), both the initial interface and decomposed interface exist at the same pressure, which is controlled by the operator. However, analogously to the effects of constraint on bulk SSE’s, the pseudo-phase decomposition in solid-state reactions takes the form:

\[
(1 - x_{\beta})x + x_{\beta}\beta \rightarrow \sum_i d_i D_i \text{ with } K_{\text{crit}}\text{RXN}
\]  

(2)

Eqn (1) and (2) differ principally in the products \(D_i\) of eqn (2) existing in a state of elevated pressure given by the local strain due to the decomposition reaction \((\text{eqn RXN})\) and some effective modulus \(K_{\text{crit}}\). This additional pressure term perturbs the reaction energy accompanying eqn (1) by a factor of \(K_{\text{crit}}\text{RXN}\), where \(V^0\) is the reference volume. Fig. 1(b) and (c) introduce using the convex hull illustration the parameter \(K_{\text{crit}}\) as a metric for how much mechanical constriction (measured by \(K_{\text{crit}}\)) is
required for the perturbed reaction to cease being thermodynamically favorable. Eqn (3) describes $K_{\text{crit}}$ mathematically as the value of $K_{\text{eff}}$ such that the energy cost of local pressurization exactly cancels the driving reaction energy ($\Delta G_{\text{RXN}}^0$). Thus, the interface reaction is expected to be thermodynamically favorable if and only if $K_{\text{eff}} < K_{\text{crit}}$.

$$\Delta G_{\text{RXN}}^0 + K_{\text{crit}} \rho_{\text{RXN}} V^0 = 0$$ (3)

In light of the effects of Fig. 1(a)–(c), Fig. 1(d) catalogs expected reactions involving SEI formation in solid-state batteries. The initial interfacing materials are taken to be the SSE and an arbitrary material “A”. During battery assembly, there exists virtually no applied mechanical constriction ($K_{\text{eff}} \approx 0$ GPa) so the interfacial chemical reaction proceeds as in a standard canonical form (i.e. the system is not yet open to lithium, hence it is not grand canonical). Following pseudo-phase conventions, the fraction of A that produces the maximum thermodynamic driving force is denoted $x_m$.

Deviation from grand canonical behavior is possible after the battery cell is fully assembled, at which point mechanical constriction may (or may not) be applied. In the absence of mechanical constriction, or when the local mechanical constriction is weak ($K_{\text{eff}} < K_{\text{krit}}$), the initially formed SEI will be able to further decay once the battery is assembled and the system becomes open to lithium. This is represented in Fig. 1(d) as SEI growth. As the battery is cycled, the voltage will likely pass in and out of the SEI’s electrochemical stability window, causing the initial SEI to continuously decay into secondary SEIs allowing continued growth. In contrast, if the local mechanical constriction is sufficiently rigid ($K_{\text{eff}} > K_{\text{krit}}$), either initially or after certain cycling steps, the following cycling of the voltage in and out of the initial SEI’s stability window will not lead to further decompositions. This process is described as “SEI maintenance” through the local stability gained from the electrochemical evolution that could contain processes of both the initial decomposition ($K_{\text{eff}} < K_{\text{krit}}$) and the further stabilization ($K_{\text{eff}} > K_{\text{krit}}$).

Towards SEI maintenance: $K_{\text{crit}}$ distribution and elemental correlation

From a functional perspective, given the inevitable formation of an initial solid-electrolyte-interphase (SEI) during battery assembly, the goal of solid-state battery design should be to focus on designing the initial SEI’s such that they don’t grow during cycling. Fig. 2 illustrates a computational process to determine such ideal initial SEI’s. In short, an ensemble of 50 randomly initialized decision trees is trained to predict $K_{\text{crit}}$ of the ground state phase at any input composition. The ensemble is then used to fine-tune the composition of known parent glass phase sulfides with good lithium-ion conductivity. The resulting child compositions should be minimally changed from the parent compositions but modulated to decrease the interface $K_{\text{crit}}$ to as low of values as is possible.

Over 80 000 SEI’s were simulated using pseudo-binary calculation of over 20 000 electronic insulators from the Materials Project with four well known ceramic-sulfide solid electrolytes – Li$_{10}$GeP$_2$S$_{12}$ (LGPS), Li$_{10}$SiP$_2$S$_{12}$ (LSPS), Li$_7$P$_3$S$_{11}$ (LPS), and Na$_2$P$_3$S$_{11}$ (NPS). Each data point consisted of the initial SEI composition, formed by simulating a chemical reaction between the SSE and the coating, and the $K_{\text{crit}}$ corresponding to stabilizing the initial SEI at a cathode voltage (e.g., from 2 V to 4 V) vs. lithium.
metal. In other words, $K_{\text{crit}}$ indicates how much local mechanical constriction is required to stop the initial SEI from decomposing when cycled to the cathode voltage region. A lower $K_{\text{crit}}$ thus also suggests that the interface is more likely to be stabilized in further cycling steps even if the initial local $K_{\text{eff}}$ is less than $K_{\text{crit}}$ as the electrochemical evolution will more easily increase the local $K_{\text{eff}}$ beyond $K_{\text{crit}}$.

Fig. 3(a) shows the distribution of $K_{\text{crit}}$ determined for interfaces of LPS paired with the various coating materials. The distribution has a mean, median, and standard deviation of 18.5, 14.8, and 13.0 GPa, respectively. The large standard deviation indicates that the value of $K_{\text{crit}}$ of an initial interface can vary considerably with composition and, thus, allows for substantial engineering to optimize the interface’s susceptibility to mechanical constriction within the constrained ensemble framework. Similarly, Fig. S1(a-c) (ESI†) show the $K_{\text{crit}}$ distribution for interfaces with LGPS, LSPS, and NPS. Fig. S1(d) (ESI†) depicts these statistical figures of merit (mean, median, standard deviation) for each SSE interface distribution. It remains true for each SSE that the distribution is quite broad, implying there is ample opportunity for engineering ideal interfaces with low $K_{\text{crit}}$ values.

The correlation between the elemental composition of SEIs and $K_{\text{crit}}$ is depicted in Fig. 3(b). Those atomic species with negative correlation (red) tend to lower $K_{\text{crit}}$ as their atomic fraction is increased. Conversely, those atomic species with positive correlation (blue) tend to raise $K_{\text{crit}}$. The highly electronegative oxygen, fluorine, and chlorine are seen to be amongst the most beneficial atomic species for maintaining a low $K_{\text{crit}}$ at the cathode voltage. This is likely attributed to these anions having high voltage redox potential. In other words, SEIs with high levels of oxygen, fluorine, and chlorine are less prone to delithiation in the considered voltage ranges. Additionally, V, Mo, and W are also seen to improve the stability of high voltage SEIs, presumably indicating that these atomic species are also unlikely to undergo reactions below 4 V.

**Designing SEIs with low $K_{\text{crit}}$**

In light of these results, it is apparent that initial SEIs could be designed to minimize $K_{\text{crit}}$ and hence decrease growth of the SEI upon cycling. In evaluating interfaces that are likely to occur in solid-state cathode, such as between LPS and common cathode materials, Fig. S2 (ESI†) shows that the $K_{\text{crit}}$ tends to be non-linear, oscillatory, with multiple local minima vs. composition. This highly nonlinear relationship between $K_{\text{crit}}$ and composition suggests that optimizing $K_{\text{crit}}$ directly is difficult, considering that the composition space is very high dimensional.

To computationally capture the nonlinear effects of composition on $K_{\text{crit}}$, we utilized an ensemble model of decision trees. The model was trained on the previously determined SEI data to predict $K_{\text{crit}}$. The conditional statements of each node are learned by the tree to most accurately reflect the underlying composition dependence of $K_{\text{crit}}$. The minimum unit of an example tree section illustrated in Fig. 4(a), together with the tree with more layers and branches depicted in Fig. S3 (ESI†), confirms the importance of electronegative anions for maintaining a low $K_{\text{crit}}$. For example, the left most branch at each layer implements a conditional statement on O or Cl. If the material has a low composition of those anions, the tree is with a higher predicted value of $K_{\text{crit}}$. Similarly, the right most branch states that if the material has a high composition of Li, F, Al, the tree is with a lower predicted value of $K_{\text{crit}}$. While each node corresponds to the evaluation of a single element, each element may be evaluated many times throughout the tree. Thus, each tree represents an approximate mapping of the $K_{\text{crit}}$ manifold. The ensemble then performs a weighted average of the predicted manifolds for the final value, $i.e.$, to predict a value for a given composition and a weighted average then gives a final prediction for the input composition. As shown in Fig. 4(b), which compares the predicted and actual $K_{\text{crit}}$ values for a test set of 10 000 SEIs, this approach can readily capture the complicated compositional dependence, especially in the lower range of $K_{\text{crit}}$. The root mean square error (RMSE) for this test set is 2.59 GPa.

Using the trained decision tree ensemble, we can modify the composition of known lithium-ion conductors to decrease the $K_{\text{crit}}$ and improve the SEI stability. To help us understand the role of each element on interface $K_{\text{crit}}$ we applied a feature-attribution model SHAP, or Shapley additive explanations, 17 to our trained ensemble to extract the influence of the composition of each element on the $K_{\text{crit}}$. More technical details about the
optimization method is described in the Methods. We tested various materials with changing levels of restriction on the elemental composition to better understand how $K_{\text{crit}}$ will change based on the doping amount. Table S1 (ESI†) shows compositional optimization of 10 well known sulfide glass SSE’s and the corresponding child phases where the composition has been changed to lower the interface $K_{\text{crit}}$ to LSPS. Child A, B, and C correspond to maximum allowed changes in the stoichiometry of each element as 0.05, 0.1, or no constraint, respectively. Note that in practice it is preferable to make the minimum possible compositional changes to lower the $K_{\text{crit}}$, as the ionic conductivity will likely fall quickly as composition is deviated. In Table S1 (ESI†), the known sulfide glasses have $K_{\text{crit}}$ values that are characteristically on the order of 15–17 GPa from our prediction. These values decrease to 10–13 GPa in child A phases, 6–9 GPa in child B phases, and 1.5–2 GPa in child C phases. This $K_{\text{crit}}$ decrease is expected to remarkably improve interface stability in actual solid-state battery cells, as the local mechanical constriction $K_{\text{eff}}$ is not necessarily at a sufficiently high level, depending on materials and battery assembly approaches.

We now use the first parent compound and its child compositions in Table S1 (ESI†) as an example to illustrate how our platform finds new compositions with lowered $K_{\text{crit}}$. Fig. 4 Decision tree method and results for composition design. (a) Illustration of the structure and mechanism of decision trees. The input composition is parsed through a hierarchy of criteria on the composition and generates output (see Fig. S3 for larger illustration, ESI†). (b) Comparison of the predicted $K_{\text{crit}}$ by the decision tree and the calculated $K_{\text{crit}}$ at the interface to LSPS at 4 V from our constrained ensemble ab initio platform, with a root-mean-square error (RMSE) of 2.59 GPa. (c) Optimization from (top to bottom) parent glassy material to child A (5% compositional change) to child B (10% compositional change) to child C (unlimited compositional change). The individual influence of each element is calculated by the SHAP model. Three intermediate states during the optimization process illustrate how the O, F, and Cl elements are introduced into the optimized Child A material as suggested by the SHAP model.
In Fig. 4c, the length of each bar (in the unit of GPa) approximates how much the corresponding element increases (red) or decreases (blue) the $K_{\text{crit}}$. The SHAP model analyzed that for the parent glass of Li$_{0.364}$P$_{0.110}$S$_{0.388}$I$_{0.138}$, the absence of O, F and Cl are dominantly responsible for the relatively high $K_{\text{crit}}$, as the first three long red bars are with O = 0, F = 0 and Cl = 0 in Fig. 4c, indicating that the presence of O, F and Cl in the compound will reduce $K_{\text{crit}}$. Meanwhile, the blue bars for existing elements of Li, I, P in the parent glass indicate that having them as existing nonzero compositions is not in conflict with the optimization goal of reducing $K_{\text{crit}}$.

It’s natural to understand that the optimization effort should be given to the features with red bars, since the length of red bar also roughly indicates how much we can further decrease the $K_{\text{crit}}$ if we fully optimize the composition of the corresponding elements starting from a given composition. We take an approach to efficiently optimize $K_{\text{crit}}$ by focusing on the element corresponding to either the longest red bar at each step, or the 2nd longest red bar if the longest bar element has just been updated in the previous step (i.e., a pseudo-longest red bar). The actual optimization process of $K_{\text{crit}}$ is also influenced by the concurrent composition of other elements brought by changing the composition of one element. For example, the child A, B, C compositions bring down the $K_{\text{crit}}$ from the 16.07 GPa (parent glass) to 12.07 GPa (Child A), 6.89 (Child B), and 2.05 GPa (Child C) with an increasing modification range of compositions. F, O and Cl are first introduced to Child A composition of Li$_{0.324}$P$_{0.098}$S$_{0.346}$I$_{0.138}$Cl$_{0.026}$F$_{0.042}$O$_{0.040}$, because these three elements all appear as either the longest or the pseudo-longest red bar in the initial intermediate optimization steps from the Parent toward the Child A composition (Fig. 4c), with Li, P, S, I compositions being forced to make slight concurrent decreases. The algorithm is very efficient in finding the right composition, as the intermediate compositions after the first 3 intermediate steps are very close to the Child A composition, with the remaining 17 intermediate steps just as a fine-tune of the composition. Through the composition optimization from Parent to Child A, B, C compounds, the total length of red bars shrinks and the $K_{\text{crit}}$ decreases generally. At Child C with unconstrained compositional change, the red bars are very short, suggesting a very limited remaining room to further decrease $K_{\text{crit}}$ through compositional modification. The general trend of adding more F, O, Cl, Br for reduced $K_{\text{crit}}$ as predicted by the SHAP model is also in good agreement with the elemental correlation map (Fig. 3(b)).

Using the predicted $K_{\text{crit}}$ of SEI at 4 V, we further predict the 4 V interface (to LGPS) decomposition energy for the SEI under various levels of composition modification (Fig. 5(a)). Similar to $K_{\text{crit}}$, the interface decomposition energy also tends to decrease with the level of modification. A $K_{\text{eff}}$ of 5 GPa can further lower...
the interface decomposition energies from that at 0 GPa while keeping the similar decreasing trend with composition modification. A $K_{\text{eff}}$ of 10 GPa can stabilize child materials at the 4 V interface to LGPS with a composition modification larger than 2.5%, for which $K_{\text{eff}} > K_{\text{crit}}$ is satisfied due to the reduced $K_{\text{crit}}$ by composition modifications (Fig. 5b). With a 15 GPa $K_{\text{eff}}$, all the materials, including the parent material will be stabilized. The trends described in Fig. 5a and b are qualitatively the same for interfaces calculated at 2 V and 3 V (Fig. S4, ESI†), except for some quantitative changes by voltage. Note that the $K_{\text{crit}}$ at 2 V are all below 4.6 GPa and that at 3 V are all above 4.9 GPa at all modifications.

Testing the model by experiment

To further test the predictability of our model, we synthesized the parent and predicted child glasses of the first row of Table S2 (ESI†) and performed cyclic voltammetry (CV) test based on batteries on normalized cathode weight, area and composition, as shown in Fig. 5(c) (Methods). Two decomposition current peaks labelled as H_L and H_R are observed at 3 V and 3.5 V, respectively. The pre-peak H_L is found to decrease monotonically with modification, while the broad peak H_R is oscillatory (Fig. 5(d)). We found that the predicted self-decomposition energy at 3.5 V at $K_{\text{eff}} = 0$ GPa in Fig. 5(b) exhibits a very similar trend to the measured H_R curve, suggesting that the change of broad peak intensity around 3.5 V with composition modification should be mainly caused by the self-decomposition current of the glasses in the initial charge, where the decomposition happens at the glass region near the many inevitable pores in the cold pressed pellet,18 where $K_{\text{eff}} = 0$ GPa. Note that without glasses, the control cathode of LGPS mixed only with amorphous conductive carbon shows a broad peak of decomposition current below 9 $\mu$A in the CV test, which defines the background contribution from the LGPS self-decomposition to the measured current that is independent from the modification to the glasses. Thus, the pre-peak H_L has to be contributed by the interface decomposition between LGPS and glass at a weak local mechanical constriction in the initial charge, consistent with the general decreasing trend for both curves of the measured H_L (Fig. 5d) and the predicted interface decomposition energy at a low $K_{\text{eff}}$ (Fig. 5a and Fig. S4, ESI†) versus modifications.

The fact that all the H_L pre-peaks are around 3 V suggests that the local $K_{\text{eff}}$ at the interface in the initial charge is around 4.6 to 4.9 GPa so that it is higher than all the $K_{\text{crit}}$ at 2 V (Fig. S4a, ESI†) but lower than all the $K_{\text{crit}}$ at 3 V (Fig. S4b, ESI†). Such an initial interface $K_{\text{eff}}$ is decided by the mechanical property of the two materials that form the interface (glass sulfide and LGPS here) and the processing and testing procedures (hand mixing, cold initial and stack press here) that define the interface contact. Only when the voltage is charged approaching 3 V, the local $K_{\text{eff}}$ starts to be smaller than $K_{\text{crit}}$ at the interface, as $K_{\text{crit}}$ increases with charge (comparing Fig. 5b and Fig. S4ab, ESI†), so that the interface decomposition happens. However, after the initial charge, the decomposition current almost vanishes in discharge and the following CV cycles below 4 V (Fig. S5, ESI†), suggesting that the local $K_{\text{eff}}$ of the glass–LGPS mixture upon the initial self and interface decompositions is increased beyond the $K_{\text{crit}}$ at all modifications both at the interface ($K_{\text{crit}} < 14$ GPa, Fig. 5b) and inside the glass ($K_{\text{self}} < 6$ GPa, Fig. S6, ESI†). Therefore, it is through these local decompositions in the initial charge that the order of $K_{\text{eff}}$ and $K_{\text{crit}}$ is switched at some low $K_{\text{eff}}$ regions, leading to the largely suppressed decomposition in further cycling, where the local $K_{\text{eff}}$ is increased beyond 14 GPa throughout the charged pellet and $K_{\text{eff}} > K_{\text{crit}}$ is generally satisfied below 4 V. Note that through the comparison between computation and experiment an upper limit of $K_{\text{eff}}$ value around 15 to 20 GPa is also estimated for crystalline phase of LPS12 and LGPS,1,4,13 Such a $K_{\text{eff}}$ is largely decided by the mechanical modulus of the electrolyte materials, which is similar for crystalline19 and amorphous20 sulfides.

We have shown that within the constrained ensemble that is unique to solid-state batteries, the use of randomly initialized decision tree ensembles has significant potential for optimizing SEIs at multiple solid–solid interfaces. Fine-tuning the SEI composition based on the decision tree results has been able to reduce the $K_{\text{crit}}$ significantly. Although the examples given here are tested at an experimental condition with strong mechanical constriction (see Methods) so that the decomposition current is largely inhibited after the initial charge regardless of the composition, in practice, the composition modification strategy for the electrolyte and coating materials could greatly reduce the burden of void elimination procedures during battery assembly and micro-crack prevention approaches during battery test, favoring the ideal manufacturing conditions at weak mechanical constriction. As a result, the method opens the door to realizable solid-state battery design and production based on mechanical constriction induced metastability.

Methods

Experiment

The amorphous solid electrolytes (P1-C6) are synthesized by high energy ball milling at 350 rpm. Stoichiometric amounts of Li$_2$S (> 99.9% purity, Alfa Aesar), P$_2$S$_5$ (> 99% purity, Sigma Aldrich), LiI (> 99% purity, Sigma Aldrich), LiF (> 99.9% purity, Sigma Aldrich), and Li$_2$O (> 97%, Sigma Aldrich) were milled for 20 hours. A total of 1 mg LGPS–amorphous SE–carbon composite cathode (weight ratio = 45 : 45 : 10) was used as the cathode. The composite was mixed and rolled into a thin film by a dry method with the addition of 5% PTFE. The diameter of the cathode film is 0.49 cm$^2$. The area and weight of all batteries were kept constant for all CV measurements. 120 mg LGPS was applied as the electrolyte while Li/graphite (capacity ratio = 2.5 : 1) was applied as an anode. The battery of Li/graphite–LPSCl–LGPS-cathode was pressed in a homemade pressurized cell, where LPSCl represents Li$_{2}$S$_{4}$Cl$_{1.5}$ that is used to guarantee a stable anode interface following our
recent approach. The batteries were initially pressed at 467 MPa while the operating pressure was kept at 250 MPa during battery test to ensure the mechanical constriction. In the CV test, a voltage ranges from the open circuit voltage to 8.5 V was applied with the scan rate of 0.1 mV s⁻¹. The control sample labelled as LGPS + C in Fig. 5 is constructed with the similar procedure but using only 0.5 mg of LGPS + C (weight ratio = 90:10) as cathode, without mixing with the amorphous glass.

High-throughput computation

Thermodynamical chemical decomposition reactions between materials (DFT data obtained from Materials Project) and an electrolyte were first calculated to generate an interphase for each material/electrolyte combination by constructing phase diagram using the Python Materials Genomics (pymatgen) library with the binary-search algorithm. The electrochemical stability of each interphase is then calculated from its grand potential phase diagram to obtain decomposition energy and reaction string for calculating \( K_{\text{crit}} \) by eqn (2) and (3). All \textit{ab initio} calculations of phase data were obtained following the Material Project calculation guidelines in the Vienna \textit{ab initio} Software Package (VASP). The mechanical constriction induced metastability calculations were performed on the computational platform we developed following the perturbation or Lagrangian optimization methods.

Decision tree learning

A decision tree consists of hierarchical computation (decision) nodes. A typical decision tree is depicted in Fig. S3 (ESI†). The input data to the decision trees is in the form \((X, y) = \{(x_1, x_2, \ldots, x_n, y)\}\) where \(x_i\) are the features and \(y\) is a target value. The decision tree can perform both the regression and classification tasks, depending on whether the nature of target variable \(y\) being continuous or a finite number of classes. Starting with the input features, each node of the tree applies a conditional statement on the value of a feature, then moves to a subsequent node based on the truth of that statement. The optimization of the tree includes choosing both the feature and a conditional statement on the value of a feature, then moves to the element with the second-largest SHAP value. Due to the composition constraint, it is possible again and optimize the next element with the most positive SHAP value. We compute the SHAP values and choose the element with the most positive influence on the \( K_{\text{crit}} \). We then optimize the composition of this chosen element under the given constraint. With the composition optimized, we compute the SHAP value again and optimize the next element with the most positive SHAP value. Due to the composition constraint, it is possible that the element that is just updated still has the highest SHAP value. In this case, we move to the element with the second-highest SHAP value (pseudo-longest red bar in Fig. 4c) in order to proceed the optimization. The optimization process ended until the \( K_{\text{crit}} \) cannot be further reduced.

Conflicts of interest

The authors declare no competing financial or non-financial interests.

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