Onset of sliding across scales: How the contact topography impacts frictional strength

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When two solids start rubbing together, frictional sliding initiates in the wake of slip fronts propagating along their surfaces in contact. This macroscopic rupture dynamics can be successfully mapped on the elastodynamics of a moving shear crack. However, this analogy breaks down during the nucleation process, which develops at the scale of surface asperities where microcontacts form. Recent atomistic simulations revealed how a characteristic junction size selects if the failure of microcontact junctions either arises by brittle fracture or by ductile yielding. This work aims at bridging these two complementary descriptions of the onset of frictional slip existing at different scales. We first present how the microcontacts failure observed in atomistic simulations can be conveniently “coarse-grained” using an equivalent cohesive law. Taking advantage of a scalable parallel implementation of the cohesive element method, we study how the different failure mechanisms of the microcontact asperities interplay with the nucleation and propagation of macroscopic slip fronts along the interface. Notably, large simulations reveal how the failure mechanism prevailing in the rupture of the microcontacts (brittle versus ductile) significantly impacts the nucleation of frictional sliding and, thereby, the interface frictional strength. This work paves the way for a unified description of frictional interfaces connecting the recent advances independently made at the micro- and macroscopic scales.

INTRODUCTION

The rapid onset of sliding along frictional interfaces is often driven by a similar dynamics than the one observed during the rupture of brittle materials. Just like a propagating shear crack, slipping starts and the shear stress drops in the wake of a slip front that is moving along the interface. This analogy particularly suits the observed behaviors of frictional interfaces at a macroscopic scale and explains that the earthquake dynamics has been studied for many decades as the propagation of shear cracks along crustal faults [1–3].

In this framework, pioneer cohesive approaches have been developed to study numerically the propagation of slip fronts, for which the frictional strength is modeled as progressively dropping with interfacial slip (often referred to as slip-weakening models) [4–7]. Using high-speed camera and photo-elasticity, slip fronts were later observed experimentally along interfaces bounding two blocks of PMMA. These “laboratory-earthquakes” confirmed the dynamics predicted by the early cohesive models [8–10].

Recent experiments [11] quantitatively demonstrated how Linear Elastic Fracture Mechanics (LEFM) perfectly describes the evolution of strains measured at a short distance from the interface during the dynamic propagation of rapid slip fronts. From this mapping, a unique parameter emerges, the equivalent fracture energy $G_c$ of the frictional interface, which was later used to rationalize the observed arrest of slip fronts in light of the fracture energy balance criterion [12, 13]. The same framework was also successfully applied to describe the failure of interfaces after coating the surface with lubricant [14]. Despite a reduction in the force required to initiate sliding, the equivalent fracture energy measured after lubrication was surprisingly higher than for the dry configuration [15]. This apparent paradox in the framework of LEFM is expected to arise during the nucleation phase, which is controlled by the microscopic nature of friction and contact. At the microscale, surfaces are rough and contact only occurs between the surface peaks, resulting in a very heterogeneous distribution of the sliding resistance [16, 17].

A class of laboratory-derived friction models [18–20] has been successfully used to rationalize some key aspects of the rupture nucleation along frictional interfaces, particularly in the context of earthquakes (critical length scales at the onset of frictional instabilities [19, 21–24], speed and type of the subsequent ruptures [25–29]). The so-called rate-and-state formulations are empirically calibrated to reproduce the subtle evolution of friction observed during experiments [16]. A direct connection with the physics of the microcontacts and their impact on the frictional strength remains however an open question and explains the increasing interest for physics-based interpretations of the rate-and-state friction laws [30–33].

To rationalize the friction coefficient of metal interfaces, Bowden and Tabor [34, 35] suggested that the microcontact junctions represent highly confined regions yielding under a combination of compressive and shear stresses. Later, Byerlee [36] proposed an alternative for brittle materials, by assuming that slipping does not oc-
cur through the plastic shearing of junctions but rather by fracturing the microcontacts, which leads to a smaller value of the friction coefficient in agreement with the ones measured for rock interfaces. From atomistic calculations, Aghababaei et al. [37–39] recently derived a characteristic size of the microcontact junction $d^*$ controlling the transition from brittle fracture (of junctions larger than $d^*$) to ductile yielding (of junctions smaller than $d^*$). As sketched in Fig. 1, these brittle and ductile failure mechanisms co-exist along two rough surfaces rubbing together. From this permanent interplay, Frérot et al. [40] proposed a new interpretation of surface wear during frictional sliding, while Milanese et al. [41] discussed the origin of the self-affinity of surfaces found in natural or manufactured materials. The link between these different microcontact failure mechanisms and the macroscopic frictional strength of the interface remains however overlooked.

In this context, the objectives of this work consist in upsaling these recent advances in the microcontacts description and unravel the impact of the heterogeneous contact topography on the frictional strength. We first present how to approximate the microcontacts failure using a convenient cohesive model. The cohesive approach is then implemented in a high-performance finite element library and used to simulate the onset of sliding across two scales. At the macroscopic level, we study the ability of an interface to withstand a progressively applied shearing, i.e. its frictional strength, while at the microscopic scale, we observe how the failure process develops across the microcontact junctions. This study culminates by discussing how small differences in the interface conditions or the size of asperity junctions, only visible at the scale of the microcontacts, can nevertheless have a significant impact on the nucleation phase and the macroscopic frictional strength.

**PROBLEM DESCRIPTION**

We consider two linearly elastic blocks of height $h/2$ brought into contact along their longitudinal face of length $l$. As presented in Fig. 1, the two blocks are progressively sheared by displacing the top surface at a constant speed $\dot{\Delta}x$, while the bottom surface is clamped. In a Cartesian system of coordinates, whose origin stands at the left edge of the contacting plane, the boundary conditions of this elastodynamic problem correspond to

\[
\begin{align*}
\mathbf{u}(x, -h/2, t) &= 0 \\
\dot{u}_x(x, h/2, t) &= \dot{\Delta}x \\
u_y(0, y, t) &= u_y(l, y, t) = 0
\end{align*}
\]

and lead to a state of simple shear, for which the shear components of the Cauchy stress tensor are $\sigma_{xy} = \sigma_{yx} = \tau$. In Eq. (1), $\mathbf{u} = \{u_x, u_y\}$ corresponds to the displacement vector and $\Box$ denotes a time derivative. At time $t = 0$, the two continua initially at rest start being progressively loaded by a shear wave whose amplitude corresponds to $\Delta\tau = \mu/c_s \Delta_x$. $\mu$ is the elastic shear modulus and $c_s$ the shear wave speed such that $t^* = h/c_s$ is the wave travel time between the top and bottom surfaces. The elastodynamic solution existing under the boundary conditions (1) and in the absence of interfacial slip is given in Fig. S1 of the Supplemental Material [42]. As illustrated in Fig. 1, sliding nucleates at small scales from the rupture of the microcontact junctions which potentially involves several non-linear phenomena (cleavage, plasticity, interlocking). As discussed by Aghababaei et al. [37], atomistic models are particularly suited to simulate these phenomena in compari-
son to continuum approaches, but are conversely disconnected from the macroscopic dynamics. The objective of this work is precisely to study this macroscopic dynamics using a generic interface law built from the failure behaviors reported in a large set of atomistic calculations [37–39, 41, 43–45]. We hence rely on a continuum description of the two solids, while the complex interface phenomena and associated dissipative processes are assumed to be constrained at the contact plane and entirely described by a “coarse-grained” cohesive law deriving from a thermodynamic potential $\Phi$. As presented in the Supplemental Material (see Fig. S2), the shape of this cohesive law and its associated potential are chosen to mimic the failure behavior of microcontact junctions computed from atomistic simulations [37, 38, 43]. As sketched in Fig. 1, sliding is assumed to initiate at the edge of a weak spot (e.g. the largest non-contacting region) existing at the very left of our model interface with a size $w_0$. Moreover, the rough contact topography sketched in Fig. 1 is idealized as a regular pattern of contacting and non-contacting junctions of microscopic size $w \ll w_0$.

Capturing the multi-scale nature of the problem requires an efficient and scalable parallel implementation of the finite element method, capable of handling several millions of degrees of freedom on high-performance computing clusters. To this aim, we use our homemade open-source finite element software Akantu, whose implementation is detailed in [46, 47] and whose sources can be freely accessed from the c4science platform [48]. More details about the numerical method and the material properties used in this manuscript are provided as Supplemental Material, which namely defines the values of the Young’s modulus $E$, the Poisson’s ratio $\nu$ and the reference interface fracture energy $G_c^{ref}$.

CHARACTERISTIC LENGTH SCALES OF THE BRITTLE-TO-DUCTILE FAILURE TRANSITION

Next, we study the onset of slip along a uniform and homogeneous (i.e. without microcontact) interface of fracture energy $G_c$ and size $(l - w_0)$. Figure 2a presents the evolution of energies observed during a typical failure event, i.e. the applied external work $W_{ext}$, the elastic strain energy $E_{el}$, the energy dissipated by fracture $E_{frac}$, and the kinetic energy $E_{kin}$. During an initial phase, the elastic strain energy builds up in the system following the dynamics predicted in the absence of interfacial slip (Fig. S1) and depicted by the black dashed line. After several back and forth reflections of the shear wave, sliding nucleates at $x = w_0$, a propagating slip front breaks the interface cohesion and releases $E_{frac} = G_c(l - w_0)$. The asterisk marks in Figs. 2a-b simply distinguish the final value of energy obtained after the complete junction failure from its transient value, i.e. $E_i^* = E_i(t \gg t^*)$. After the complete interface failure, an eventual excess of mechanical energy ($W_{ext}^* - E_{frac}^*$) remains in the system and takes the form of elastic vibrations in absence of any other dissipative process.

Figure 2b describes the evolution of energies observed during another failure event, during which sliding initiates for a significantly lower applied external work, exactly balancing the energy dissipated in fracture ($W_{ext}^* = E_{frac}^*$). Perhaps surprisingly to some readers, these quantitatively different sliding events arise within two systems having identical elastic properties and interface fracture energy $G_c$. These different dynamics emerge solely from the size of the fracture process zone, which represents how stresses concentrate and damage spreads at the edge of the contact junction. When the size of the process zone $l_{pz}$ is comparable to the resisting junction size $(l - w_0)$, sliding motion develops along a damage band stretching over the entire length of the interface with an energy balance similar to the one observed in Fig. 2b. Conversely, if $l_{pz} \ll (l - w_0)$, sliding initiates in the form of a slip front propagating from $x = w_0$ and leading to a more violent rupture as described in Fig. 2a.

The limit of an infinitesimally small process zone corresponds therefore to a singular shear (mode II) crack, whose propagation initiates according to LEFM criterion $K_{II} = K_c$ [49]. $K_c$ is the interface fracture toughness, which can be computed from the fracture energy as

$$K_c = \sqrt{\frac{G_c}{1 - \nu^2}}. \quad (2)$$

$K_{II}$ is the stress intensity factor, which depends on the far-field shear stress $\sigma_{xy}^\infty$, the crack size $w_0$ and a dimensionless factor $\chi$ accounting for the geometry:

$$K_{II} = \chi \sigma_{xy}^\infty \sqrt{\pi w_0}. \quad (3)$$

For the edge crack configuration of interest, $\chi$ can be approximated as 1.12 [49], such that the rupture is expected to initiate when

$$\sigma_{xy}^\infty = \frac{1}{1.12} \sqrt{\frac{G_c}{\pi w_0 (1 - \nu^2)}}. \quad (4)$$

According to Griffith’s criterion, the failure of the interface arises once the associated drop of potential energy is sufficient to balance the energy dissipated into fracture [49, 50]. For the limit of a singular crack ($l_{pz} \ll (l - w_0)$), this condition can be estimated as

$$W_{ext}^* \geq E_{frac}^* + E_{el}^{lefm}. \quad (5)$$

In Eq. (5), $E_{el}^{lefm}$ is the strain energy at the onset of a singular rupture and is approximated using Eq. (4) as

$$E_{el}^{lefm} \approx \frac{1}{2\mu} \int_\Omega (\sigma_{xy})^2 d\Omega = \frac{G_c}{(1.12)^2 \pi w_0 (1 - \nu^2)} hl. \quad (6)$$

Finally, LEFM predicts that the size of the process zone at the onset of the rupture scales according to the
FIG. 2. The ratio of the process zone size to the length of the resisting junction mediates the work required to initiate sliding. (a) and (b) present two typical time evolutions of the energetic quantities prior to the rupture onsets, which occur, respectively, at $t = 92t^*$ and $t = 35t^*$. The two events share the same elastic properties and $G_c = 4G_{ref}^c$, but their respective interface cohesive laws lead to $l_{pz}/(l - w_0) = 3.5 \cdot 10^{-2}$ and $l_{pz}/(l - w_0) = 3.5$. The latter are visible on the associated shear stress profiles presented for the two interfaces before the onset of sliding. The dashed lines in (a) and (b) describe the theoretical build-up of elastic strain energy expected in the absence of interfacial slip according to the dynamics presented in Fig. S1 of the Supplemental Material. (c) Normalized external work required to initiate sliding between the two bodies as function of the ratio between the process zone size $l_{pz}$ and the resisting junction size $(l - w_0)$ for different types of interface properties and geometries. The ratio of $l_{pz}$ to the junction size selects the failure mechanism from strength-controlled ductile failure ($W^{*\text{ext}} = E_{\text{frac}}^{*}$) to toughness-controlled brittle rupture ($W^{*\text{ext}} \geq E_{\text{frac}}^{*} + E_{\text{El}}^{*}$).

The evolution described in Fig. 2c is equivalent to the transition reported in the framework of tensile fracture [51] from strength-controlled fracture ($F^{*\text{ext}} \sim \tau_c \cdot (l - w_0)$) for large process zones toward toughness-controlled failure with shorter process zones ($F^{*\text{ext}} \sim \sigma^{\infty}_{xy} \cdot l$), where $\sigma^{\infty}_{xy}$ is function of $K_c$ (see Eq. (4)). $F^{*\text{ext}}$ corresponds to the macroscopic force required to trigger sliding, i.e to reach the interface frictional strength.

As illustrated in Fig. 1, an equivalent brittle-to-ductile transition exists in the failure of the microcontact asperities. Aghababaei et al. [37] revealed how a characteristic junction size $d^* = \lambda G_c \tau_c^{-2} \mu$ mediates this transition from the brittle rupture of the apexes of junctions larger than $d^*$ to the ductile yielding of junctions smaller than $d^*$. In Eq. (8), $\lambda$ is a dimensionless factor accounting for the geometry (typically in the range of unity) and, therefore, $l_{pz}$ (Eq. (7)) corresponds
to the same characteristic length scale than \(d^*\) (Eq. (8)). Remarkably, there is a direct analogy between the brittle-to-ductile failure transition (controlled by \(l_{pz}\)) presented in Fig. 2c using the cohesive approach. The latter represents therefore a powerful tool to unravel the impact of the microcontacts failure on the macroscopic frictional strength of interfaces. Next, we select two types of interface properties with the same fracture energy \(G_c = G_c^{\text{ref}}\) and with process zone sizes that are much smaller than the size of the domain. We later refer to these two systems as interface \(A\) (\(l_{pz,A}/l = 9 \cdot 10^{-4}\)) and interface \(B\) (\(l_{pz,B}/l = 4.5 \cdot 10^{-2}\)). Under the homogeneous conditions considered in this section, the interfaces \(A\) and \(B\) rupture with a crack-like dynamics (as \(l_{pz} \ll l - w_0\)) at similar magnitudes of external work (see the blue circles in Fig 2, which are recalled in Fig 3).

In the next section, the frictional strength of interfaces having a heterogeneous microscopic topography is studied in light of the characteristic junction size \(d^*\). The size of the microcontact junctions \(w\) is chosen in order to discuss the cases where \(w\) is respectively larger/smaller than the characteristic junction size of the interfaces \(A/B\) (\(d^*_A < w < d^*_B\)). The characteristic junction sizes are computed using \(\lambda \equiv 3\) in Eq. (8), such that \(d^* \equiv l_{pz}\). This value of \(\lambda\) corresponds to the one estimated for three-dimensional spherical asperities in [37].

**ROUGH CONTACT TOPOGRAPHY AND FRICTIONAL STRENGTH**

As sketched in Fig. 1, two solids come into contact along a reduced portion of the interface, between the peaks of the microscopically rough surfaces. To model the effect of this very heterogeneous topography, we now introduce an idealized array of microscopic gaps and junctions of size \(w = 0.05w_0 = 0.005l\). In order to keep the total energy dissipated into fracture unchanged \((E_{\text{frac}} = G_c^{\text{ref}}(l - w_0))\), the fracture energy of the microscopic junctions is set to \(2G_c^{\text{ref}}\). As shown by the red circles in Fig. 3, the interfaces \(A\) and \(B\) have significantly different frictional strength in presence of the heterogeneous microstructure and interface \(A\) is roughly twice stronger than interface \(B\) \((W_{\text{ext},A}^* \approx 2.2W_{\text{ext},B}^*\) and \(F_{\text{ext},A}^* \approx 1.7F_{\text{ext},B}^*\)). This major difference is caused by the introduction of a new length scale \(w\) in the systems, which exactly stands between the characteristic length scales \(d^*_A\) and \(d^*_B\).

As presented in Fig. 4b, along interface \(B\) \((d^*_B > w)\), several microcontact junctions start damaging and slipping during the initial loading phase. The stress concentration at the edge of the critical junction spans several gaps and contacting asperities. Their individual properties are thereby homogenized within this large process zone and result in a quasi-homogeneous frictional response driven by the strength-dominated ductile failure. Conversely, for interface \(A\) \((d^*_A < w)\), the shear stress sharply concentrates at the very edge of the microcontact junctions (cf. Fig. 4a) whose local toughness directly controls the onset of failure. Indeed, applying Griffith criterion to this setup implies that an infinitesimal advance \(da\) of the rupture requires the release \(da \cdot 2G_c^{\text{ref}}\) of potential energy, which is theoretically twice larger than in the homogenized situation \((da \cdot G_c^{\text{ref}})\). Such toughening mechanism will therefore become stronger for larger contrasts between the local toughness of the microcontacts and the average macroscopic toughness of the interface.

**DISCUSSION**

Between two realistic rough surfaces in contact, a dense spectrum of junction sizes forms the real contact area, which often barely exceeds few percents of the apparent area of the contact plane [16, 40]. In this context, our results reveal how the characteristic junction size \(d^*\) gives the length scale over which the microscopic properties of the rough surfaces can be averaged to estimate the
FIG. 4. Zooming at the vicinity of the critical junction \((x = u_0)\) reveals the origin of the frictional strength difference between interface \(A\) and \(B\). Colors depict the shear stress profile existing before the onset of sliding while an artificial vertical displacement \(u_y(x, y) = u_x(x, y)\) is applied to help visualizing the slip profile along the interface (200 times magnification).

The shear strengths existing between the top and bottom surfaces are depicted with a gradation from black \((\tau^{\text{str}} = \tau_c)\) to white \((\tau^{\text{str}} = 0)\). (a) For interface \(A\), the shear stress and slip localize at the edge of the critical junction, magnifying its local toughness. (b) Along interface \(B\), frictional slip and shear stress stretch across several microcontact junctions and gaps within a larger process zone which averages the local frictional strengths. The sketches located in the top right of each plot associate the failure of the coarse-grained junctions simulated along the interfaces \(A\) and \(B\) to the corresponding failure mechanism of surface asperities discussed in Fig. 1.

The heterogeneous microstructure has a strong impact on the nucleation phase but is much less significant during the subsequent rupture dynamics. The interested readers can find further discussions about the rupture dynamics of such heterogeneous interfaces in the Supplemental Material.

CONCLUSION

The onset of sliding between two elastic blocks brought into contact arises at different scales. Macroscopically, a slip front nucleates at the critical spot of the interface and expands along the contacting plane, driven by the same dynamics than a propagating shear crack. At a smaller scale, at which surfaces are rough, sliding initiates by the progressive failure of the contacting asperities, which typically occurs either through the ductile yielding of the contact junctions or the brittle failure of the asperities. In this work, we study how these microscale failure mechanisms respectively impact the macroscopic frictional strength of the interface. To this aim, we construct a coarse-grained cohesive model and relate the characteristic length scale of the cohesive law \(l_{pz}\) to the characteristic junction size \(d^*\) controlling the brittle-to-ductile transition during microcontacts failure.

We then simulate the onset of sliding across two scales using large finite element models. At a macroscopic scale, we measure the frictional strength of the interface (i.e. the amount of shear force/external work required to trigger sliding), while at a microscopic scale we observe the failure of the microcontact junctions. Our work reveals how the selection of the microcontact failure mechanism has a direct and significant impact on the macroscopic frictional response. Indeed, the strength of the junctions smaller than \(d^*\) can be averaged (cf. responses of interface \(B\) in Fig. 3), whereas the strength of the microcontact junctions larger than \(d^*\) are individually impacting the macroscopic frictional behavior of the interface (cf. responses of interface \(A\) in Fig. 3). Any perturbation of this characteristic junction size (coating, lubrication) can thereby have a strong impact on the interface frictional strength.

In this context, the brittle-to-ductile transition discussed in this work might bring an interesting avenue to rationalize the “slippery but tough” behavior of lubricated interfaces discussed in the introduction. As reported by Bayart et al. [15], the lubrication significantly increases the critical slip distance \(\delta_s\) and the interface fracture energy \(G_c\). Moreover, a reduction of the interface adhesion also leads to an increase of the characteristic junction size \(d^*\) [44, 45]. Dry contact can hence be viewed as a strong but fragile interface, where slip initiates by a sharp concentration of the shear stress and damage zone at the edge of the microcontacts, followed by the abrupt brittle failure of individual microcontacts. After lubrication, the damage zone distributed over multiple microcontact leads to the strength-dominated ductile failure of several junctions, resulting macroscopically into a more slippery yet tougher interface.

The difference between \(d^*_A\) and \(d^*_B\) significantly impacts the frictional strength which cannot be systematically predicted from the homogenized interface properties. However, the dynamics of the subsequent ruptures propagating along the interfaces \(A\) and \(B\) are similar at a macroscopic scale and comply with the dynamic fracture predictions (LEFM) for homogenized interface properties (see Fig. S3 in the Supplemental Material). Hence,
frictional strength. Indeed, during the ductile yielding of asperities smaller than \( d^* \), the strengths of several neighboring microcontact junctions and gaps are averaged within the process zone. Conversely, the shear stress sharply concentrates at the edges of microcontact asperities larger than \( d^* \) and magnifies their individual toughnesses. In this context, any modification of the characteristic junction size \( d^* \) (lubrication, coating) or the microcontact topography (sanding) will thereby impact the macroscopic frictional strength (even if such modifications are only visible at a microscale and do not change the homogenized interface properties).

Finally, the observations and discussions presented in this manuscript also find implications in our understanding of the failure of heterogeneous media, particularly in the context of multi-scale and hierarchical materials, for which the microstructure organization can be tuned to enhance the overall material properties [52, 53].

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SUPPLEMENTAL MATERIAL

To prevent confusions with the core manuscript, labels referring to figures or equations of the Supplemental Material are preceded by the letter “S”.

In Eq. (S3), $\tau_c$ and $\delta_c$ are respectively the maximum strength and critical slip of the interface characterizing the exponential traction-separation law sketched in Fig. S2, for which the fracture energy corresponds to

$$G_c = \int_0^\infty \tau d\delta_c = c\tau_c \delta_c.$$  

FIG. S1. Elastodynamic solution in the absence of interfacial slip. The dynamic fields are mediated by the vertical propagation of a shear wave front characterized by $\Delta \tau = \mu/c_s \Delta \tau$. $t^* = h/c_s$ is the time needed by the front to travel between the top and bottom surfaces and $n \in \mathbb{N}$ is the total number of reflections observed at the top boundary.

Numerical method

The elastodynamic equation is solved with a finite element approach using a lumped mass matrix coupled to an explicit time integration scheme based on a Newmark-$\beta$ method [54]. The stable time step is defined as function of the dilatational wave speed $c_d$ and the spatial discretization $\Delta s$ as

$$\Delta t = 0.7 \frac{\Delta s}{c_d},$$  

with $\Delta s$ being typically set to $\frac{h}{1000}$ in this work. For the large simulations of interfaces with a heterogeneous microstructure, the discretization is brought to $\frac{h}{1000}$ leading to about 70M degrees of freedom. The virtual work contribution of the frictional plane is written as

$$\hat{W}(t) = \int_0^t \tau(x, t) \delta(x, t) dx,$$  

with $\hat{\square}$ denoting a “virtual” quantity and $\delta(x, t) = u_x(x, 0^+, t) - u_x(x, 0^-, t)$ being the interfacial slip between the top and bottom surfaces. The shear traction acting at the interface $\tau$ is assumed to derive from an exponential Rose-Ferrante-Smith universal potential $\Phi$ [55] and is expressed as

$$\tau = \frac{\partial \Phi}{\partial \delta_x} = \frac{\delta_x}{\delta_c} \tau_c e^{1 - \frac{\delta_x}{\tau_c}}.$$  

Material properties

The results are discussed in the manuscript with adimensional scales but the material properties of Homalite used in the simulations are given to the reader for the sake of reproducibility: Young’s modulus $E = 5.3$ [GPa], Poisson’s ratio $\nu = 0.35$, shear wave speed $c_s = 1263$ [m/s], and typical interface fracture energy $G_{c}^{ref} = 23$ [J/m$^2$].

Slip front dynamics

The objective of the manuscript is to study the impact of the microscopic roughness at nucleation. Hereafter, we briefly comment the subsequent rupture dynamics observed along the heterogeneous interfaces $A$ and $B$. As presented in the manuscript, the heterogeneous microstructure has a significant impact on the nucleation phase as different magnitudes of external force/work are required to onset sliding along these two interfaces. Nevertheless the subsequent rupture dynamics are macroscopically similar and comply with LEFM predictions for
FIG. S2. From left to right: Typical force versus slip profile observed during the shearing of two interacting asperities in atomistic simulations (see [38] for a detailed presentation of the method and setup). Such behavior can be conveniently described by the exponential cohesive law given in Eq. (S3) and derived from a Rose-Ferrante-Smith [55] type of universal binding potential. The exponential cohesive law (Eq. (S3)) allows for saving the cost of describing the fine details of asperity contact and, in return, the coarse-grained junctions can embed the microcontacts failure behavior into the macroscopic response of frictional systems. More notably, these coarse-grained junctions also reproduce the crucial observations of the atomistic simulations: the brittle-to-ductile transition in the failure of microcontact junctions controlled by an identical characteristic length scale (cf. Section Characteristic length scales of the brittle-to-ductile failure transition in the core manuscript).

homogenized interface properties. As the difference between $d_A^*$ and $d_B^*$, as well as the size of the microcontact asperities $w$ are buried within the near-tip region (region I) in Fig. 1 of the main manuscript), the square root singularity is governed by the average properties of the interface, i.e. by the average fracture energy $G_{c}^{\text{ref}}$.

In Fig. S3, the stress profiles are measured at a macroscopic distance ($h/25 \gg w$) from the contacting plane as it is the case during experiments [11, 13, 14]. In both situations, the stress profiles present the K-dominance predicted by LEFM for dynamic shear cracks with an associated dynamic energy release rate balancing the average fracture energy $G_{c}^{\text{ref}}$. The linear elastic stress solutions used in Fig. S3 are described in the next subsection.

One could argue from LEFM that two identical interfaces whose dynamic failures are mediated by the same fracture energy $G_{c}^{\text{ref}}$ should also rupture for the same amplitude of far-field shear stress. This correct statement for homogeneous interfaces breaks down in presence of microscopic heterogeneities, which strongly impact the rupture initiation (cf. red versus blue dots in Fig. 3 of the core manuscript). Therefore the propagating fronts nucleate under very different shearing conditions along interfaces $A$ and $B$. The subsequent dynamic ruptures propagate then at different velocities (faster along interface $A$) and explain the difference in stress amplitude existing between the two interfaces in Fig. S3. The presence of elastic waves in the stress profile of interface $A$ is a second difference visible in Fig. S3. These high-frequency radiations, which were studied in details in [57], arise only during the interplay of dynamic fronts with heterogeneities of size larger than the process zone, and therefore mainly for interface $A$. Nevertheless, their wavelength and amplitude will decay for microcontacts junctions smaller than the two orders of magnitude considered in our simulations and become out of the resolution of macroscopic experiments.

Dynamic fracture mechanics

For a detailed presentation of the dynamic fracture theory, the reader is redirected to the reference textbooks [2, 63, 64]. For a mode II shear crack moving at speed $v_c$, the dynamic energy balance is expressed from the dynamic stress intensity factor $K_{II}$ and a universal function of the crack speed $A_{II}$:

$$G_c = G = \frac{1 - \nu^2}{E} K_{II}^2 A_{II}(v_c),$$

(S5)

with

$$A_{II}(v_c) = \frac{\alpha_s v_c^2}{(1 - \nu)Dc_s^2},$$

(S6)

where $\alpha_{s.d} = 1 - v_c^2/c_{s.d}^2$, and $D = 4\alpha_d \alpha_s - (1 + \alpha_s^2)^2$. As for the static crack depicted in Fig. 1 of the main manuscript, stresses immediately ahead of a dynamic front are dominated by a square root singular contribution. The latter can be expressed in a polar system of coordinates $(r, \theta)$ attached to the crack tip and as func-
FIG. S3. At a macroscopic distance from the interface the evolutions of the stress fields observed during the dynamic failure of the heterogeneous interfaces A (top) and B (bottom) comply with LEFM predictions for an interface fracture energy corresponding to the average value $G_{ref}^c$. On the left panels, shear stress at the vicinity of the propagating slip front is mapped using the same color scale. To mimic the experimental measurements, the white lines highlight the position along which the components of the Cauchy stress tensor are presented on the right panels in red. The stress fields predicted by LEFM at the vicinity of a shear crack are plotted in blue for a fracture energy equal to $G_{ref}^c$. Note that the shift visible in the simulation profiles of $\sigma_{xy}$ is caused by the shear wave traveling ahead of an accelerating shear crack which is not included in LEFM solutions of Eq. (S7) [6, 56].

FIG. S4. Higher-order non-singular contributions (i.e. $n > -1/2$ in Eq. (S8)) have little influence on the stress solutions reported in Fig. S3. The red dots and blue curves are the same data presented in Fig. S3 for interface A. The computations of the blue curves include the square root singular contribution plus the first higher-order term ($n = -1/2, 1/2$ in Eq. (S8)). The green curves present the same predictions by considering only the singular contribution ($n = -1/2$).

The expression of the dynamic stress intensity factor $K_{II}$ [63]:

$$
\sigma_{xx} = -\frac{K_{II}}{\sqrt{2\pi r}} \frac{2\alpha_d}{D} \left\{ \left( 1 + 2\alpha_d^2 - \alpha_s^2 \right) \frac{\sin \frac{1}{2} \theta_d}{\sqrt{\gamma_d}} - \left( 1 + \alpha_s^2 \right) \frac{\sin \frac{1}{2} \theta_s}{\sqrt{\gamma_s}} \right\},
$$

$$
\sigma_{xy} = \frac{K_{II}}{\sqrt{2\pi r}} \frac{1}{D} \left\{ 4\alpha_d \alpha_s \frac{\cos \frac{1}{2} \theta_d}{\sqrt{\gamma_d}} - \left( 1 + \alpha_s^2 \right)^2 \frac{\cos \frac{1}{2} \theta_s}{\sqrt{\gamma_s}} \right\},
$$

$$
\sigma_{yy} = \frac{K_{II}}{\sqrt{2\pi r}} \frac{2\alpha_s (1 + \alpha_s^2)}{D} \left\{ \frac{\sin \frac{1}{2} \theta_d}{\sqrt{\gamma_d}} - \frac{\sin \frac{1}{2} \theta_s}{\sqrt{\gamma_s}} \right\},
$$

(S7)

with $\gamma_{s,d} = \sqrt{1 - (v_s \sin \theta/c_{s,d})^2}$ and $\tan \theta_{s,d} = \alpha_{s,d} \tan \theta$. 

Just as in the Williams series describing static cracks [65], non-singular contributions could be added to describe stresses evolution far from the tip (cf. region (III) in Fig. 1 of the main manuscript). As described in [11], the singular expressions of Eq. (S7) can be generalized to higher-order terms, as example for the shear stress component which becomes:

$$\sigma_{xy} = \sum_{n=-\frac{1}{2}, \frac{1}{2}, \frac{3}{2}, \ldots} A_n (2\pi r)^n \frac{1}{D} \left\{ 4\alpha_d \alpha_s (\gamma_d)^n \cos(n\theta_d) - (1 + \alpha_s^2)(\gamma_s)^n \cos(n\theta_s) \right\},$$  \hspace{1cm} \text{(S8)}

where $A_{-\frac{1}{2}}$ is equivalent to $K_{II}$ in Eq. (S7).

As the profiles of stress are measured above the contact plane, the first non-singular contribution $A_{\frac{1}{2}}$ is also considered during the mapping. The excellent agreement with LEFM predictions reported in Fig. S3 was obtained with

$$K_{II} = \sqrt{\frac{G_{c}^{\text{ef}} E}{(1 - \nu^2)A_{II}(v_c)}}$$  \hspace{1cm} \text{(S9)}

and by seeking for the position of the front $x_{tip}$, its propagation velocity $v_c$ as well as $A_{\frac{1}{2}}$ giving the best predictions of the simulated stress profiles according to a nonlinear least-squares regression [66–68]. As shown in Fig. S4, the non-singular contribution has a limited influence on the resulting mapping. This comparison with LEFM predictions suggests that dynamic slip fronts propagate along interfaces $A$ and $B$ with the same macroscopic fracture energy $G_{c}^{\text{ef}}$ corresponding to the homogenized value. A more quantitative comparison of the apparent fracture energy of heterogeneous interfaces could be the topic of a subsequent work and requires a precise and independent measurement of the rupture speeds.