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Pushing the Performance Limits of Electrical Contact Materials

Connecting Nano to Macro

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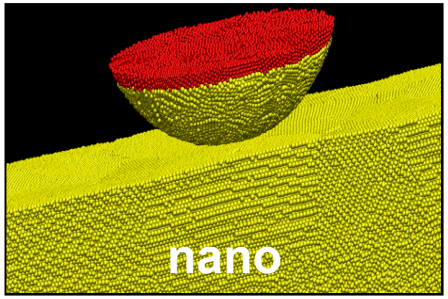

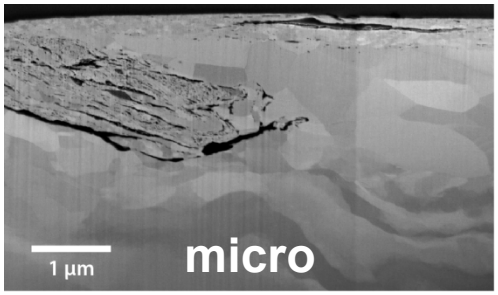

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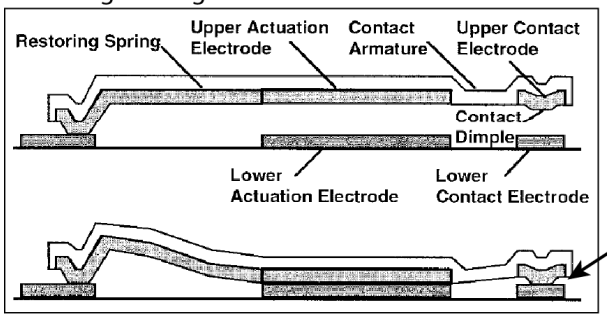
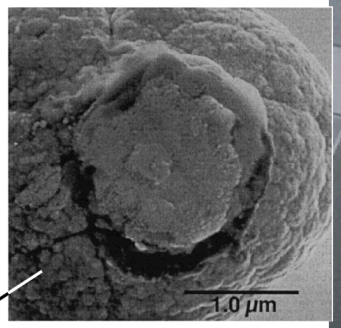
Metals are widely used tribological materials – electrical contacts

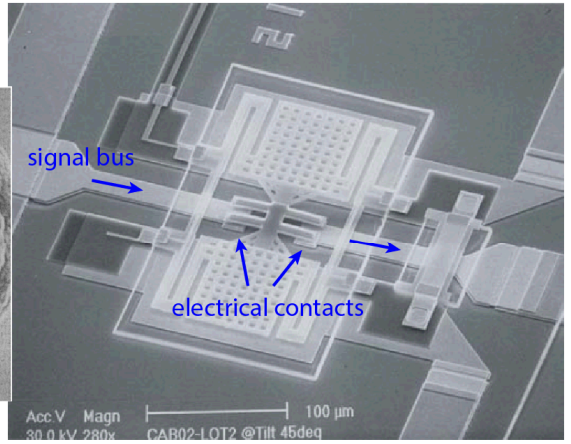
RF Micro Electromechanical Systems (MEMS)

switching GHz signals



Source: D. Hyman and M. Mehregany, IEEE Trans. & Pack. Tech. 22-3, 1999



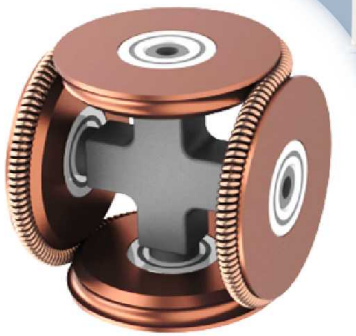


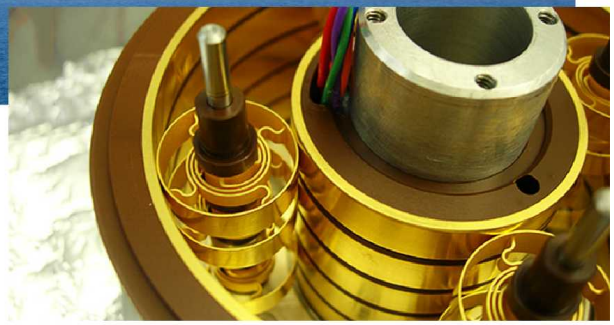
Source: Rockwell Scientific metal-metal switch

Electronics (e.g. PCB blade connectors):

200 - 500 nm thick electroless hard gold

Aerospace and Energy



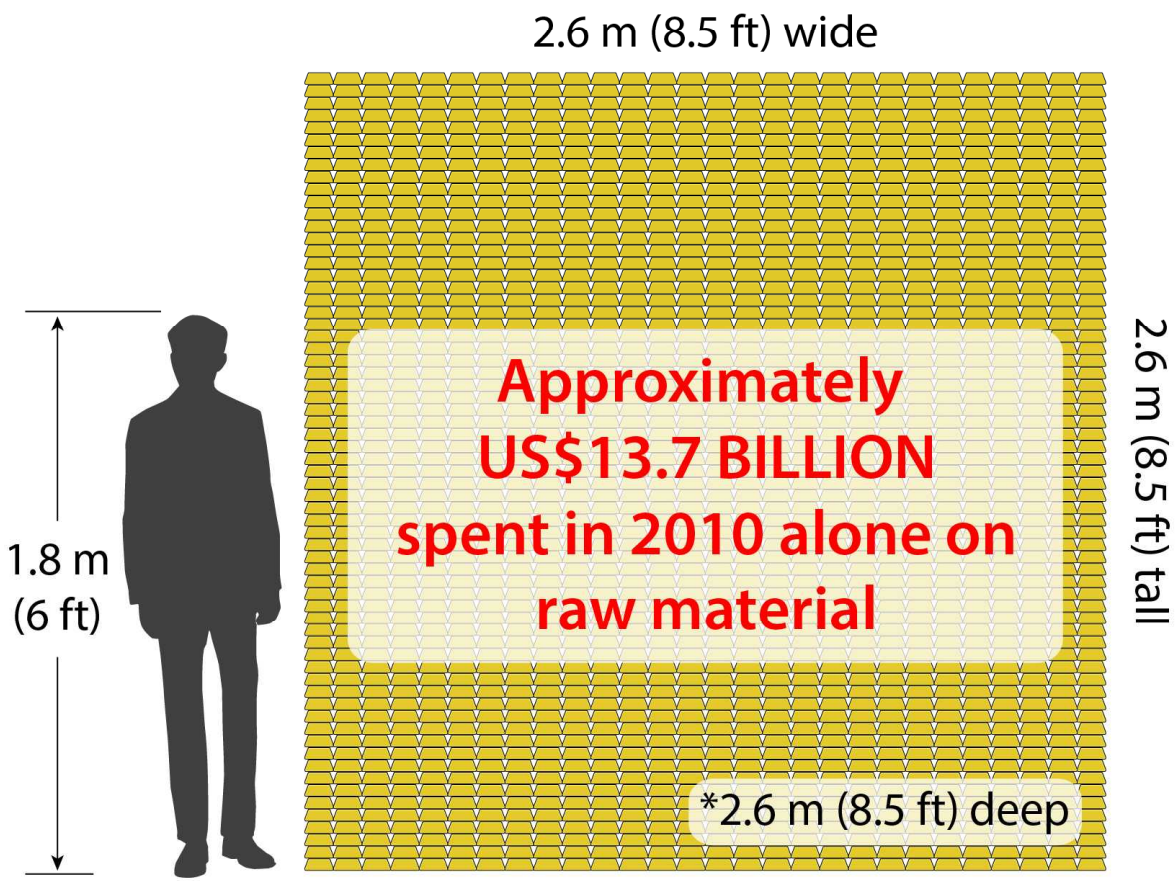


Source: Honeybee Robotics (http://www.honeybeerobotics.com/portfolio/rolling-contact-connector/)

The Gold Standard… how much gold you may ask? TONS per year

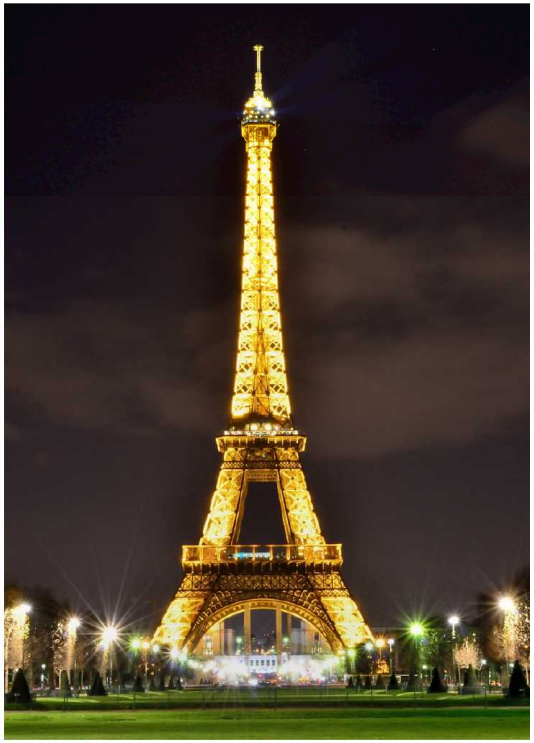
An estimated 300 metric tons/year of gold used in electronics related applications, most of it in electroplated connectors and contacts (11% of yearly amount mined)

Equivalent to a cube comprised of ~25,000 standard gold bars (12 kg/26.4 lb each)...



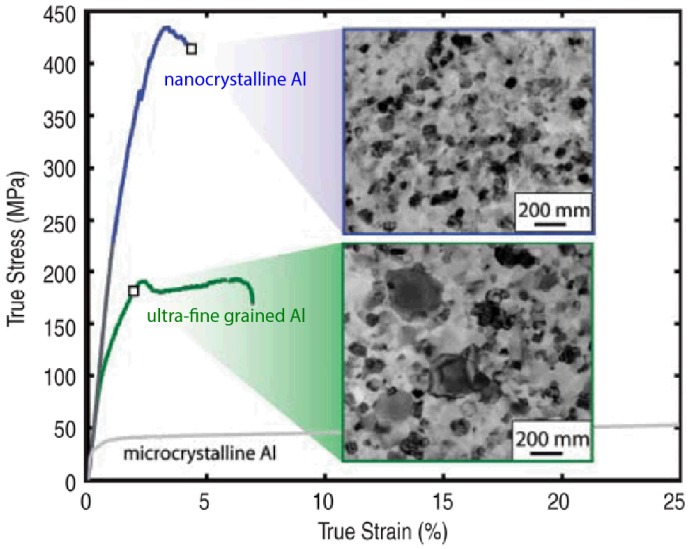
Reference: Gold Survey, Gold Fields Mineral Services Ltd., 2010

... or enough to clad the surface of the Eiffel Tower with 70 µm of pure gold ***every year***

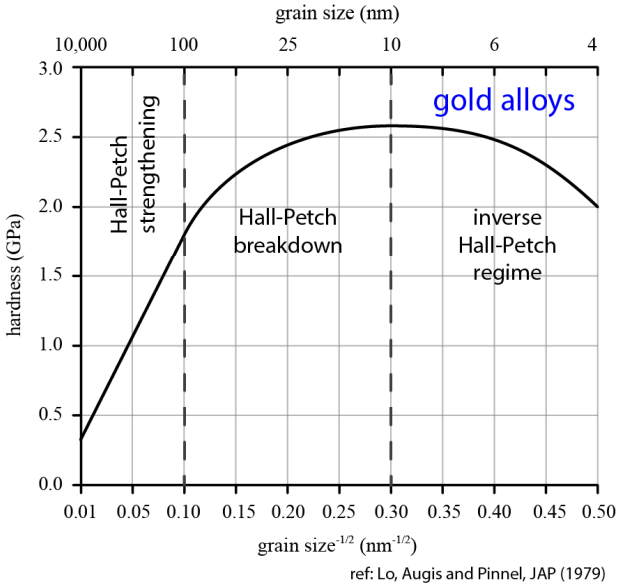


Engineering advantages of nanocrystalline (NC) alloys and metal-matrix composites

higher yield strength

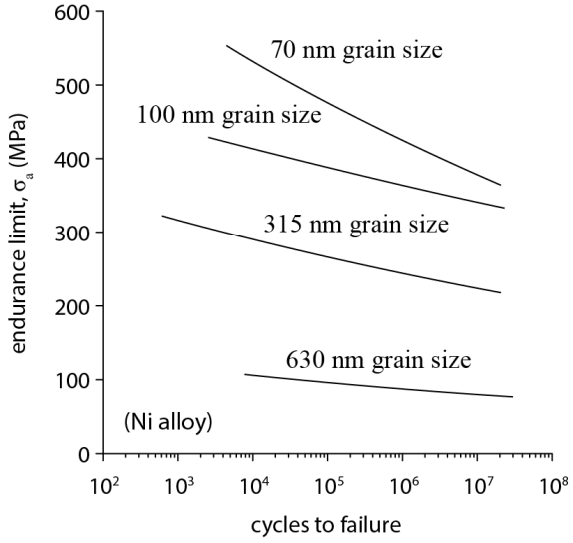


higher hardness

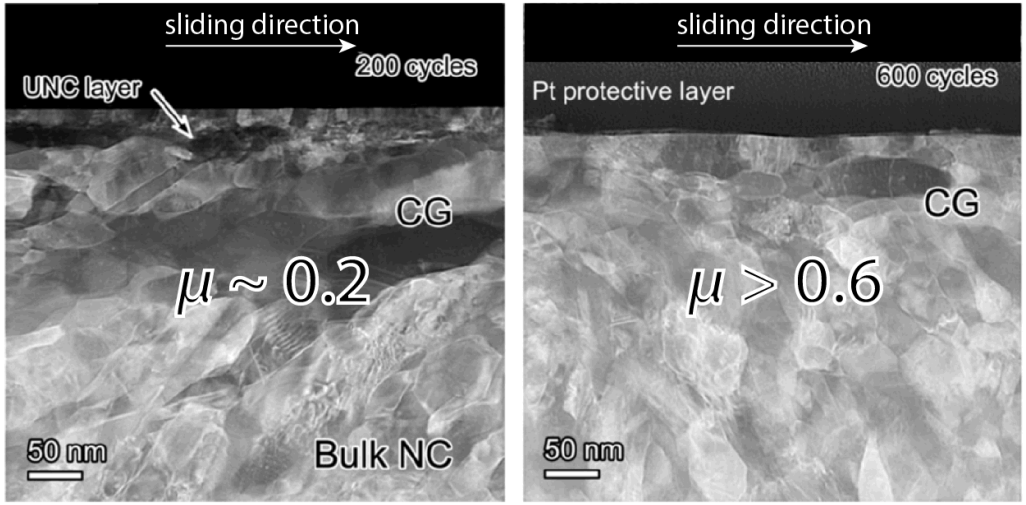


ref: Lo, Augis and Pinnel, JAP (1979)

higher fatigue strength (endurance limit)



lower friction and wear rates -- *but why?*

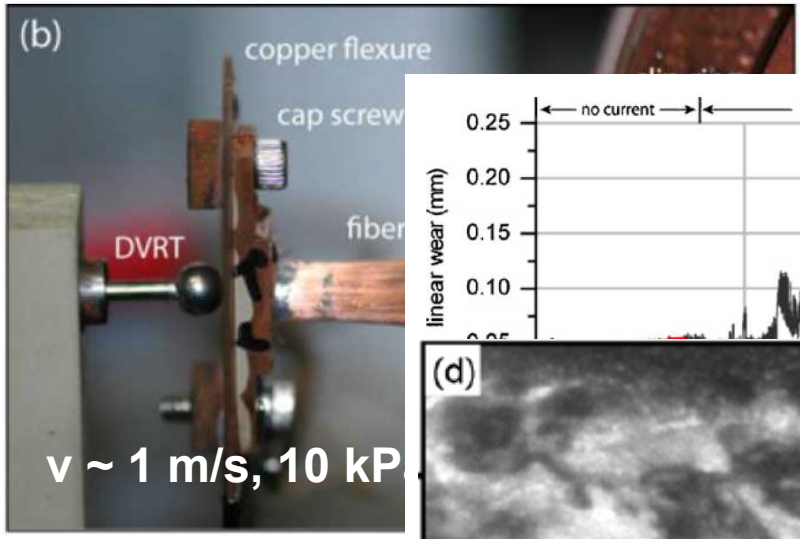


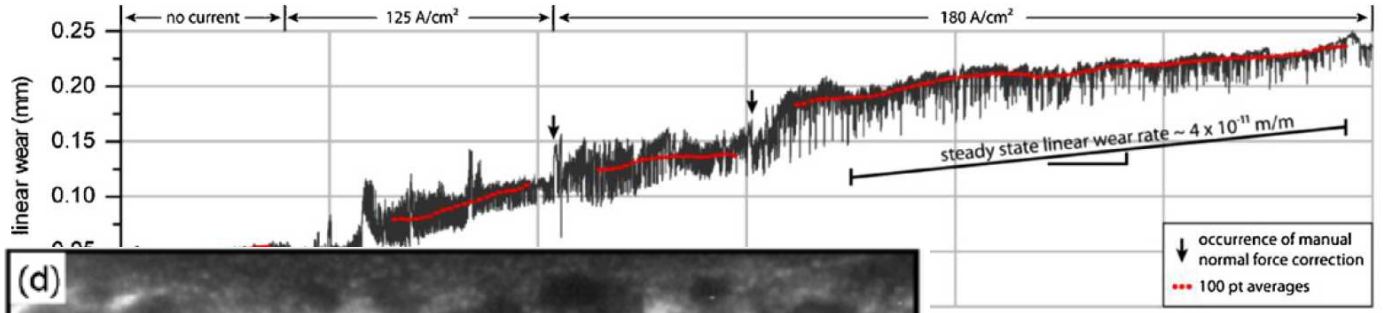
steady-state cross-sections of wear tracks

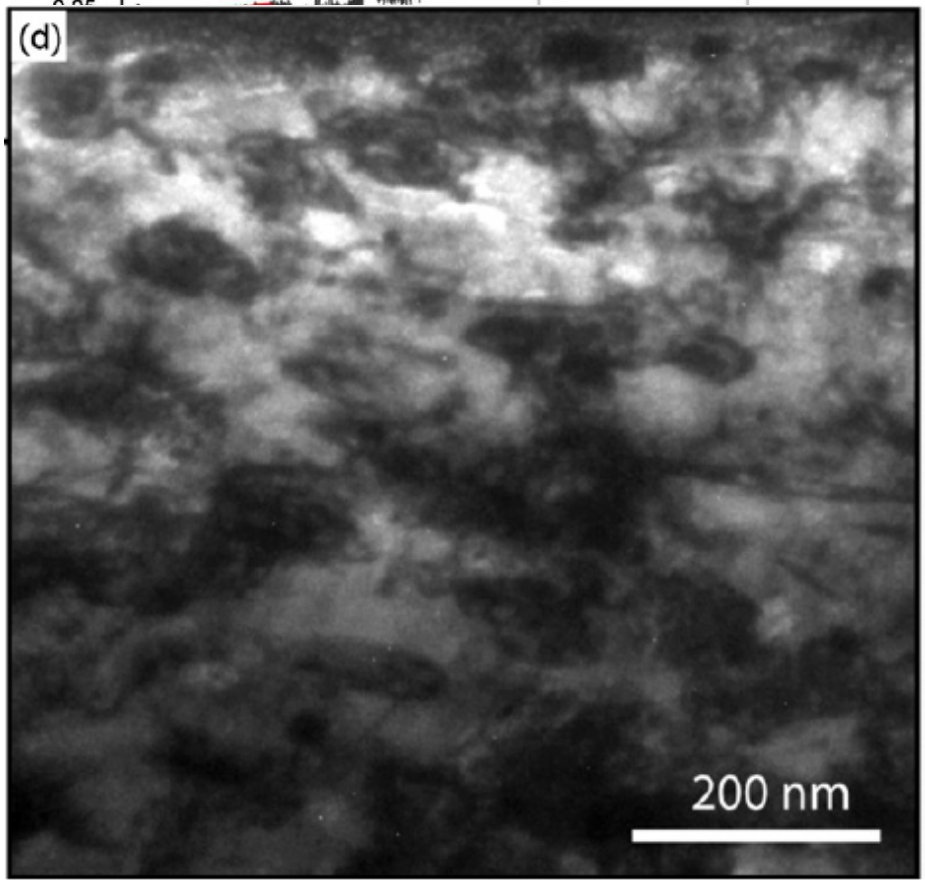
Low friction linked to nanocrystalline surface grain size – even with pure metals

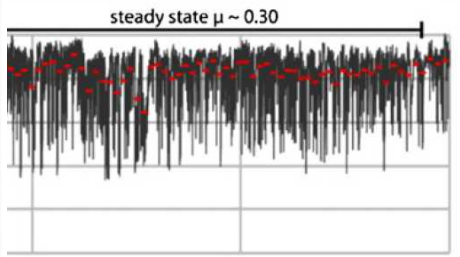
wear rate of ~ 1 nm per kilometer

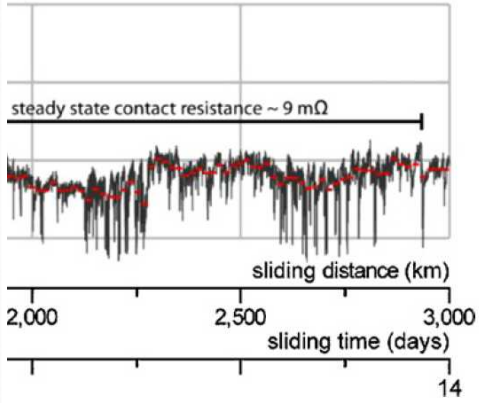
μ_{ss} ~ 0.3







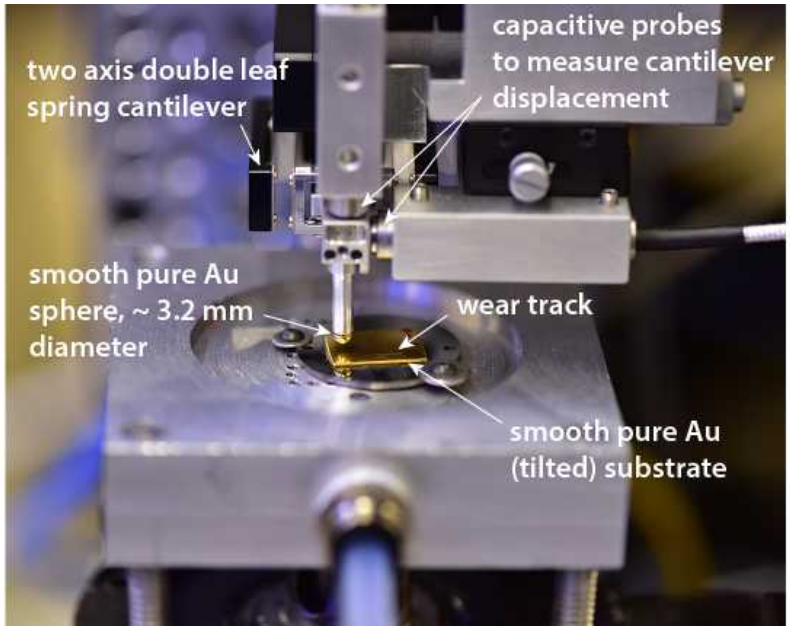




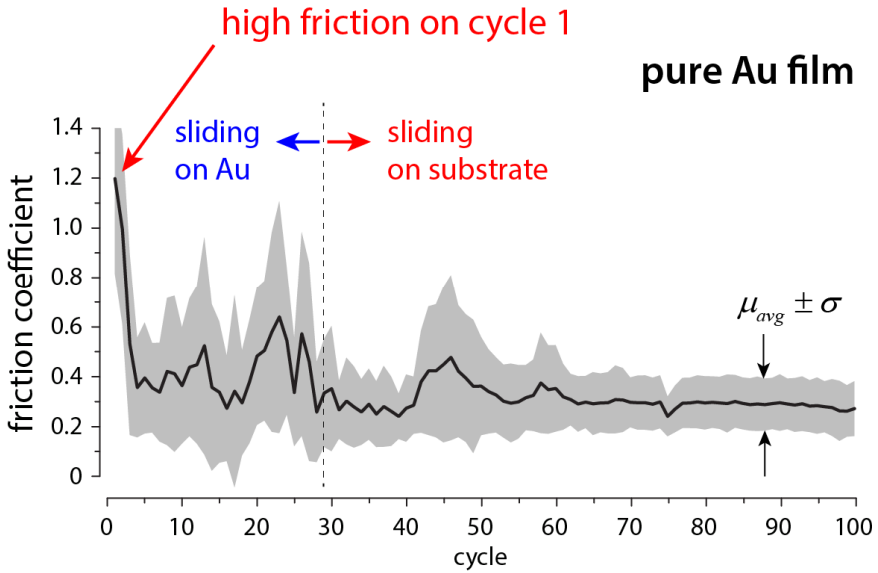
Low friction associated with nanocrystalline surface for a Cu-C system

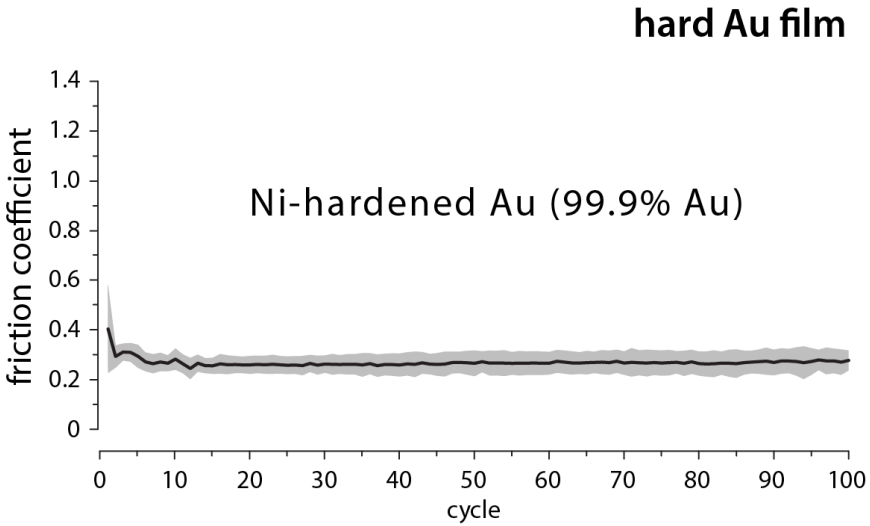
ref: Argibay et al., Wear 2010

Example of low friction with unlubricated metal contacts – Hard Au vs Pure Au



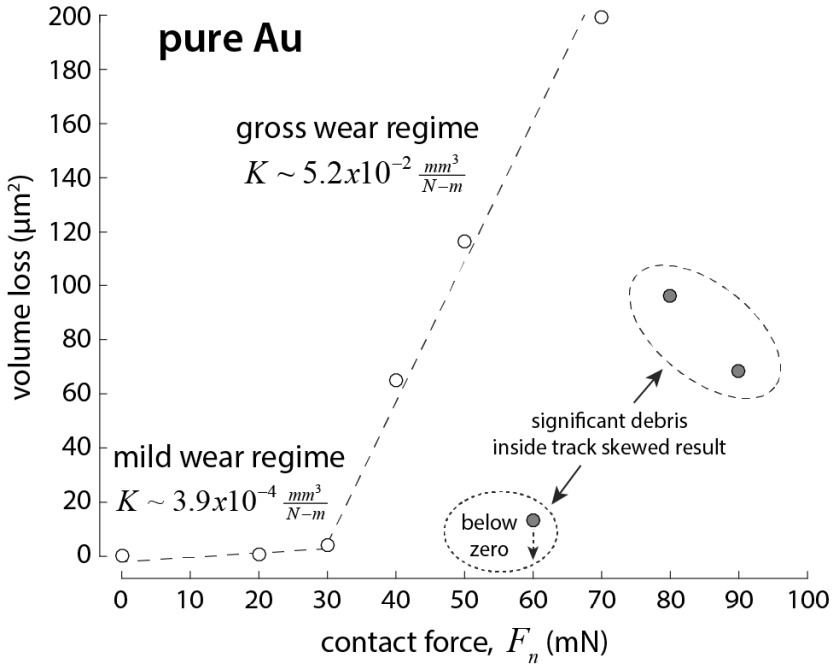
normal force = 100 mN
ball radius = 1.6 mm
speed = 1 mm/s

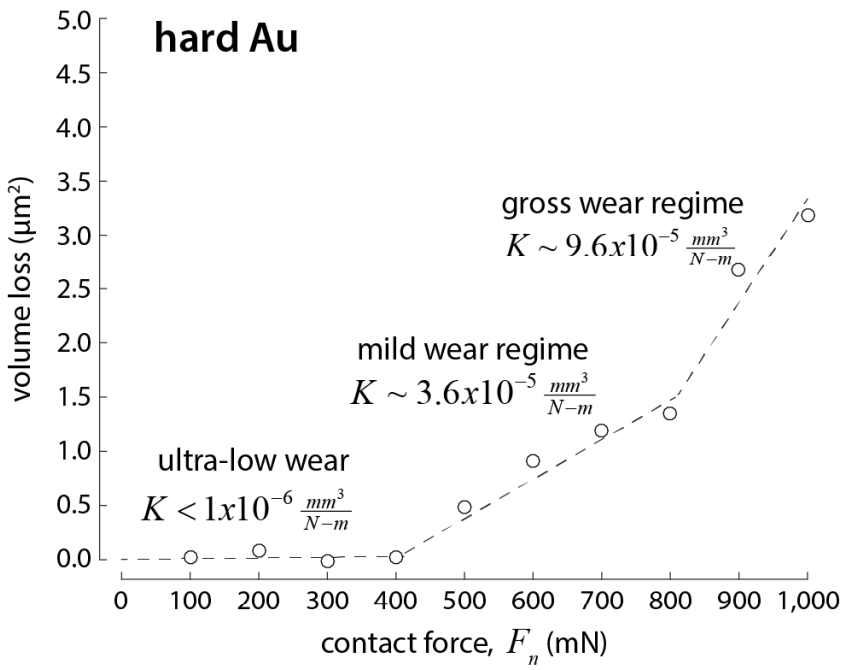




Example of low wear with unlubricated metal contacts – Hard Au vs Pure Au

Note x- and y-axis scale difference!



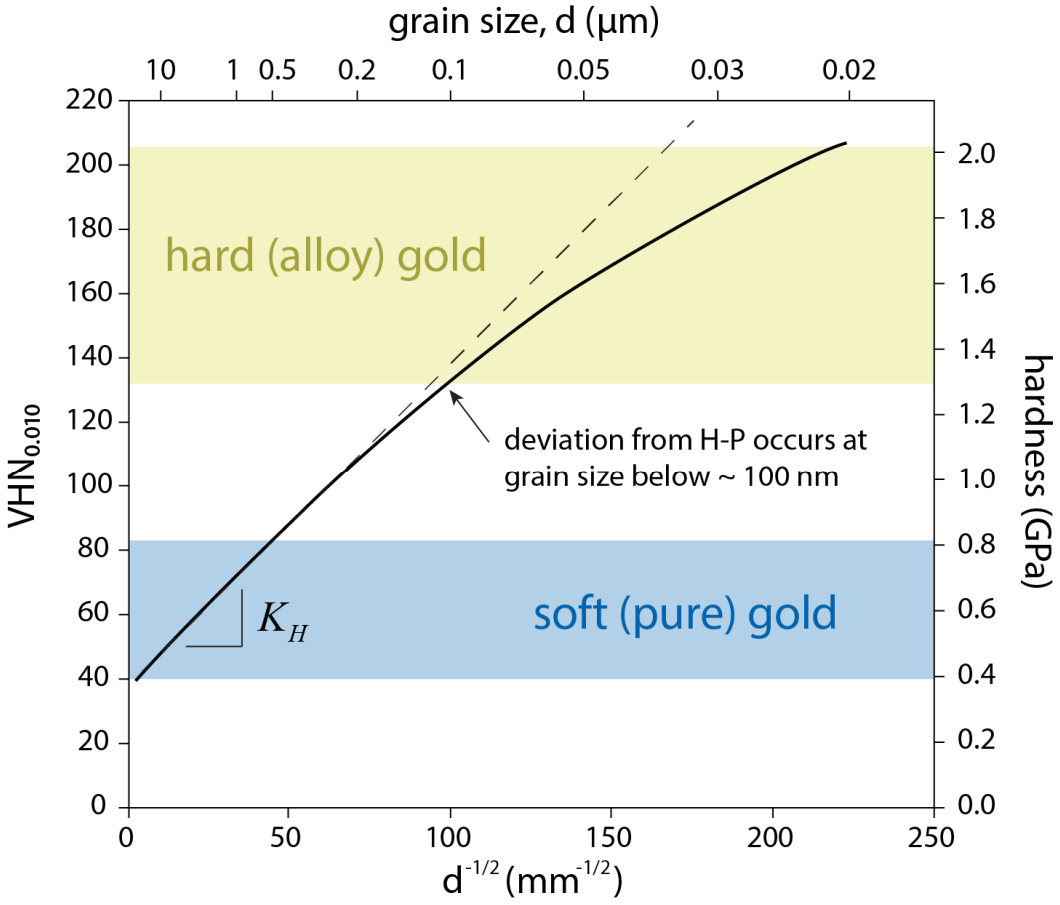


<u>> 10,000x</u> wear rate reduction with addition of 0.1% Ni

Archard wear = rate is linearly dependence to contact force

low wear and friction with pure Au is possible!

Alloying produces stable NC metal in bulk… alloys -> low μ at higher stresses



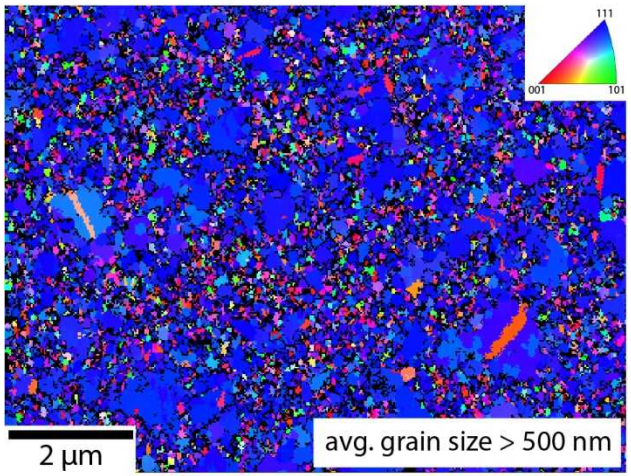
Reference: C. Lo, J. Augis, and M. Pinnel, JAP (1979)

Stable NC grain size can be also achieved by adding non-metal species!

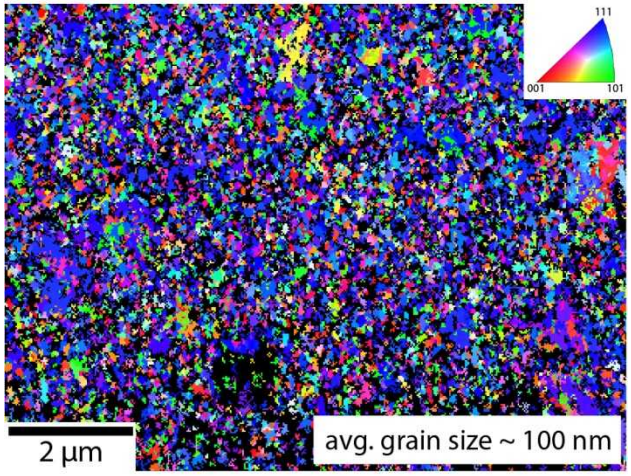
refs: Argibay et al. JAP (2015) and Wear (2013)

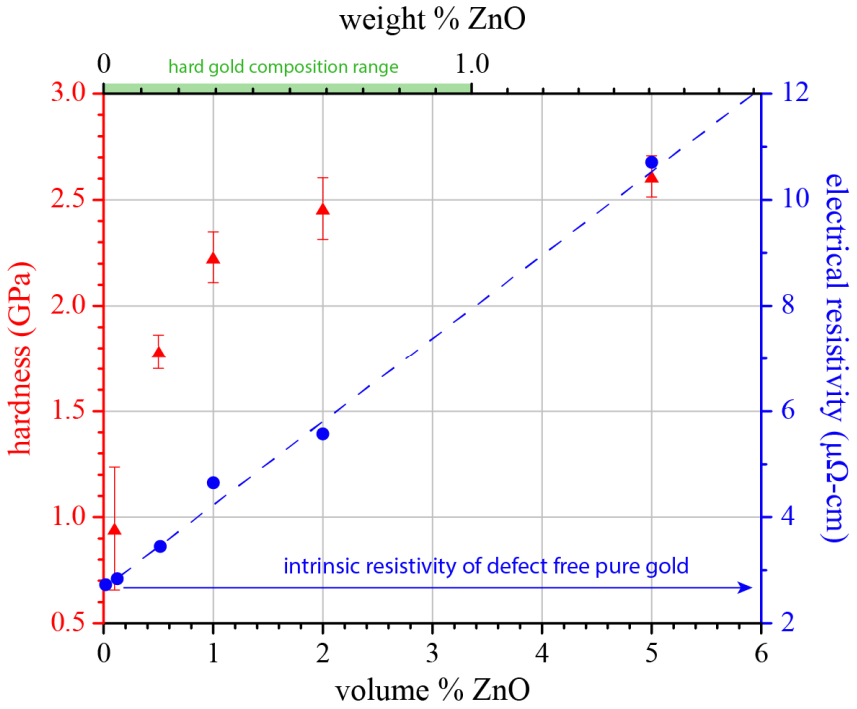
Surface-normal (planar) EBSD maps of e-beam deposited films:

pure Au film

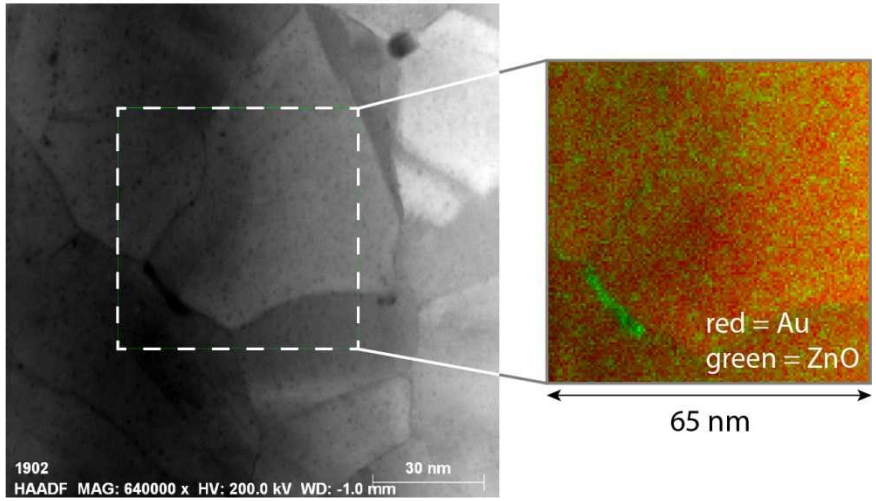


Au film with 0.1 vol. % ZnO

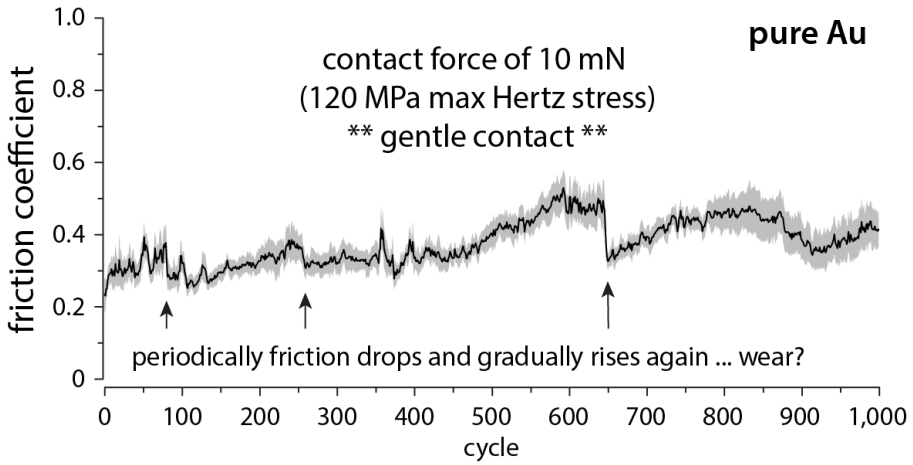




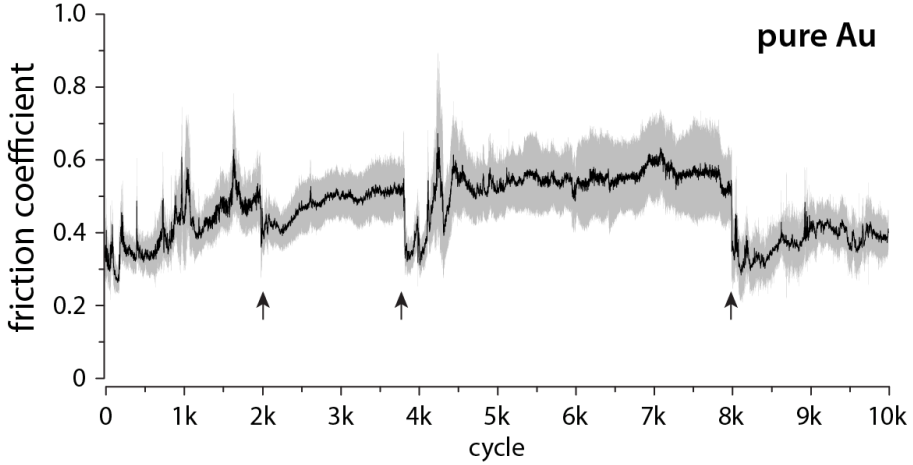
Cross-sectional composition map of a Au- 5 vol. % ZnO film showing grain boundary segregated ZnO in a Au matrix



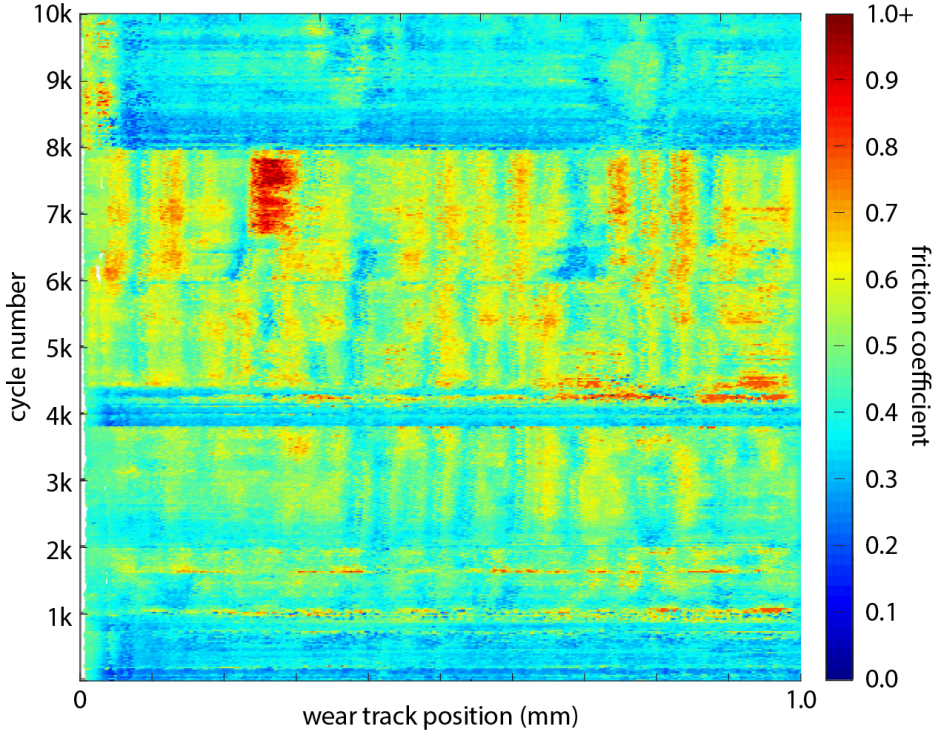
However, low friction IS possible even with pure Au sliding against hard Au



now a longer experiment to 10k cycles:

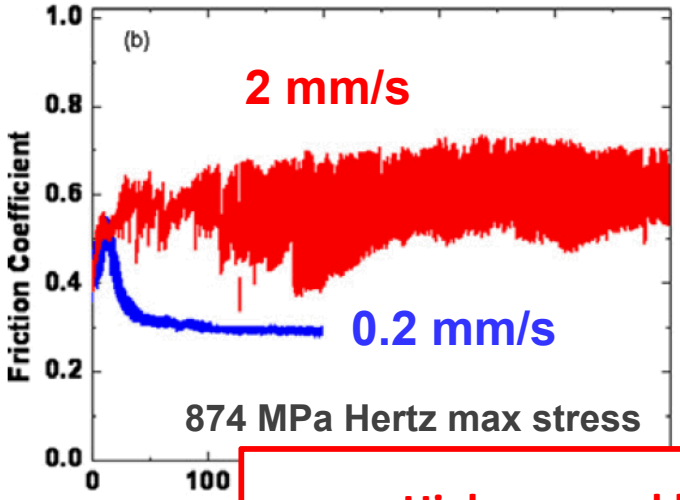


Friction Mapping Reveals More

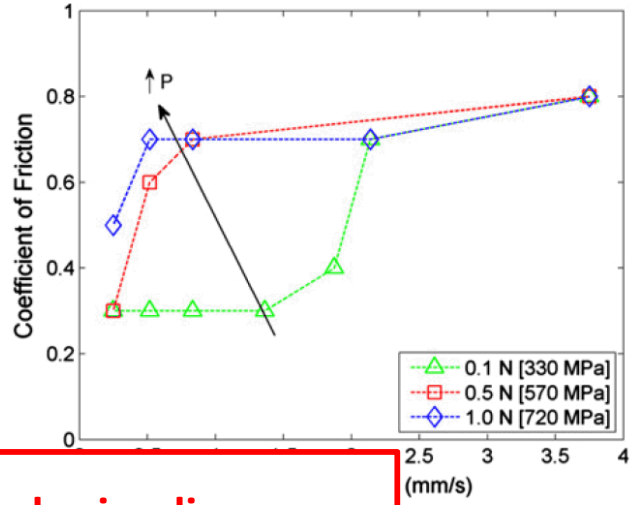


Evidence of threshold stress and speed for low friction to prevail

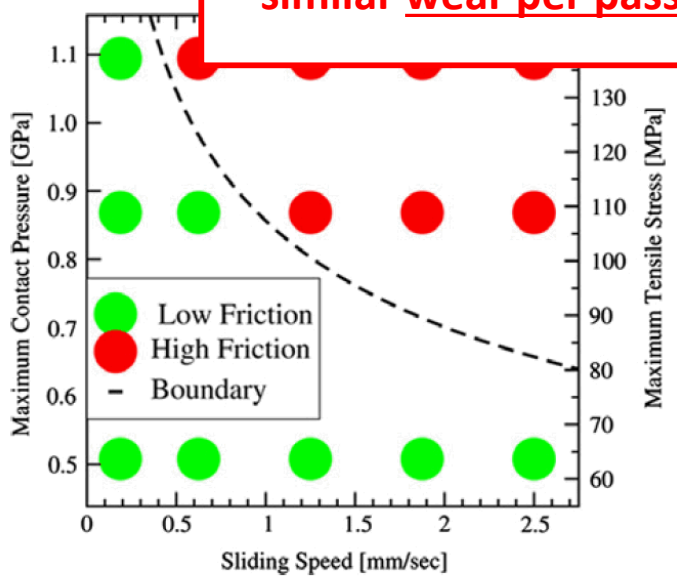
pure Ni data from S. V. Prasad et al., Scripta Mat. (2011)

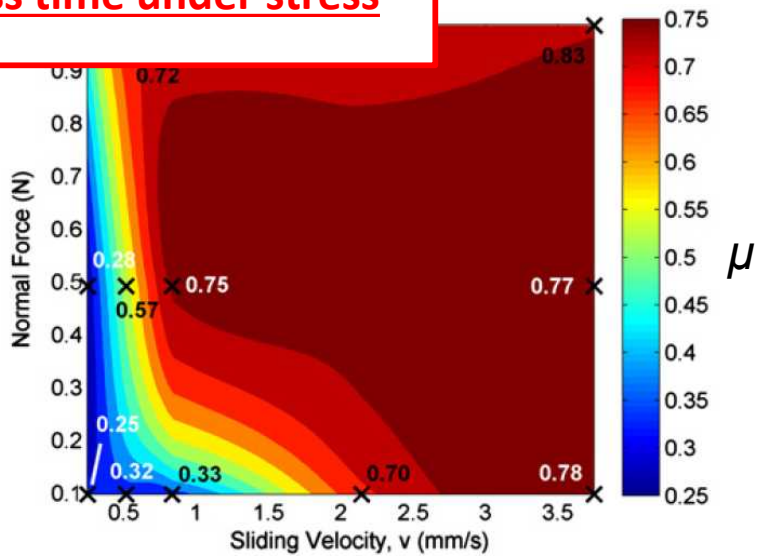


Ni-Fe data from H.A. Padilla et al., Wear (2013)

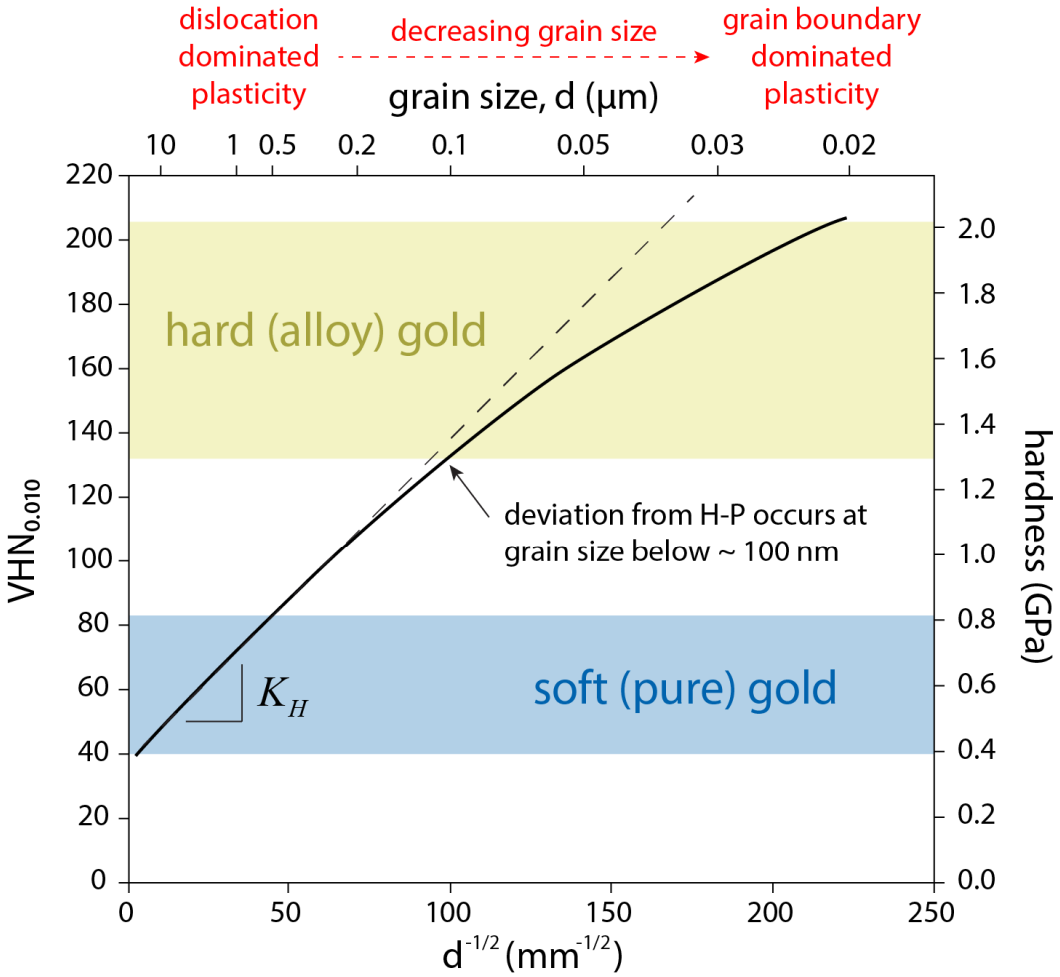


Higher speed but same cycles implies
similar wear per pass but less time under stress





Why "hard" Au? Hardness increase associated primarily with grain size reduction



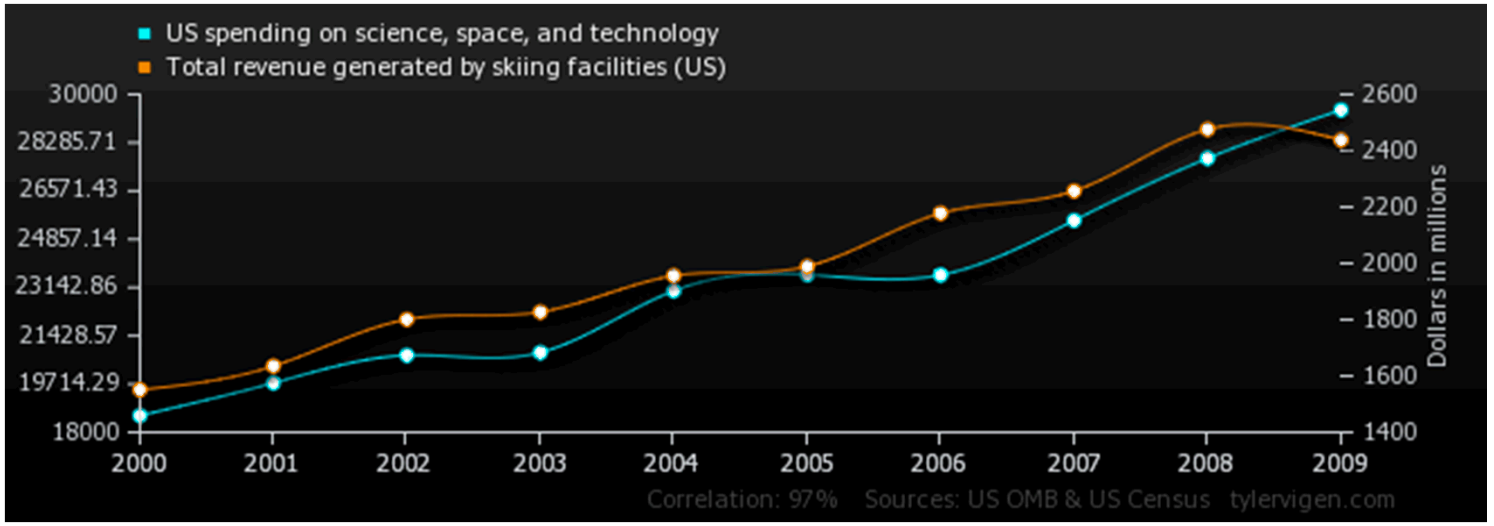
Reference: C. Lo, J. Augis, and M. Pinnel, JAP (1979)

The widely held misconception that <u>hardness increase is the source of low wear and friction</u> is loosely attributed to the notion that real contact area drops with hardness:

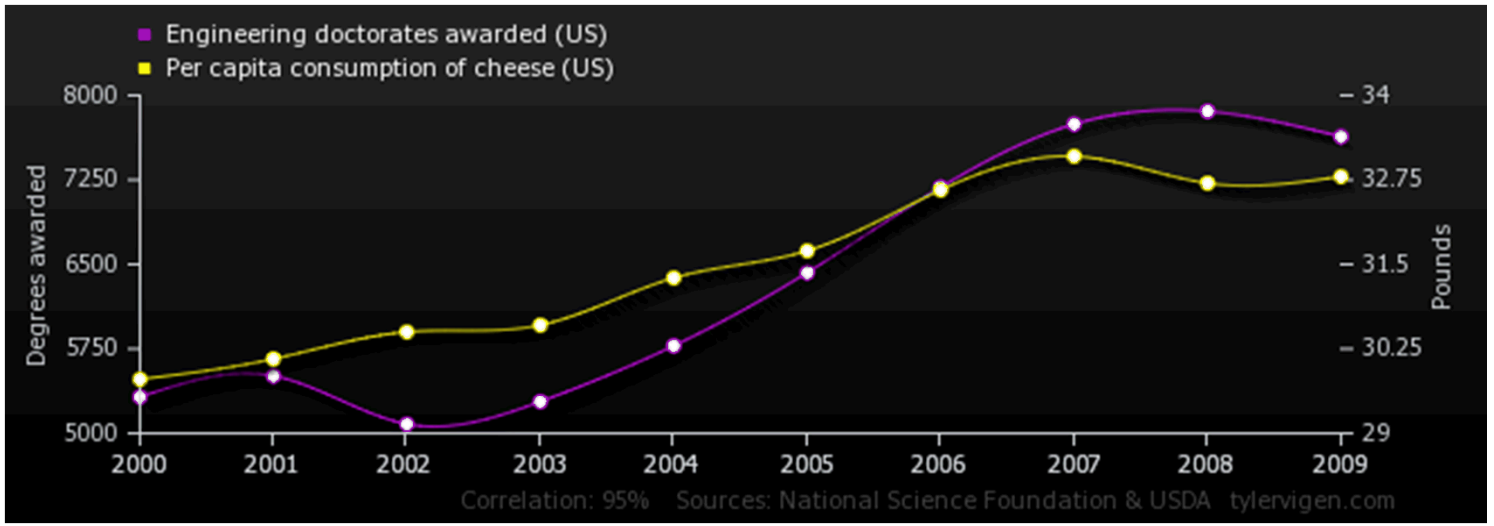
$$A_r \cong \frac{F_n}{H}$$

... for metal contacts the real area is a function of hardness and contact force. (Bowden & Tabor, 1939)

Correlation is not causation…



$R \sim 0.97$



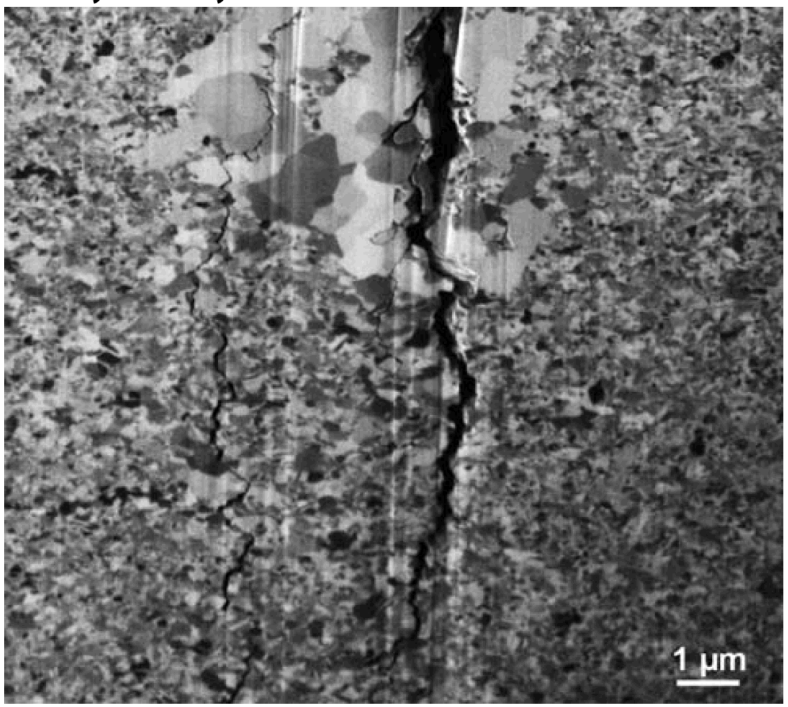
$R \sim 0.95$

hardness is not the answer
(the answer seems to be grain boundary stability and grain size evolution)

Grain size **stability** remains the key challenge to widespread adoption of NC metals

stress-driven grain growth

initially nanocrystalline Ni-Mn

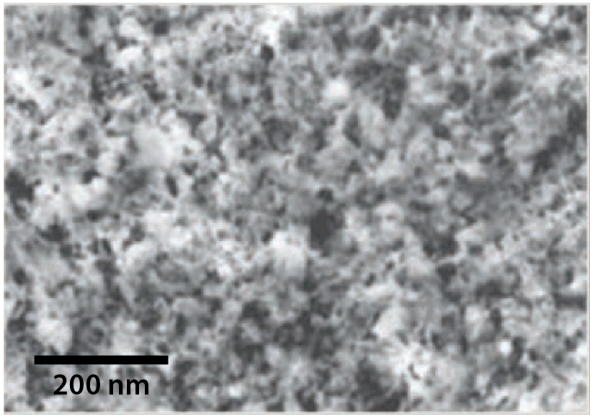


ref: Padilla & Boyce, Exp. Mech. (2010)

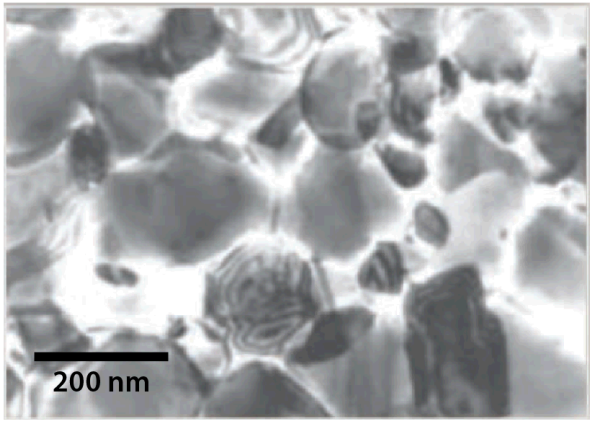
This implies contact stress can drive coarsening...

thermally-driven grain growth

initially nanocrystalline Ni



after anneal at 300°C for 30 minutes



... and contact heating can drive coarsening (Blok, Jaeger, Archard, Lim and Ashby)

Two routes to stabilize nanocrystalline metals – kinetic and thermodynamic

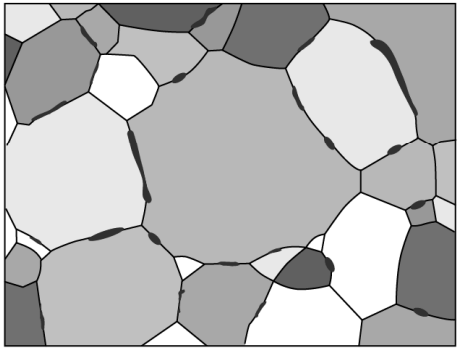
ref: Simoes et al., Nanotech. (2010)

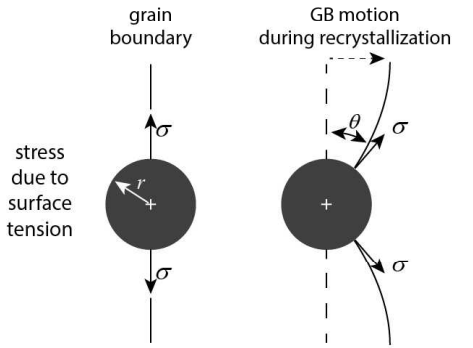
Grain growth is essentialy driven by grain boundary described by speed of grain boundary motion (speed), v

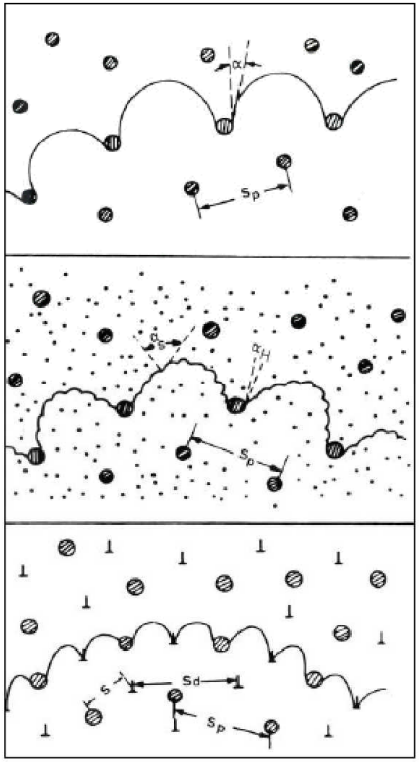
$$v = M \cdot P = M_o \exp\left(-\frac{Q_m}{kT}\right) \cdot \frac{2\gamma_o}{r}$$

Limit the ***kinetics*** of recrystallization (traditional quasi-stability)

e.g. Zener pinning, solute drag, porosity







drag force: $f_D = 2\pi r\sigma \cos\theta \sin\theta$

M = grain boundary mobility
P = pressure on grain boundary
γ_o = interfacial energy per unit area
r = mean grain radius

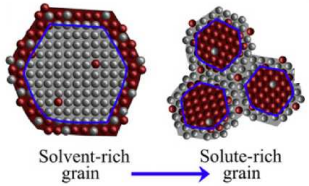
Weissmüller (1993), Kirchheim (2002), and Schuh (2012) have made significant contributions toward understanding and achieving ***thermodynamic*** stability by lowering grain boundary energy through solute segregation

Regular Nanocrystalline Solution (RNS) Model:
ref: Chookajorn et al., Science, 2012

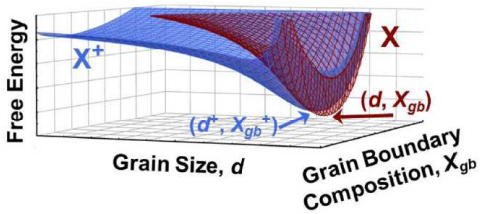
$$\Delta G^{mix} = (1 - f_{gb})\Delta G_c^{mix} + f_{gb}\Delta G_{gb}^{mix} + z\nu f_{gb}(X_{gb} - X_c)\left[(2X_{gb} - 1)\omega_{gb} - \frac{1}{zt}(\Omega^B\gamma^B - \Omega^A\gamma^A)\right]$$

$$dG = \left[\gamma - \frac{N_\beta}{A}\Delta G_{seg}\right]dA$$

Grain structure model:
segregated 2-phase metal system



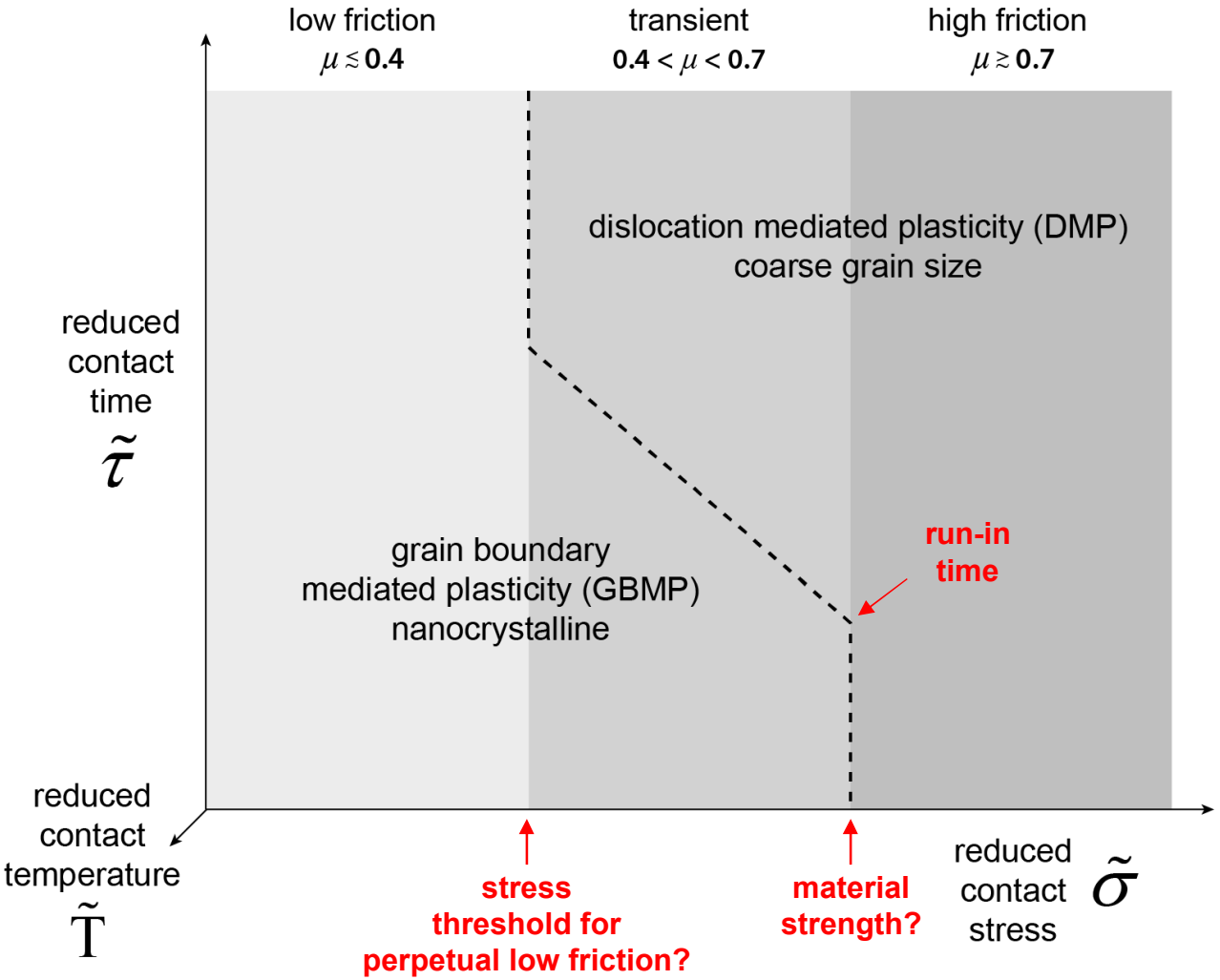
ref: Murdoch et al., Acta Mat. (2013)



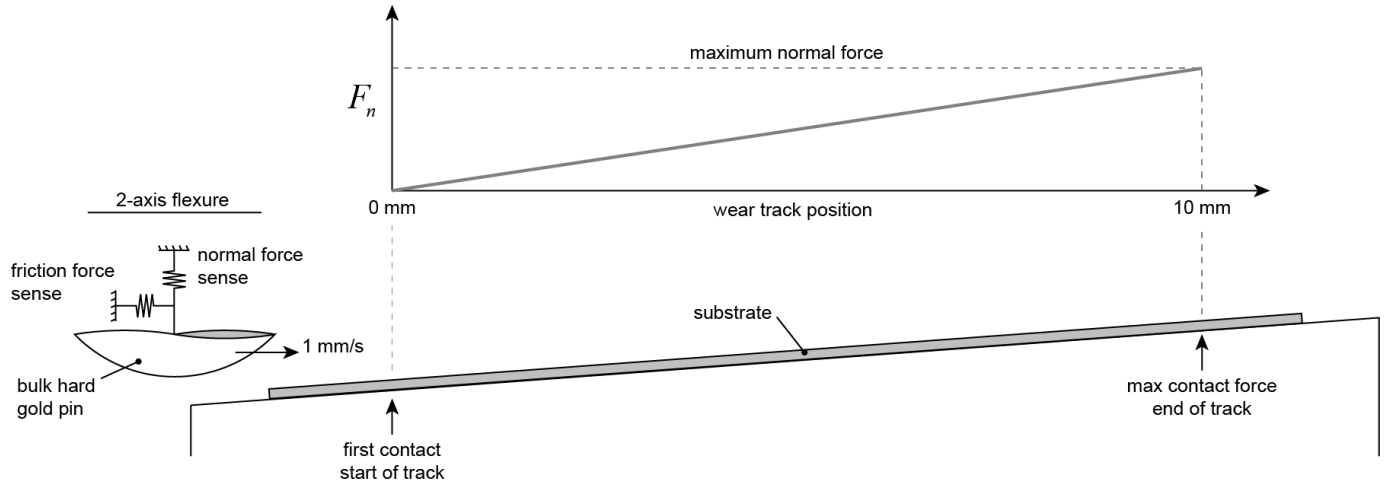
ref: Murdoch et al., Acta Mat. (2013)

What does a microstructure-based friction regimes model look like?

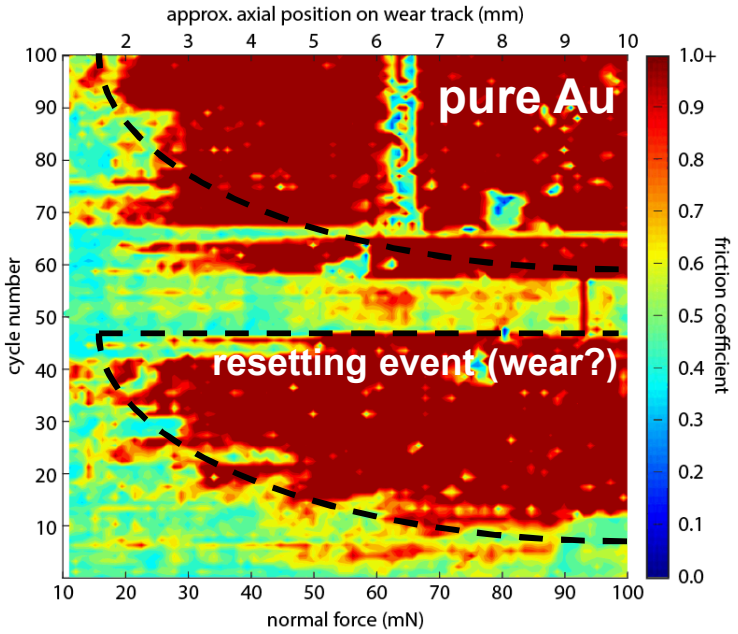
Based on recent experimental and MD simulation results, we arrived at a new model that predicts (some) friction behavior regimes based **exclusively on materials parameters and microstructure evolution models**

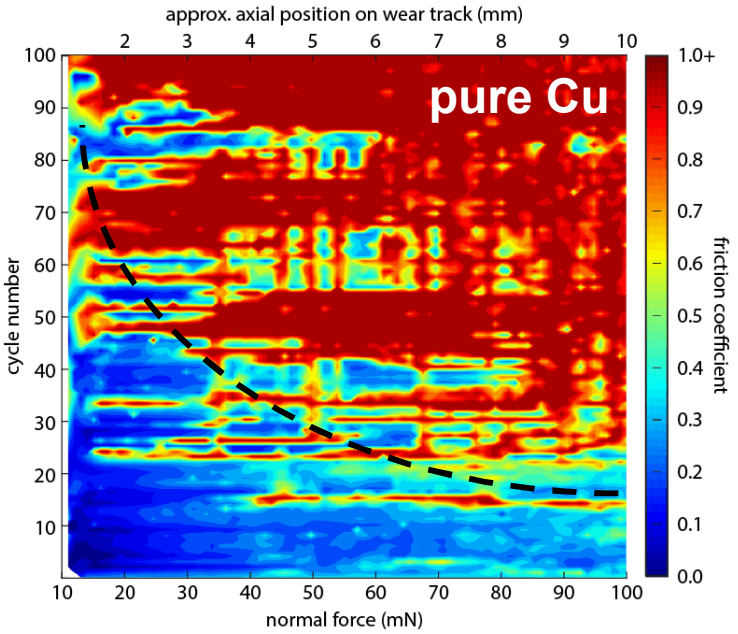


Ramped contact force experiments and friction mapping reveals much



Messy (tribology…), but there is stress dependent envelope!

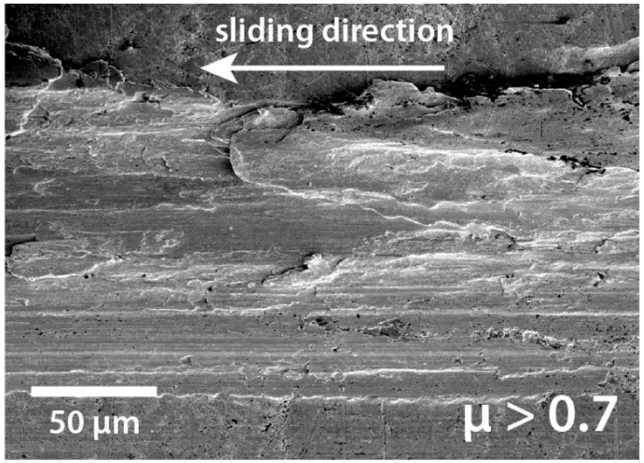


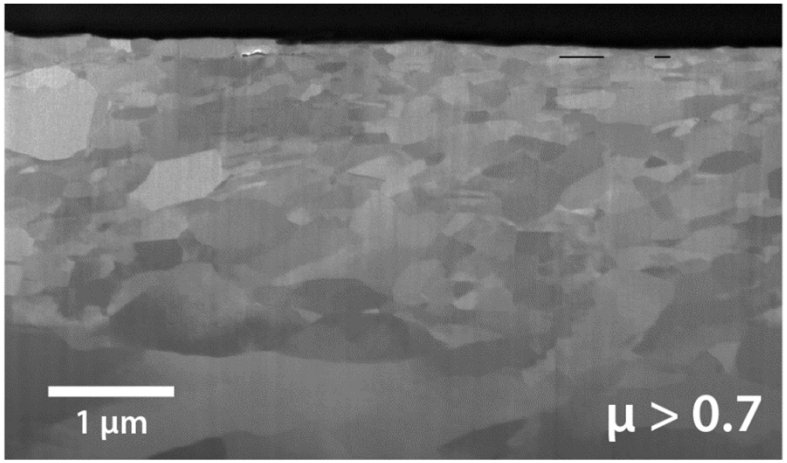


Again, low friction associated with nanocrystalline surface.. see shear banding too

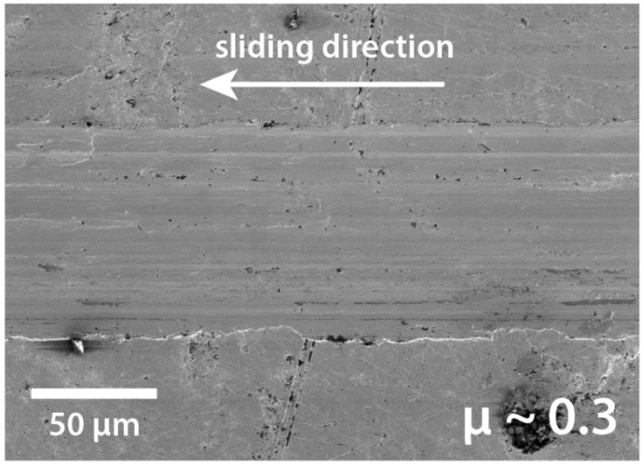
comparing pure Au surfaces and microstructures where low and high friction were measured:

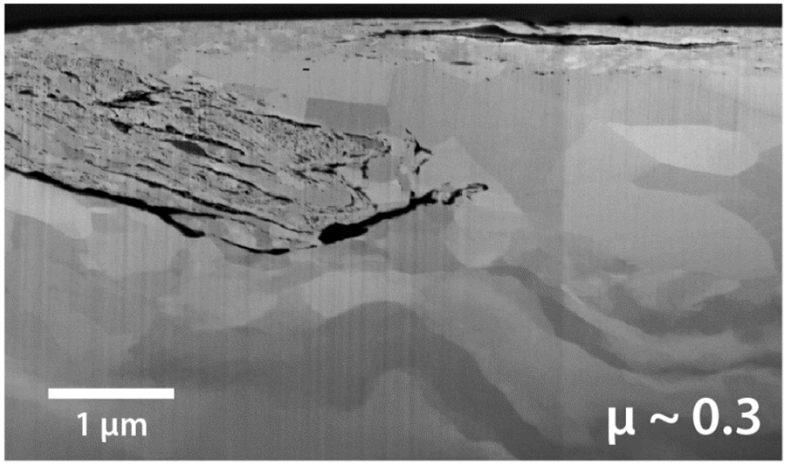
high friction



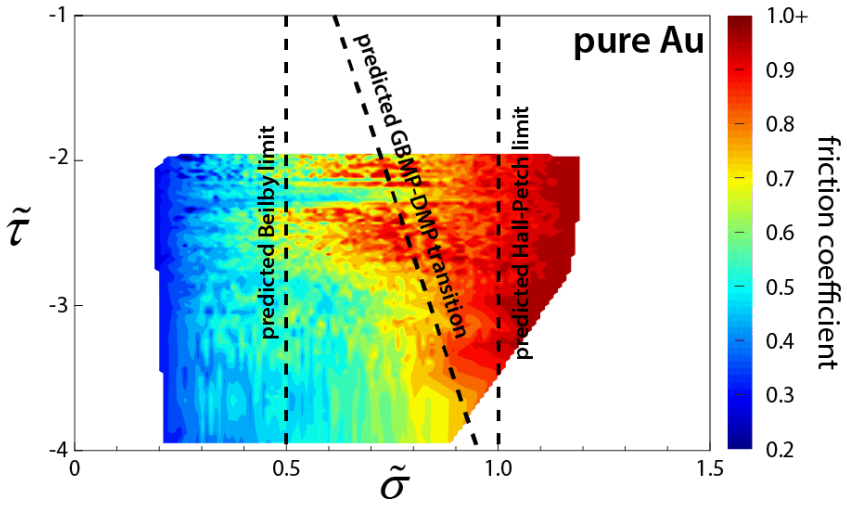


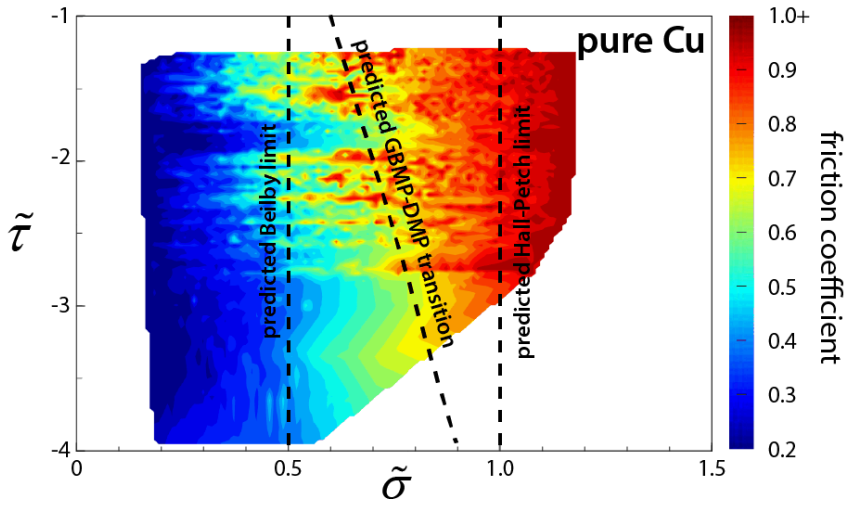
low friction



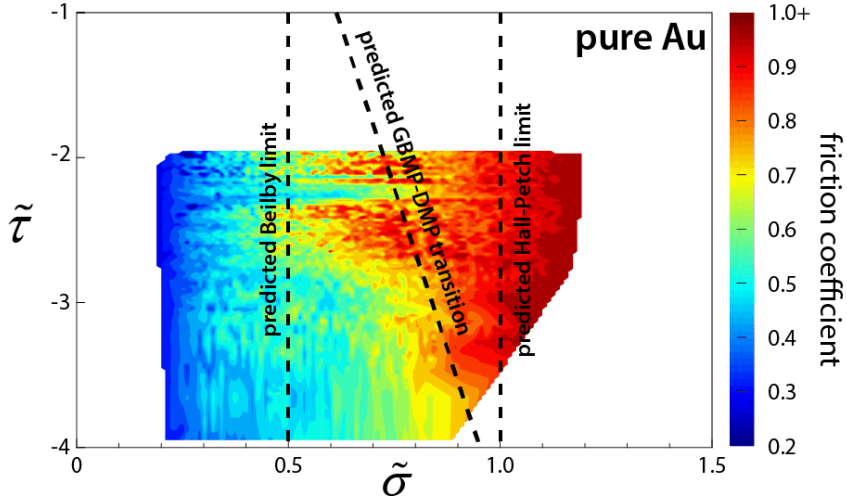


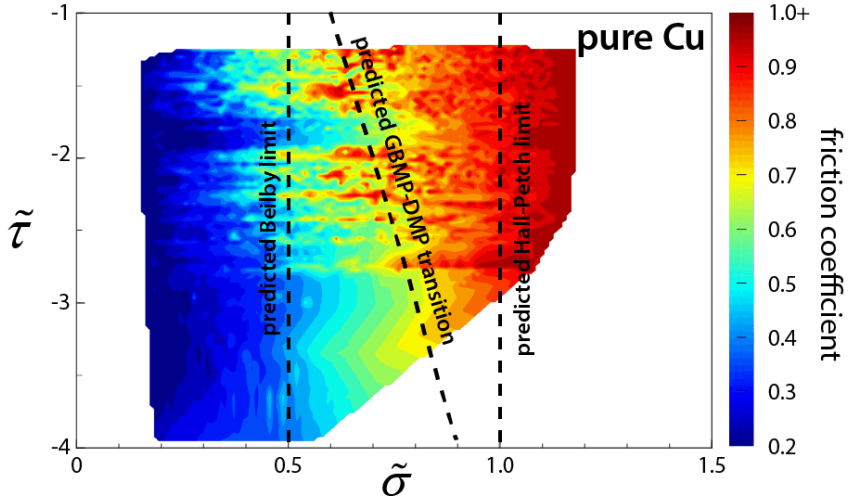
The punchline first: this is what the reduced data looks like





Predictions are pretty good… Note: no fitting or fudging, all bounds predicted!



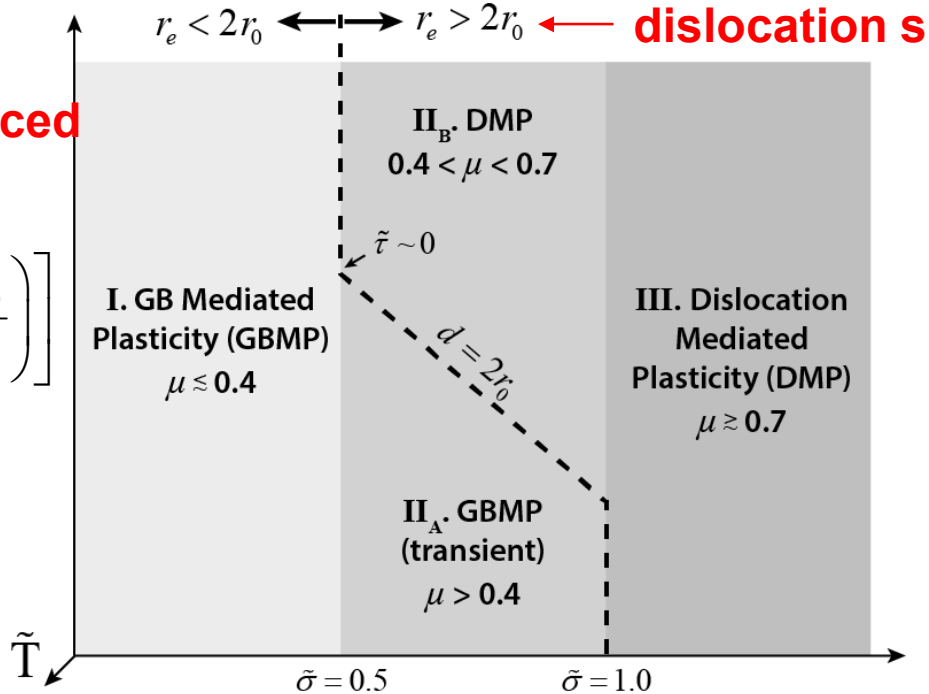


$r_e < 2r_0$ ⟷ $r_e > 2r_0$ ← **dislocation splitting distance is ke**

and formulated reduced time parameter:

$$\tilde{\tau} = \log_{10}\left[\left(\frac{2a_c}{v_s}\right)\left(\frac{4\gamma_{gb}M_0}{r_0^2}\right)\right]$$

contact time / grain growth



Found reduced stress parameter:

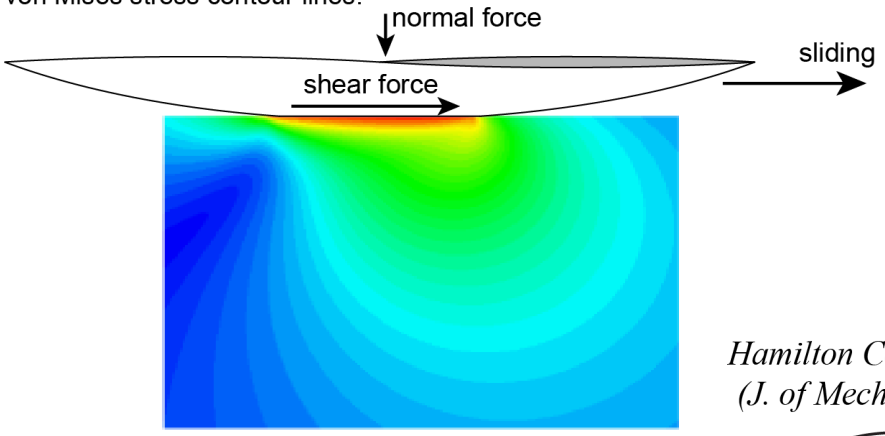
Hamilton Contact Model

$$\tilde{\sigma} = \frac{\sigma_{surf,\max}}{\sigma_\infty}$$

theor. strength

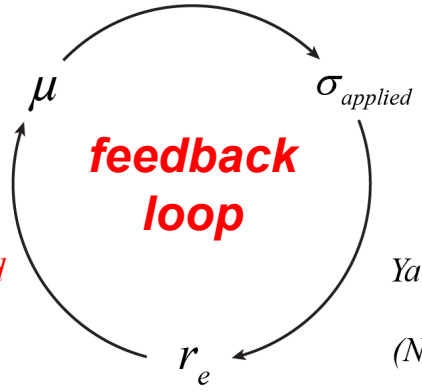
The connection between friction and microstructure evolution

von Mises stress contour lines:



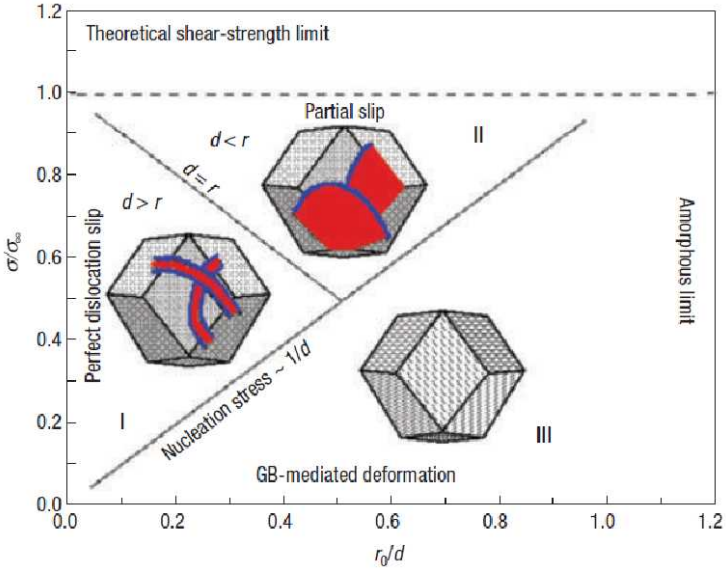
$$\sigma_{a,\max} = \frac{3F_n}{2\pi a^2}\left[\frac{1-2\nu}{3} + \frac{(4+\nu)}{8}\pi\mu\right]$$

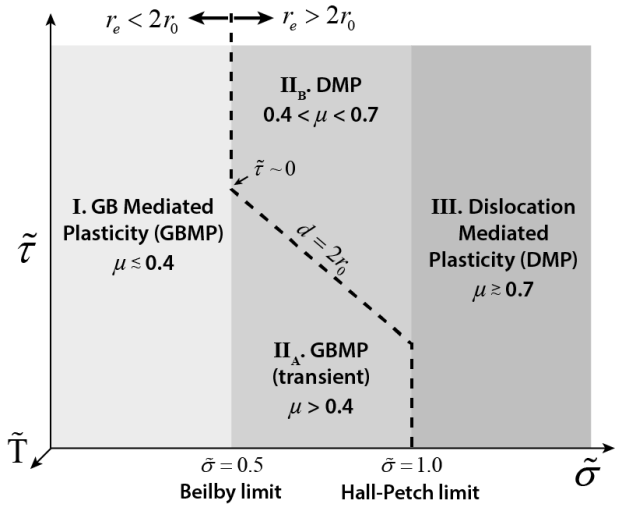
Hamilton Contact Model (J. of Mech. Eng. 1983)



proposed model

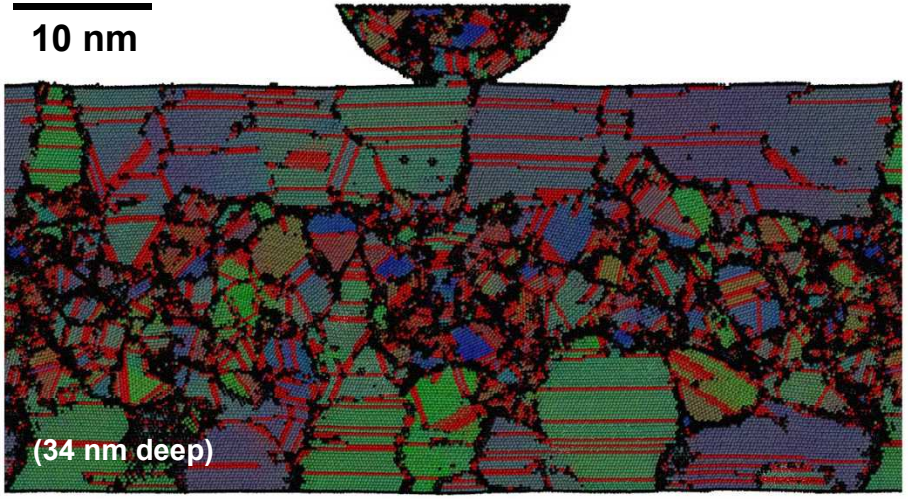
Yamakov et al. model (Nature 2004)





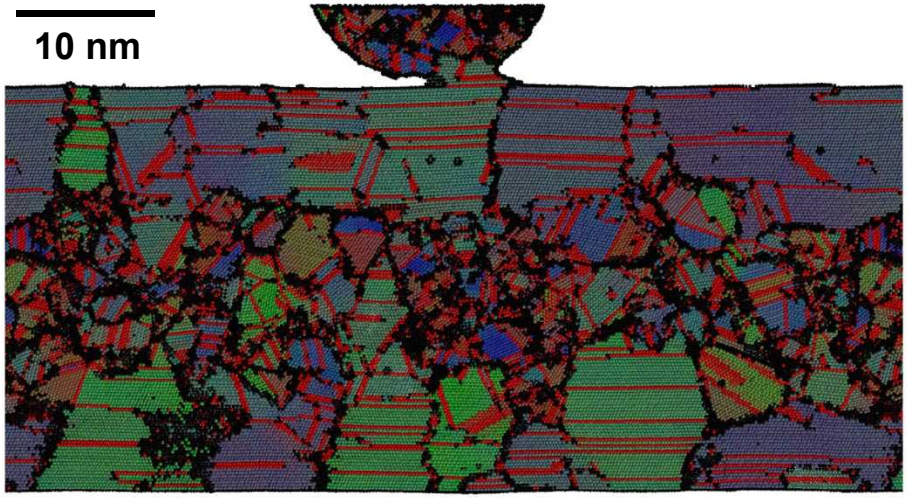
What did we learn from MD simulations…

initial microstructure
of Ag and Ag-Cu alloy
(no sliding yet)

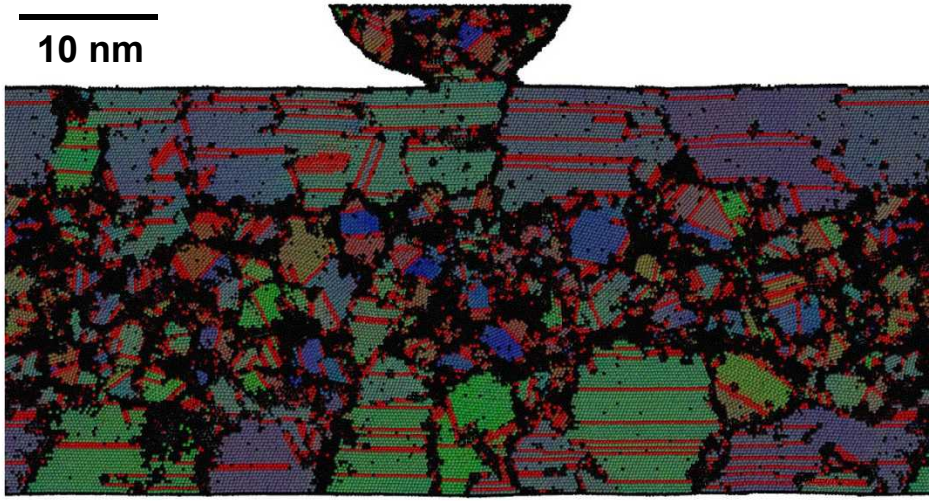


300 MPa contact stress
300 K temperature
2 m/s sliding speed

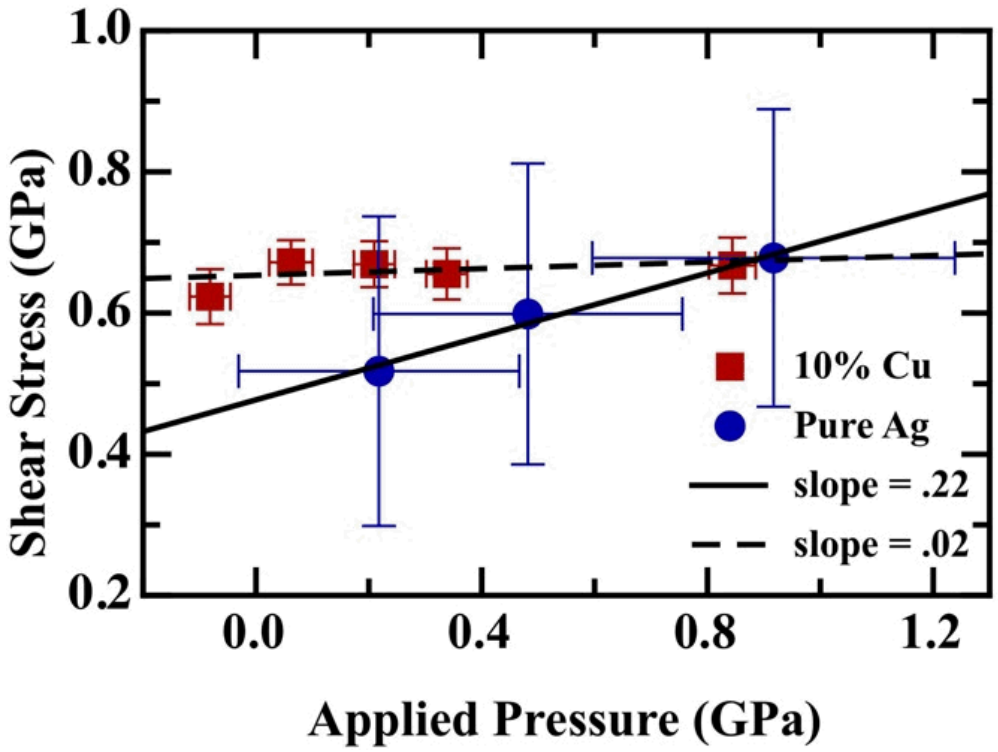
pure Ag after 4 nm of sliding



Ag-10% Cu alloy after 4 nm of sliding

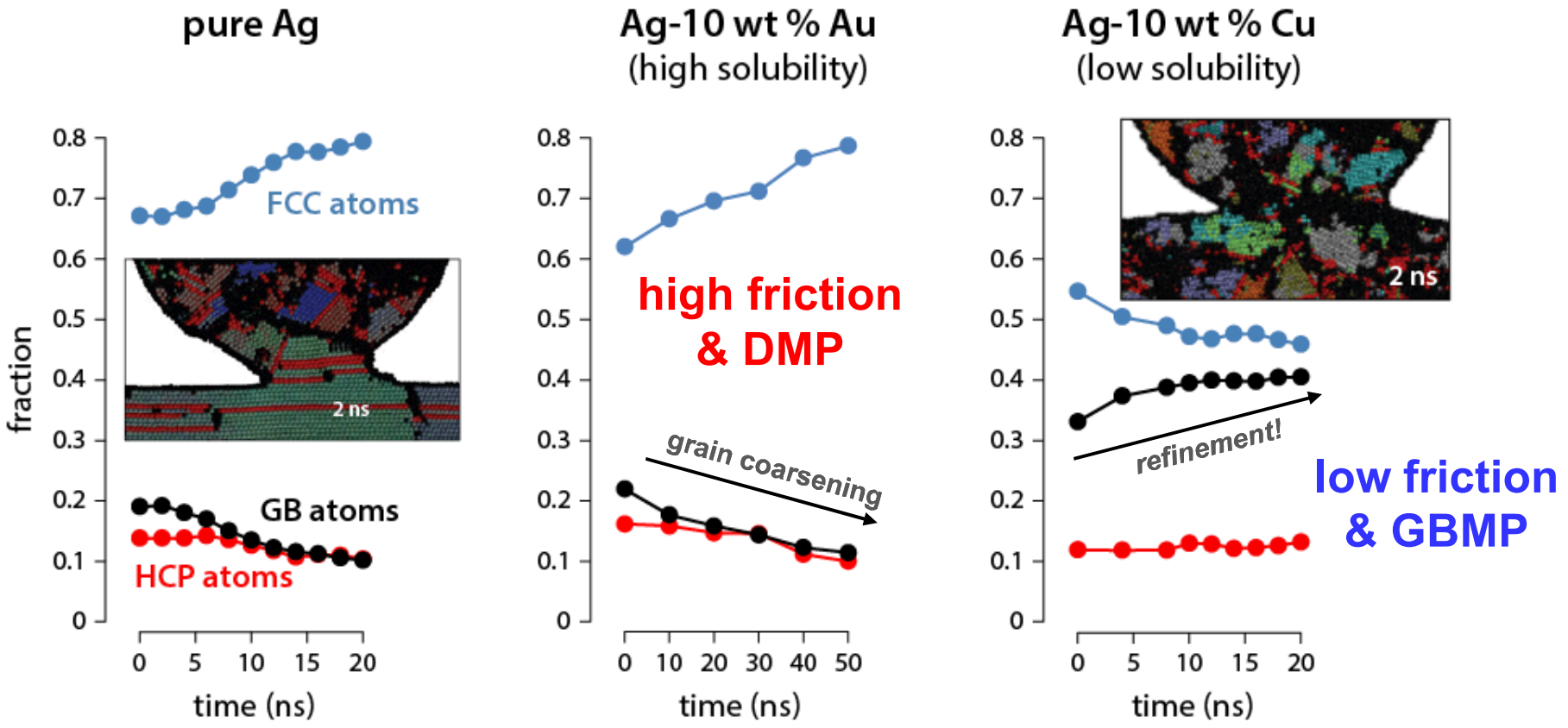


MD shows us that alloying lowers friction



- AgCu is similar to hard gold (AuNi, AuCo…)
- Friction coefficient is the slope of line
- Change in shear accomodation changes the friction

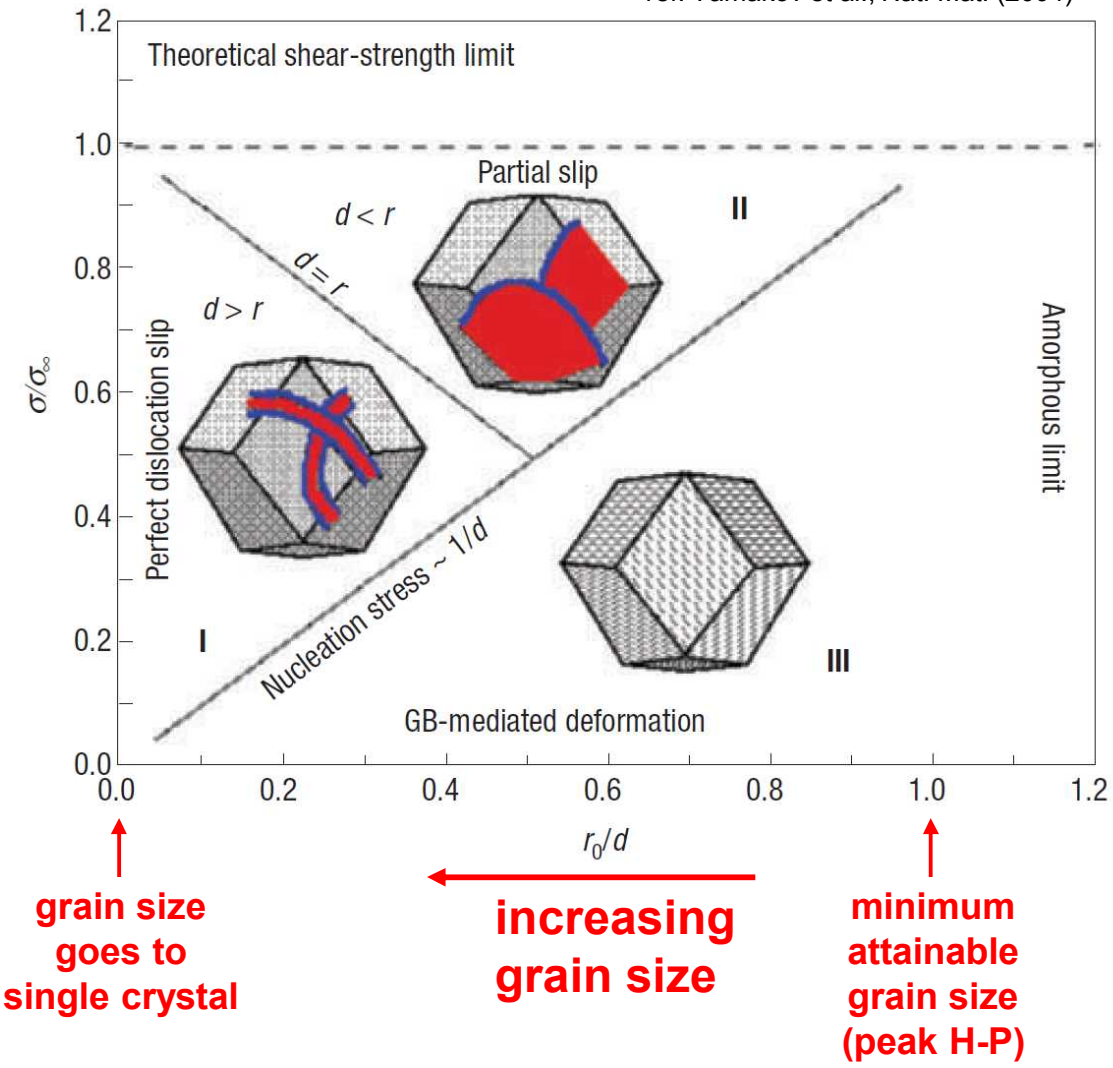
Alloying also impacts microstructural evolution



- Experiments: alloying reduces grain size and stabilizes grain boundaries
- Simulations: alloying mitigates stress-driven grain growth at interface and promotes defect (primarily GB) mediated plasticity
- Connection: stabilizing grain boundaries reduces friction

Evolution of the Yamakov et al. (Nat. Mat. 2004) deformation mechanisms model

ref: Yamakov et al., Nat. Mat. (2004)



grain size goes to single crystal

increasing grain size

minimum attainable grain size (peak H-P)

Equilibrium (zero stress) dislocation splitting distance:

$$r_0 = \frac{(2+\nu)Gb^2}{4\pi(1-\nu)\gamma_{sf}}$$

Stress-dependent splitting distance:

$$r_e = \frac{r_0}{1-\sigma_a/\sigma_\infty}$$

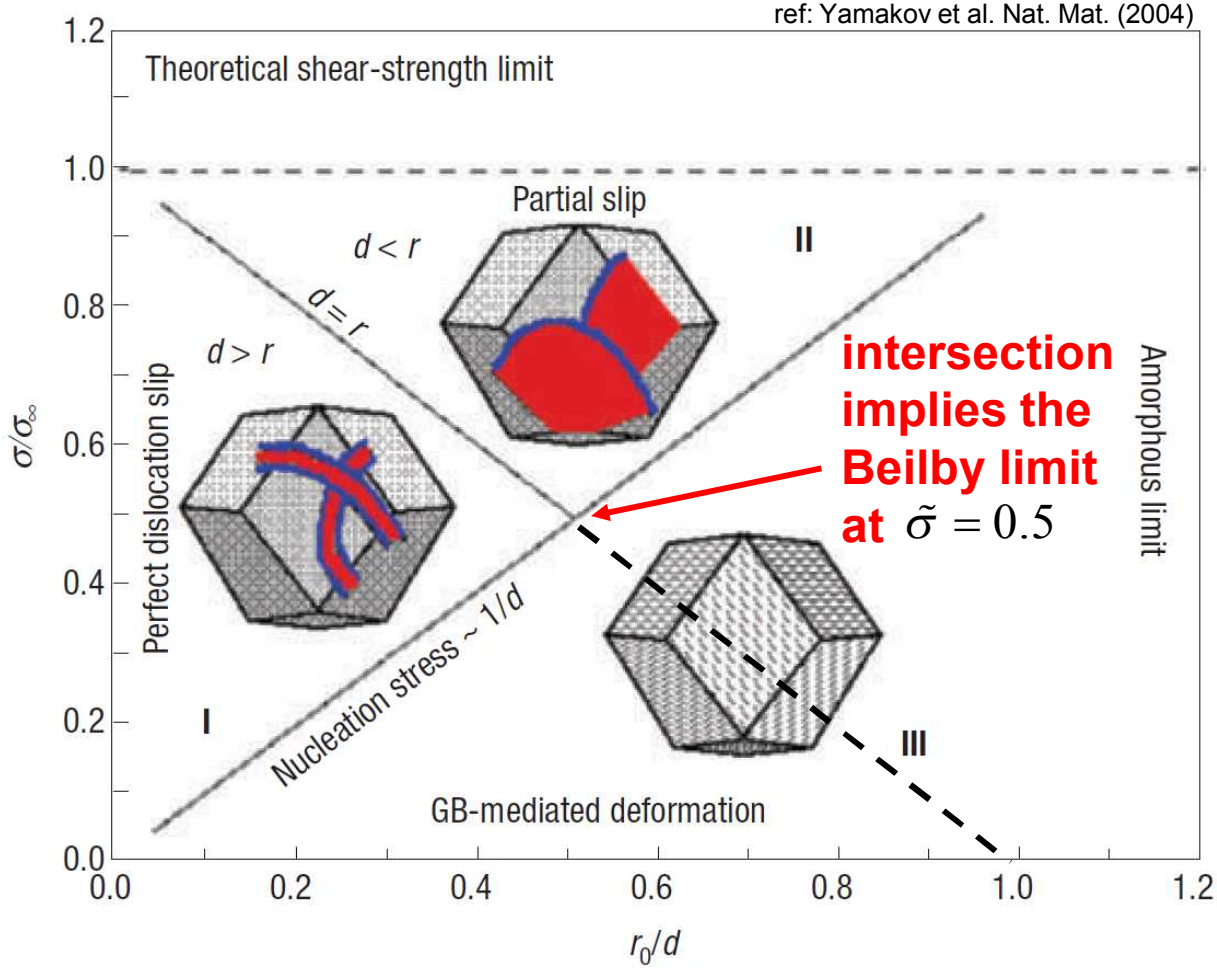
Theoretical strength, grain size where Hall-Petch reaches max:

$$\sigma_\infty = \frac{2\gamma_{sf}}{b}$$

Ref: Froseth et al., Acta Mat. (2004)

Evolution of the Yamakov et al. (Nat. Mat. 2004) deformation mechanisms model

ref: Yamakov et al. Nat. Mat. (2004)



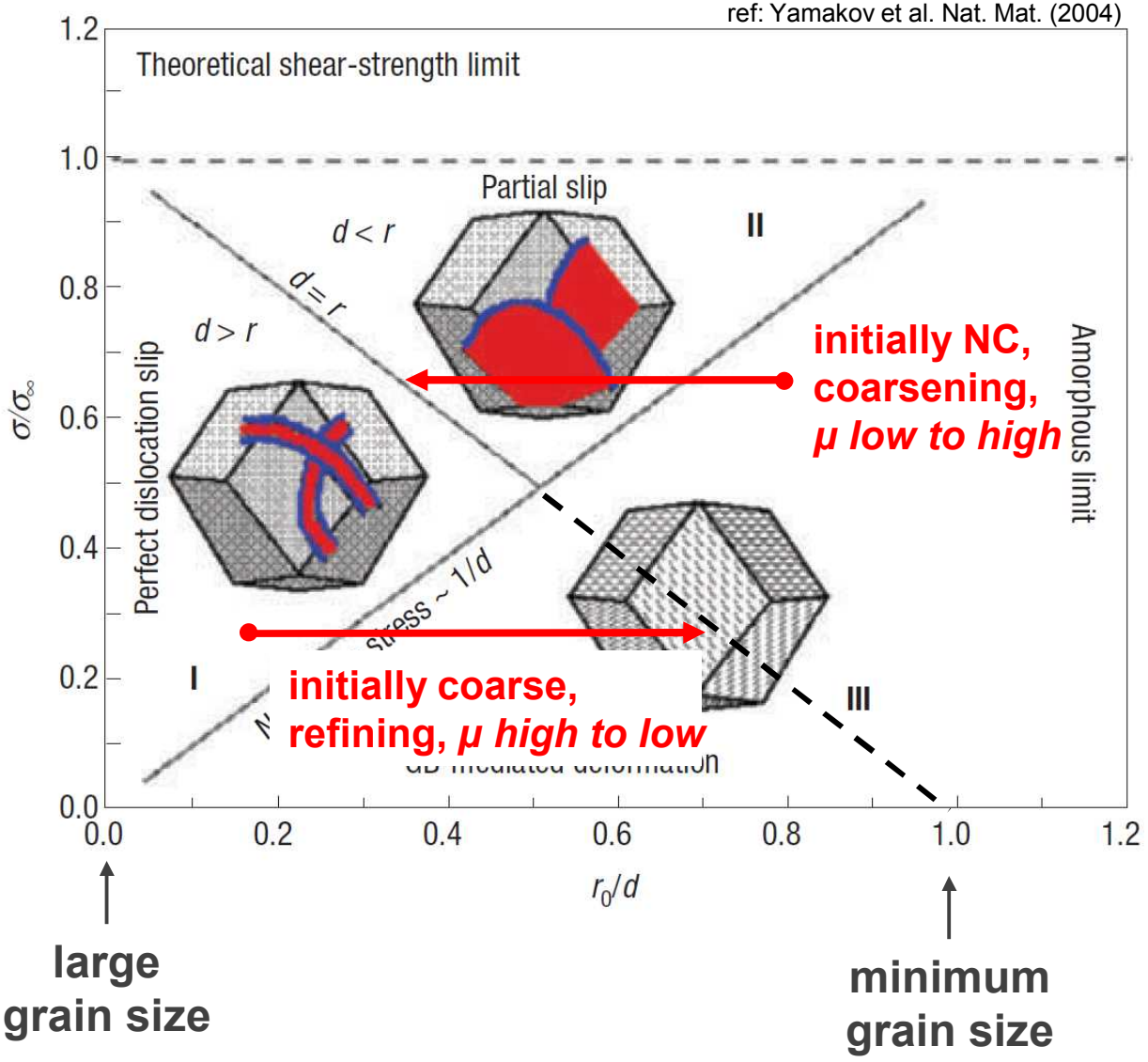
Assumptions:

1. grain size goes to splitting distance, $\rightarrow d$

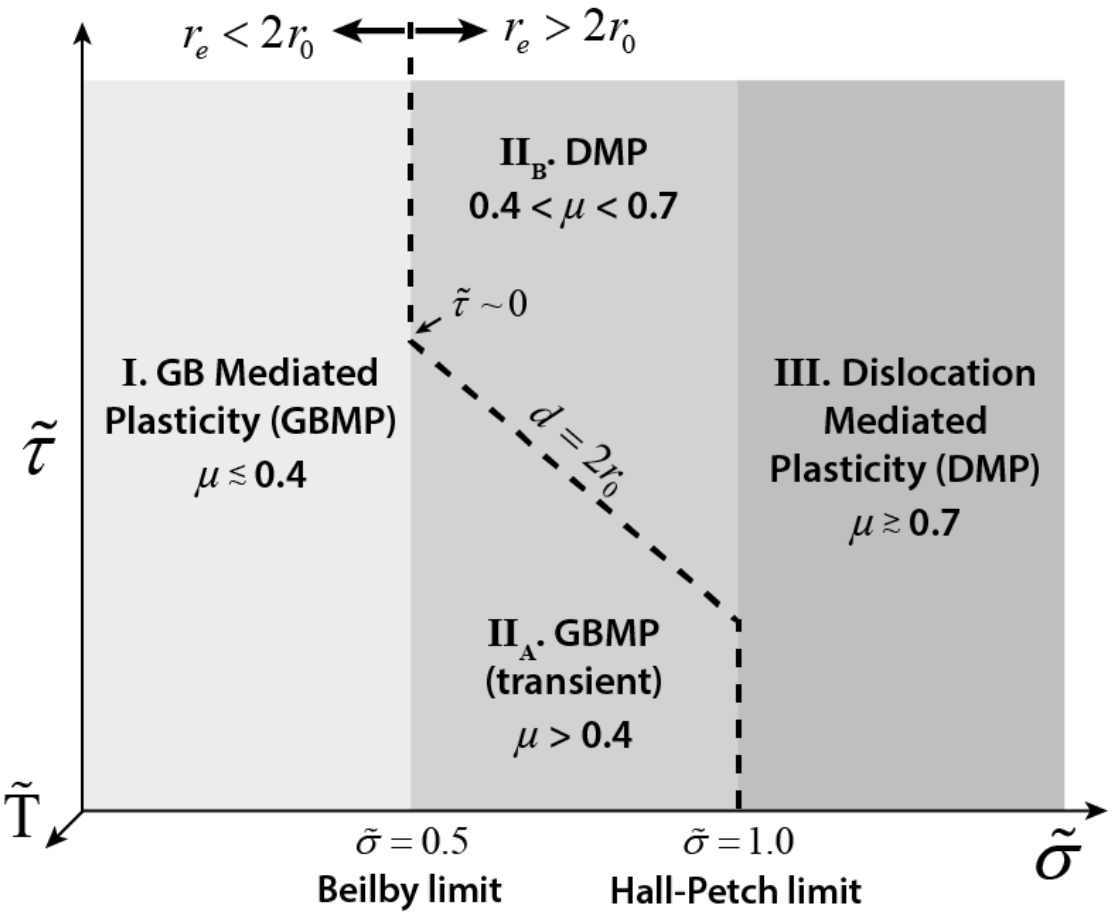
2. nucleation stress goes as inverse grain size, $\tilde{\sigma} \propto \frac{1}{d}$

But this says nothing about evolution… coarse grain surface can be driven to NC

ref: Yamakov et al. Nat. Mat. (2004)

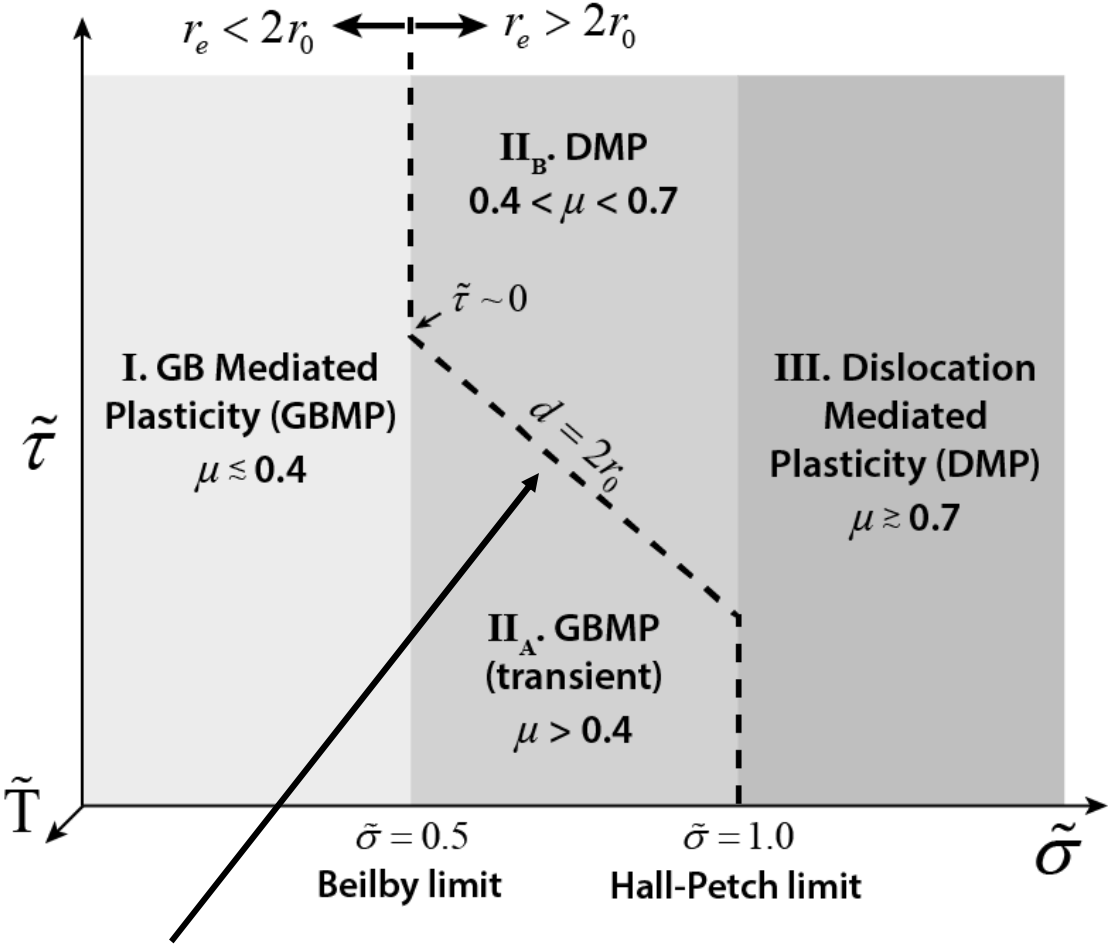


Returning to the microstructure-based friction regimes model…



We assume that wear events reset the surface relatively fast, where in the right conditions even coarse grained material is first **rapidly refined** then **gradually coarsened** via cyclic stress

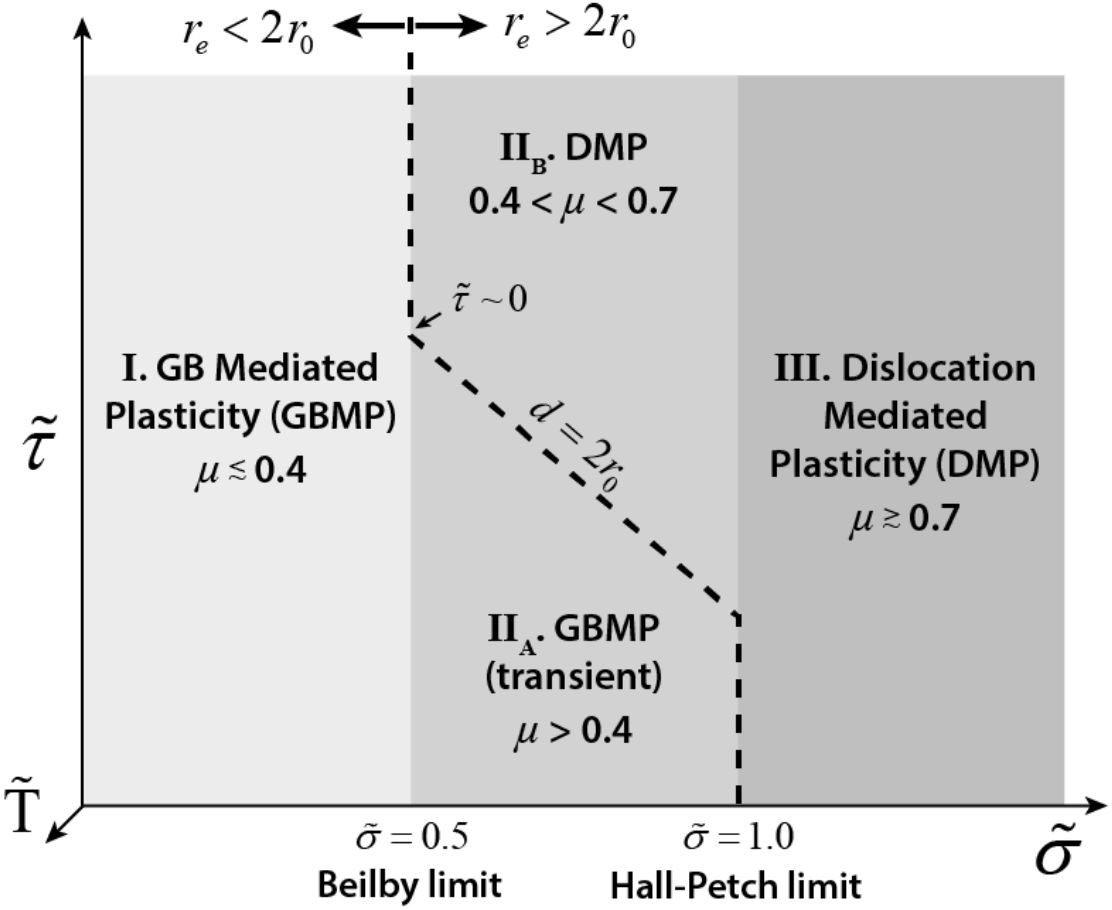
Returning to the microstructure-based friction regimes model…



Then contact time until high friction is governed by stress- and temperature-dependent grain boundary speed:

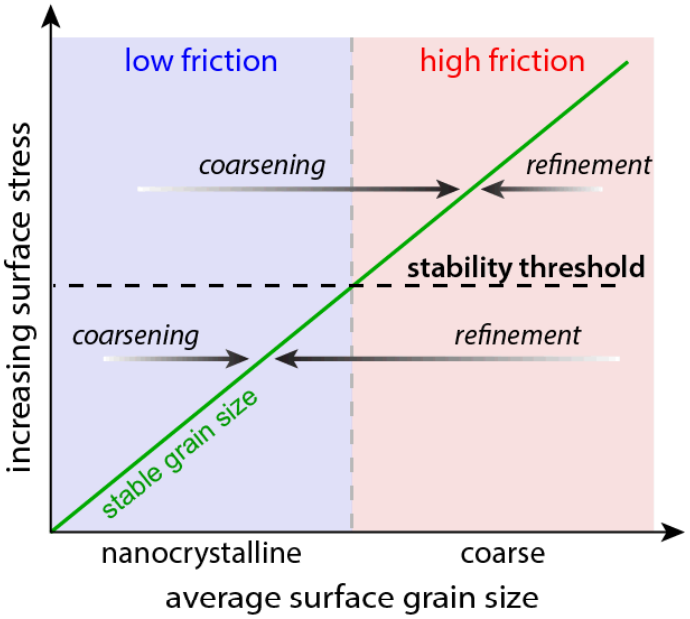
$$v_{gb} = \frac{4\gamma_{gb}}{d} M_0 \exp\left[\frac{-Q}{kT}\right] \exp\left[\frac{\left(\sigma - \frac{1}{2}\sigma_\infty\right) V^*(d)}{kT}\right]$$

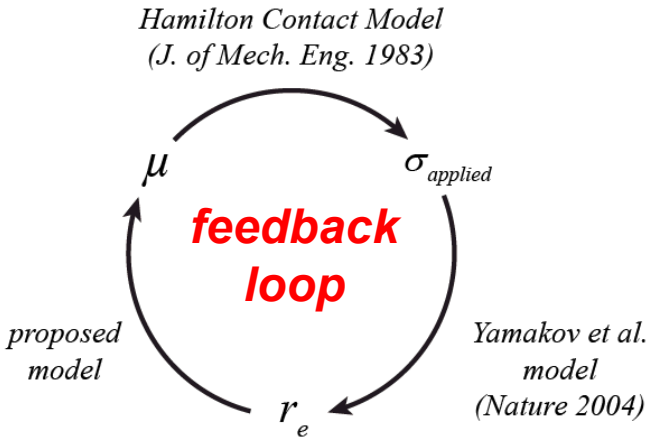
What about boundary lubrication of metal contacts (e.g. graphite, DLC, MoS_2)?

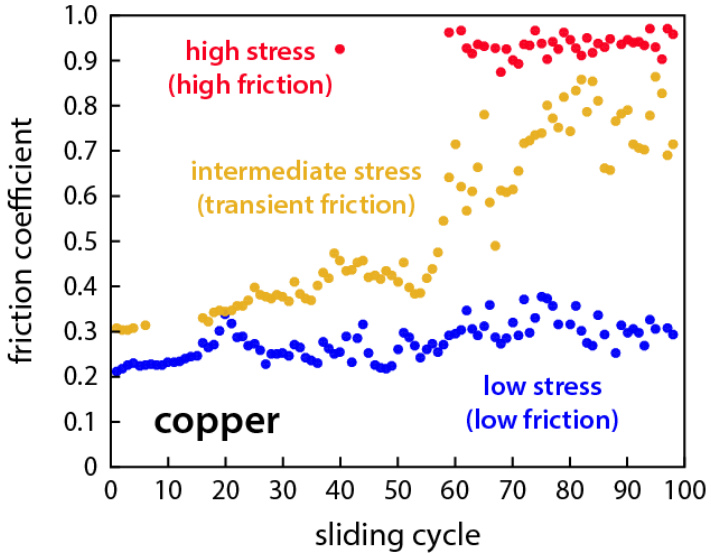


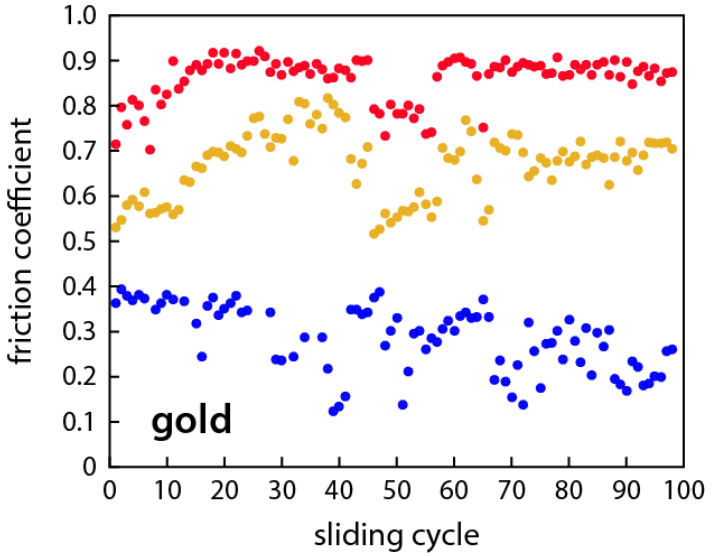
Boundary lubrication (e.g. graphite, MoS_2, engine oil) mitigates commensurate contact –
thus it is possible to achieve low friction at higher normal force

Ok, that was a lot of information. Big picture is…

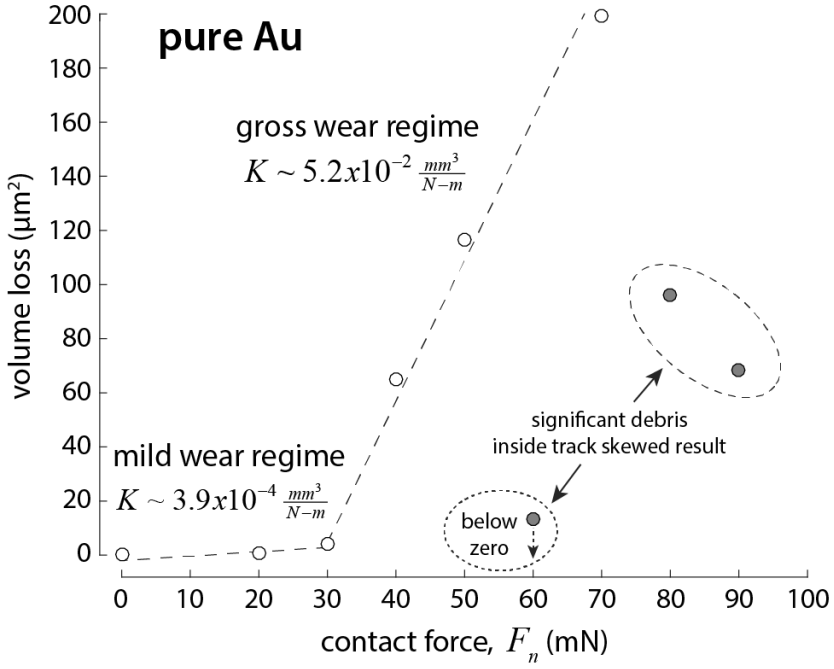


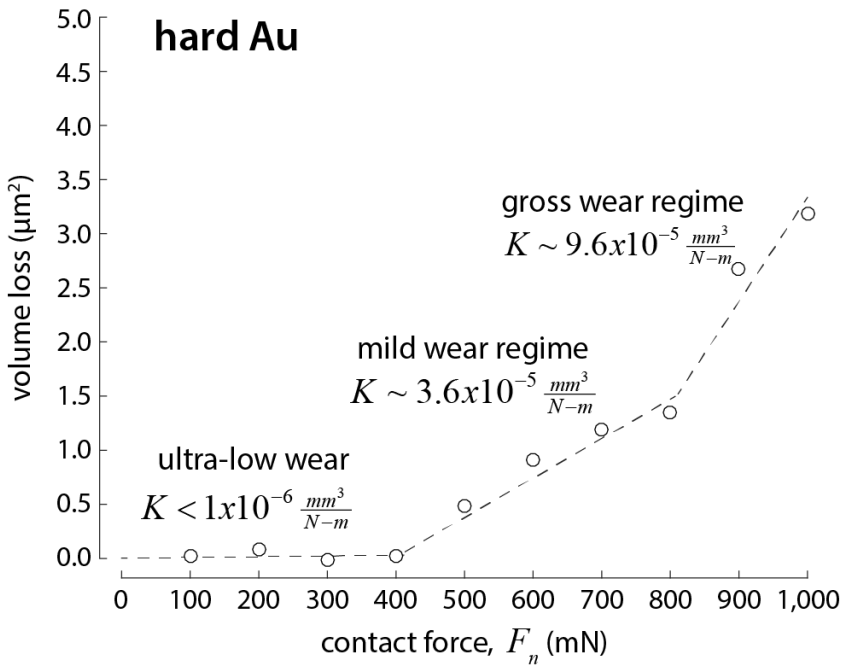






One more look at wear





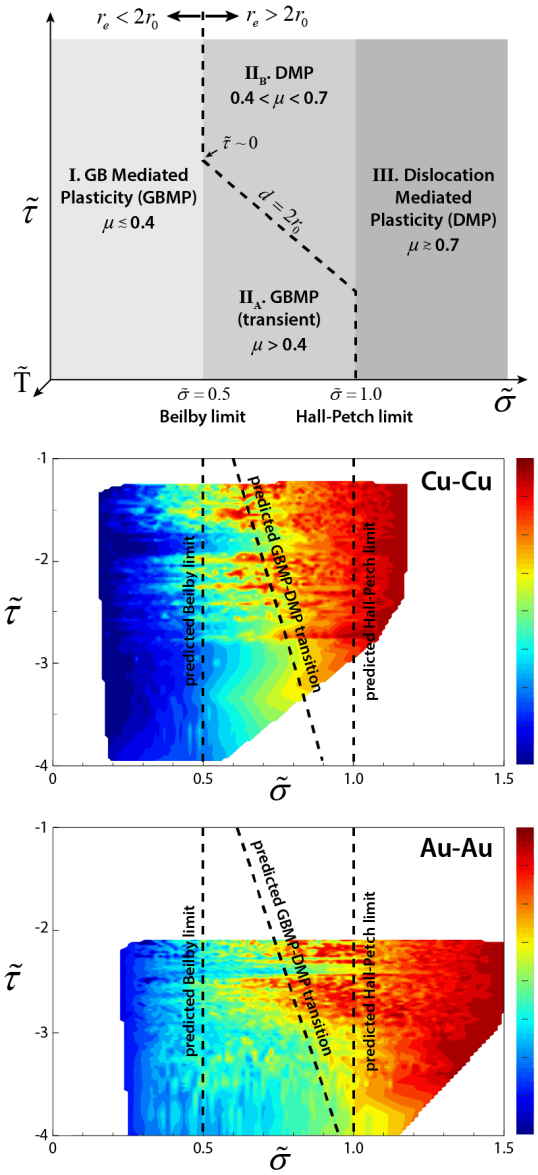
So…

friction regimes described by Hamilton (max surface stress)

wear regimes described by Holm (max bulk stress)

Only the beginning, much left to do… some exciting prospects below

- So far only applied to FCC metals. Apply to BCC metals, ionic solids -- ductility observed in nanoparticles of alumina
- Now exploring the temperature axis: optimizing high current density electrical sliding and rolling contacts
- Clearly there are other regimes and boundaries that have not been identified…
- Low friction regime is result of a competition between wear and stress-driven grain growth
- Can we ***determine*** stacking fault energy or grain boundary mobility for alloys?
- Can we model competing wear? …difficult, but maybe

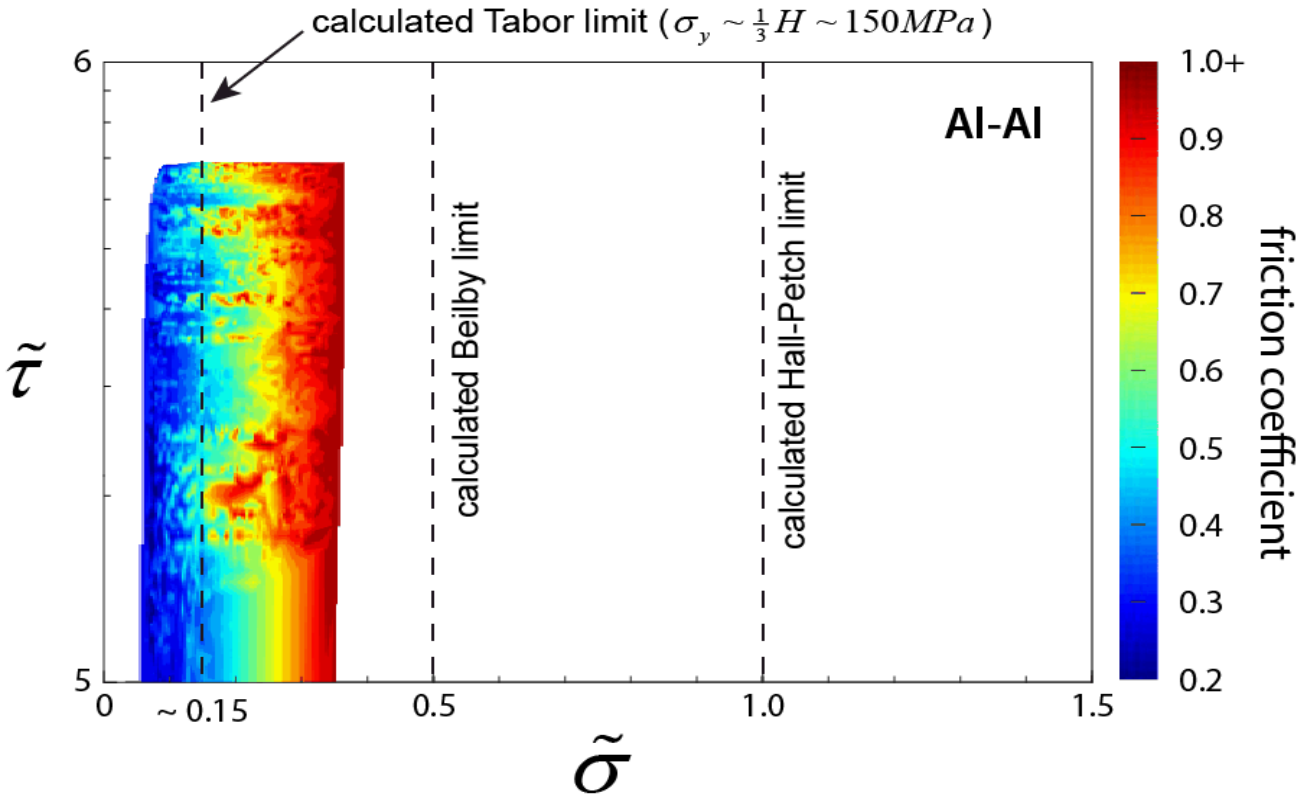


Appendix Slides

Material Properties and Calculated Parameters

property	material system				
	Au	*Cu*	*Al*	*Ni*	*units*
shear modulus, G	27	48	27	76	GPa
Poisson ratio, v	0.44	0.36	0.35	0.31	-
lattice constant, a	4.08	3.61	4.05	3.52	A
Burgers vector, b	2.88	2.55	2.86	2.49	A
SFE, γ_{sf}	45	78	166	128	mJ/m^2
GBE, γ_{gb}	378	625	324		mJ/m^2
HAGB mobility, M_0	3.84×10^{-6}	30	2×10^{-2}		m/s-Pa
HAGB activation energy, Q	1.33	2.01	1.05		$x10^{-19}$ J
calculated parameters					
equilibrium splitting distance, r_0	8.7	5.9	2.0	2.8	nm
σ_∞	312	611	1,117	1,808	MPa
$\sigma(r = 2r_0)$	156	306	580	904	MPa

Aluminum and the Tabor limit



High stacking fault energy… BUT low strength!

Classical attempts to define wear & friction regimes were empirical/phenomenological

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VIEWPOINT SET No. 14

WEAR-MECHANISM MAPS

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(Received August 15, 1989)
(Revised October 16, 1989)

WEAR-MECHANISM MAPPING: THE APPROACH

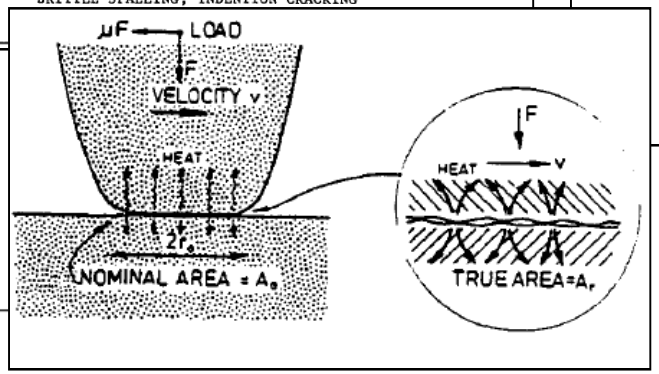
Wear is the loss or transfer of material when contacting surfaces slide. In general, the wear rate W (defined here as the volume loss per unit area of surface per unit distance slid) depends on the bearing pressure F/A_n (where F is the load carried by the contact and A_n is its nominal area), on the sliding velocity, v, and on the material properties and geometry of the surface (Figure 1):

$$W = f(F/A_n, v, \text{Mat. Props.}, \text{Geometry}) \quad (1)$$

But one such equation is not enough. There are many mechanisms of wear, each dependent in a different way on the variables. The dominant mechanism, at any given F and v, is the one leading to the fastest rate of wear. Table 1 lists some of the mechanisms encountered in wear studies of metals and of ceramics; it includes wear by melting, by chemical change induced by frictional heating, by low-temperature plasticity and by brittle fracture.

TABLE 1: MECHANISMS OF WEAR

METALS	CERAMICS
SEIZURE	SEIZURE (?)
MELT WEAR	MELT WEAR
SEVERE-OXIDATIONAL WEAR	THERMALLY-INDUCED STRUCTURE CHANGE
MILD-OXIDATIONAL WEAR	THERMAL CRACKING AND SPALLING
PLASTICITY-DOMINATED WEAR	BRITTLE SPALLING; INDENTION CRACKING
ULTRA MILD WEAR	



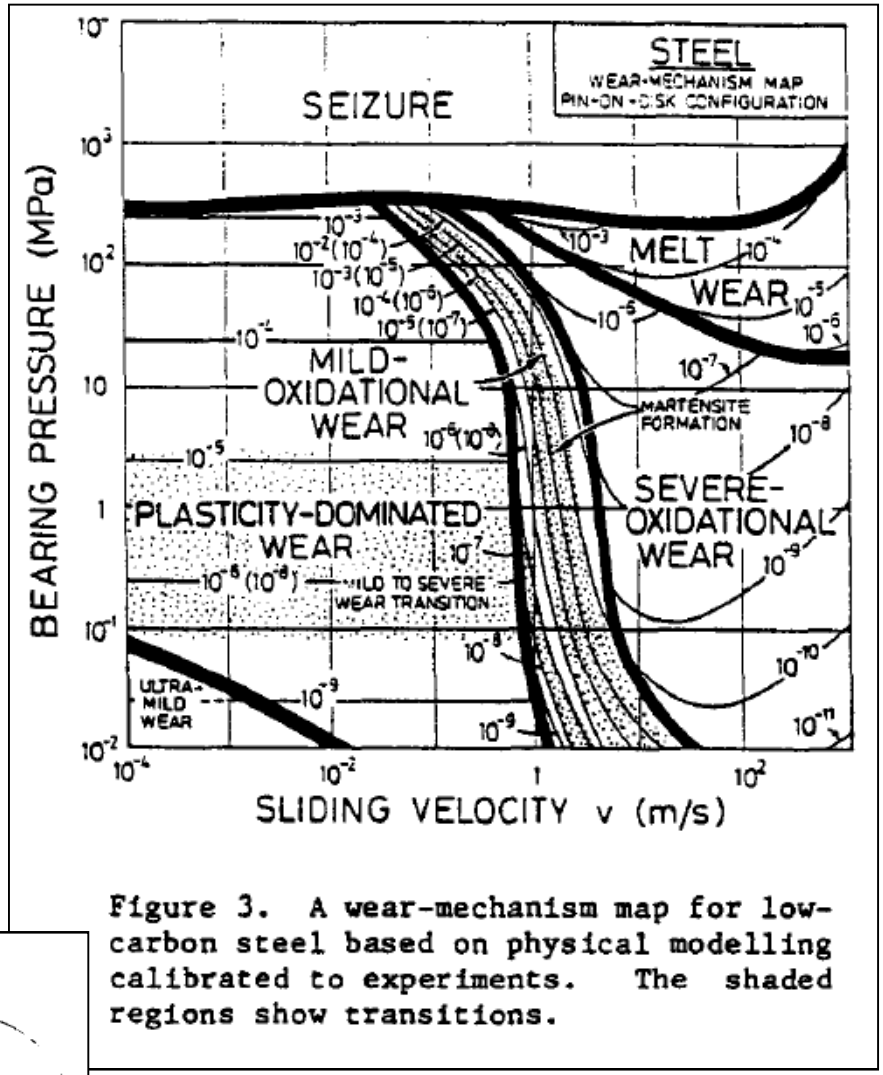
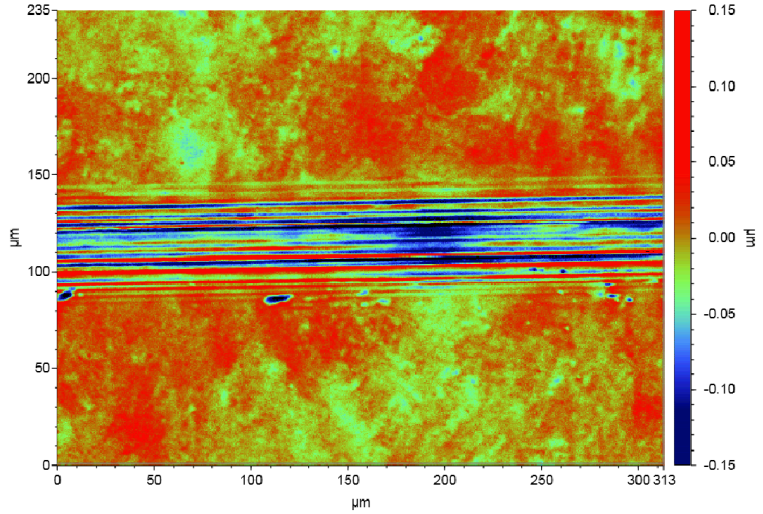


Figure 3. A wear-mechanism map for low-carbon steel based on physical modelling calibrated to experiments. The shaded regions show transitions.

Wear analysis of pure and alloy gold surfaces along wear track for ramped force test

Wear tracks analyzed using a scanning white light interferometer, sample image shown below:



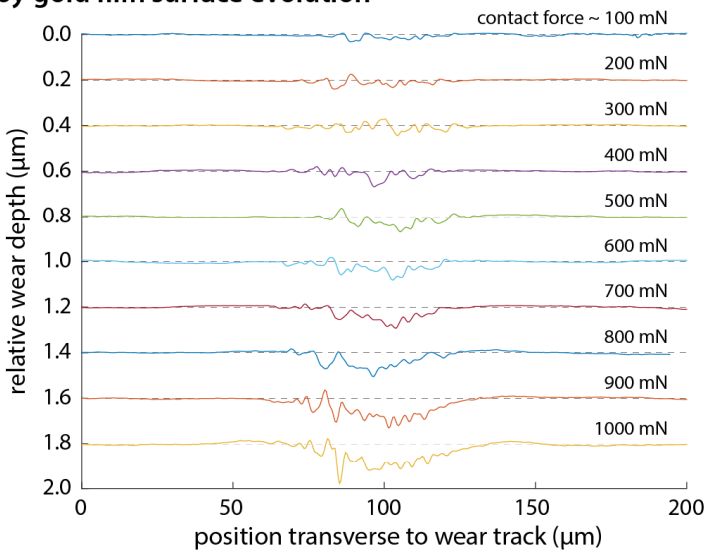
Images taken at 1 mm intervals along 10 mm long wear tracks

Each image then collapsed into a single line plot showing the average wear track cross-section (right images)

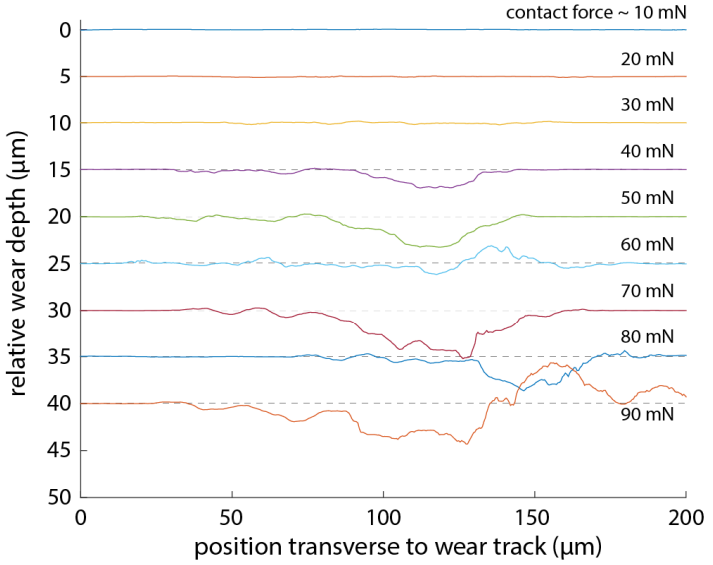
Wear at each interval calculated based on contact force average in this part of the track, number of cycles, and volume loss

Change in contact force along length of image (313 µm) was about +/- 3% of max load

alloy gold film surface evolution

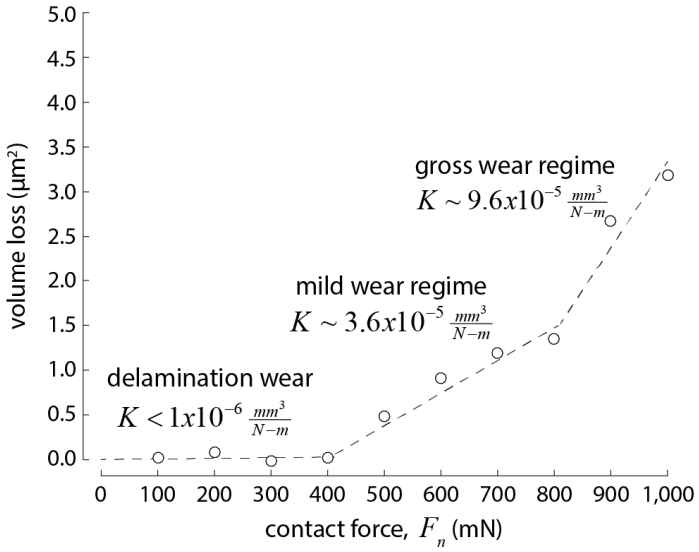


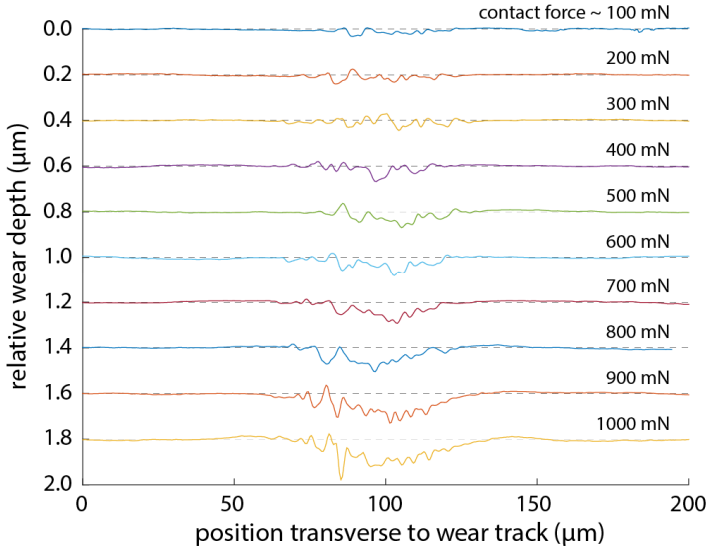
pure gold substrate surface evolution



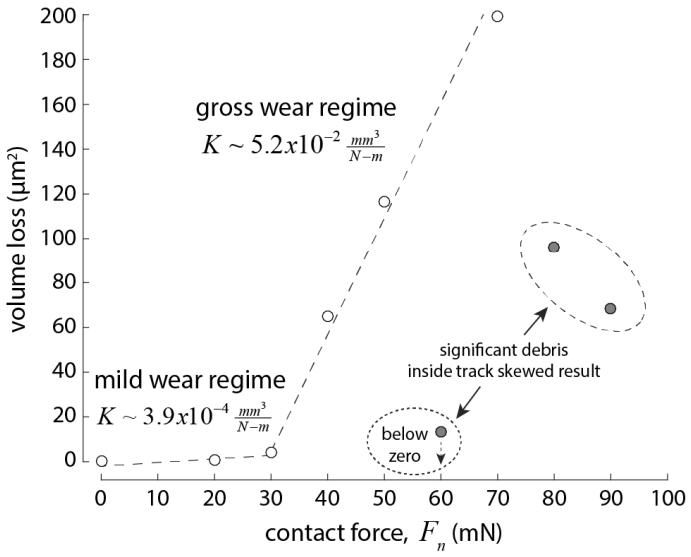
Observed three wear regimes

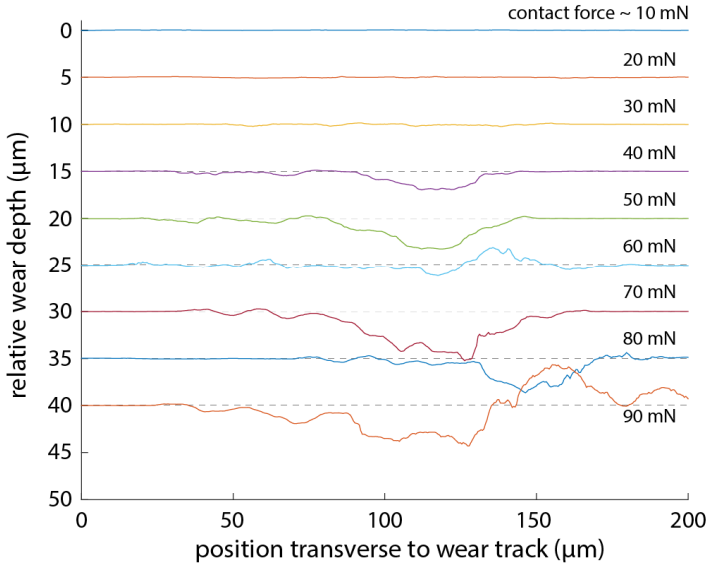
alloy gold film surface evolution





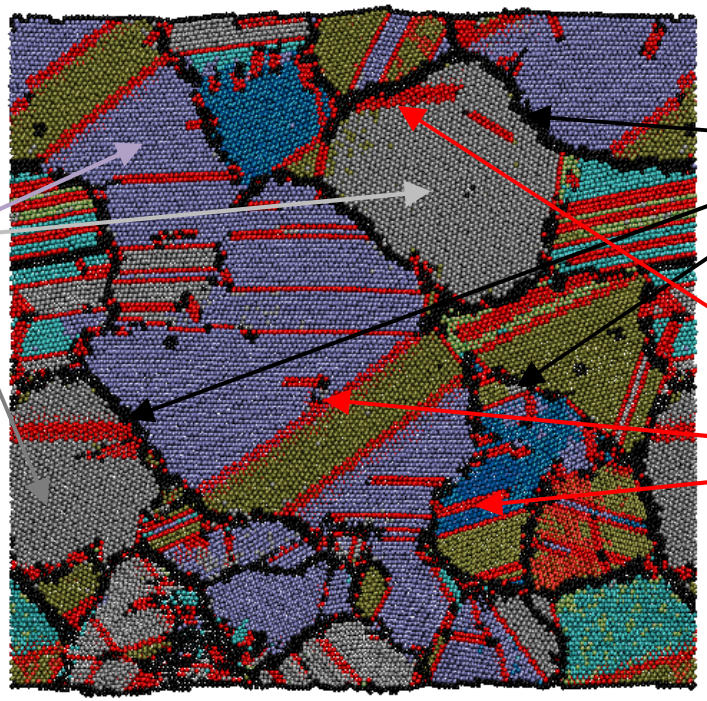
pure gold substrate surface evolution





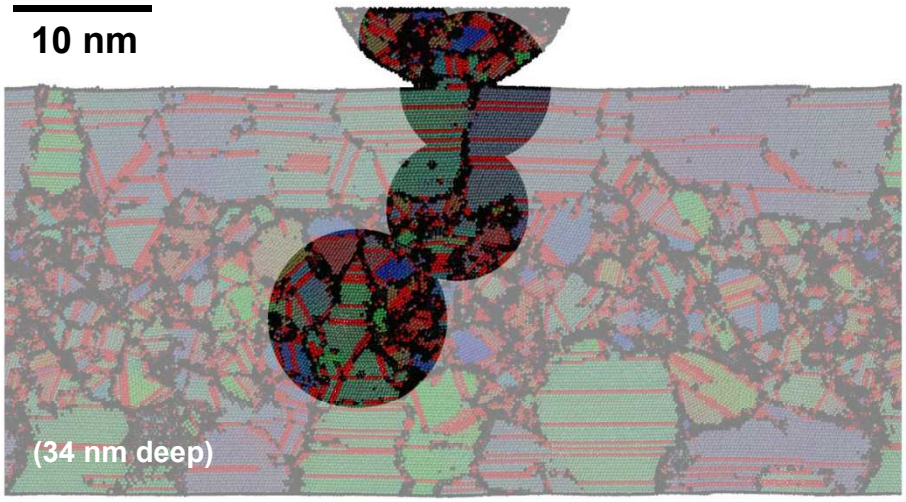
MD Simulations: how to interpret the following images…

Cross-sectional slices of a 3D space filled with atoms



- Locally FCC atoms colored according to Euler angle
- Locally HCP atoms colored red – twins & stacking faults
- Otherwise colored black – grain boundaries

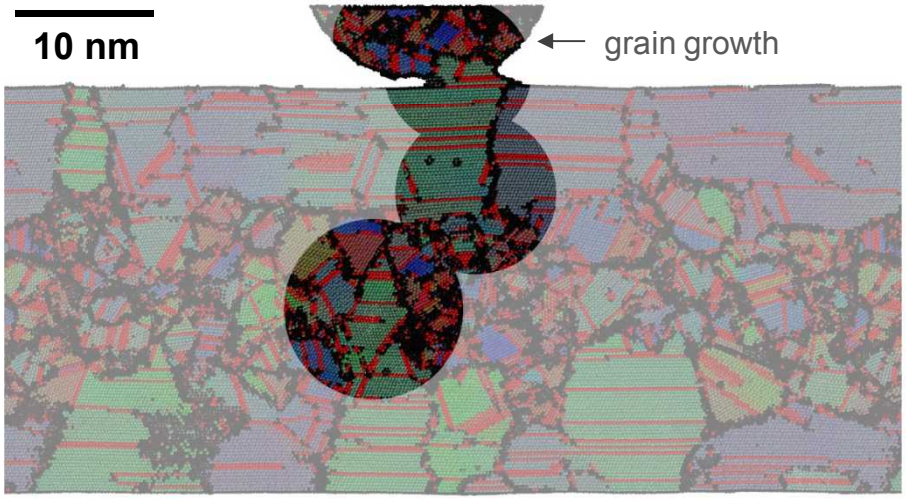
Comparison between pure and alloyed Ag grain evolution: stabilization thru alloying



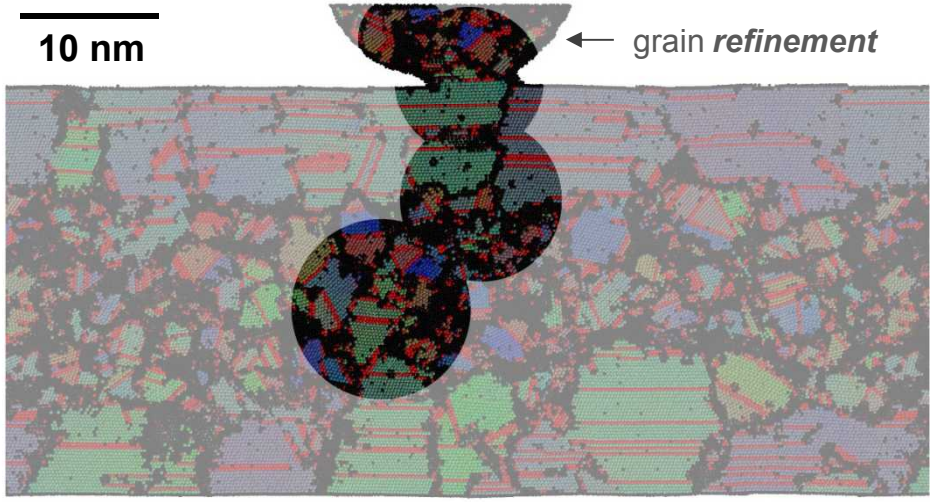
initial microstructure
of Ag and Ag-Cu alloy
(no sliding yet)

300 MPa contact stress
300 K temperature
2 m/s sliding speed

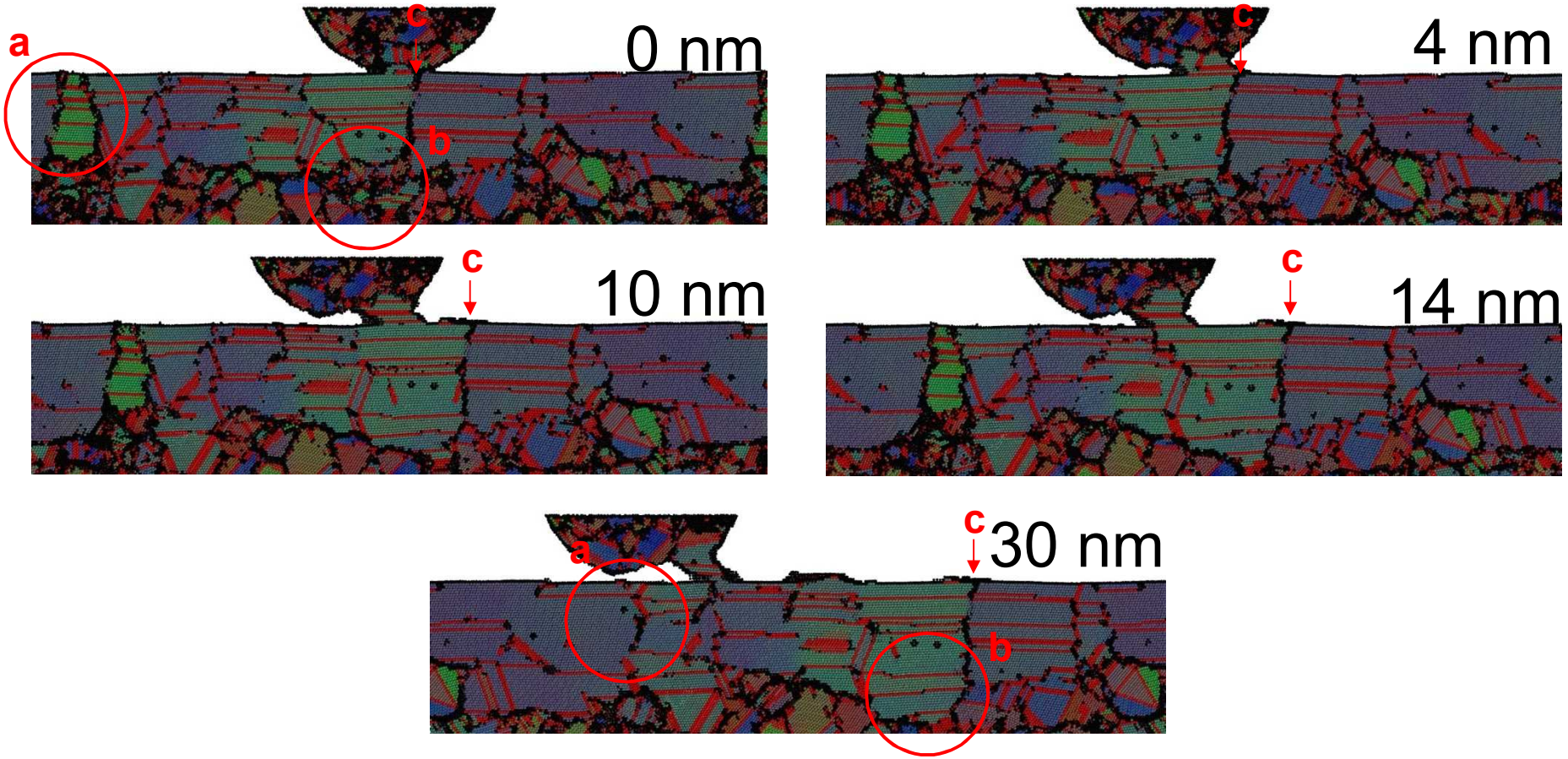
pure Ag after 4 nm of sliding



Ag-10% Cu alloy after 4 nm of sliding

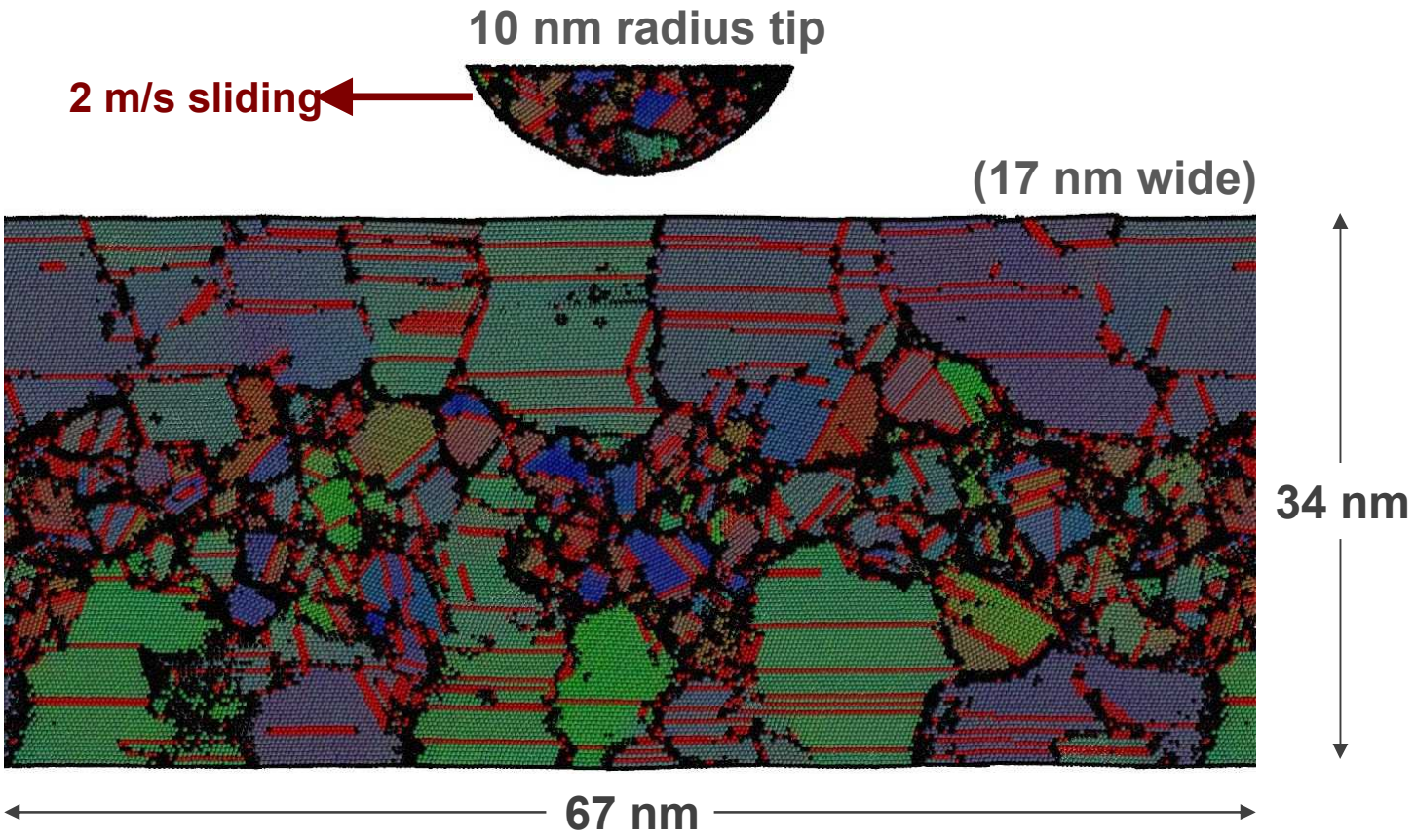


Another look, now at pure Au tip/slab contact evolution over a longer sliding time



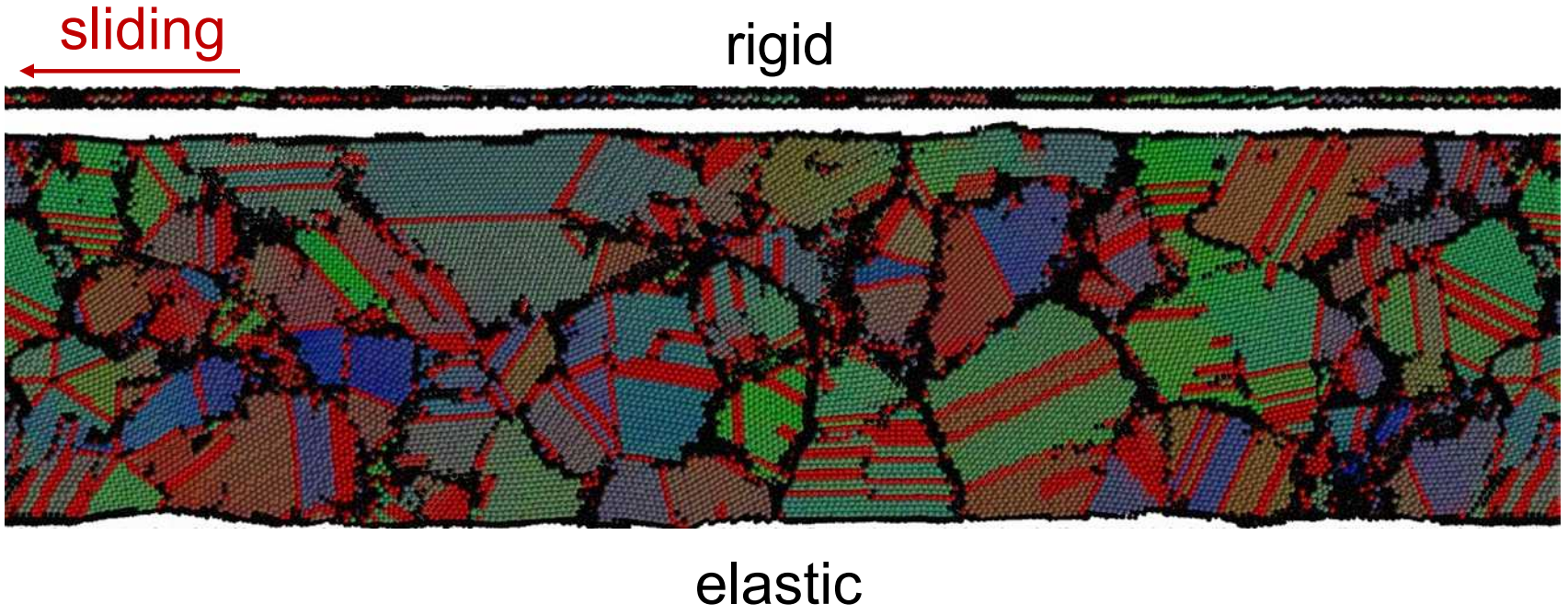
- Initially distinct grains
- After shear (adhesive load), coalescence – now a mode II crack
- Single grain forms across interface – stress induced grain growth

Tip based friction simulations : this is what the initial condition looks like



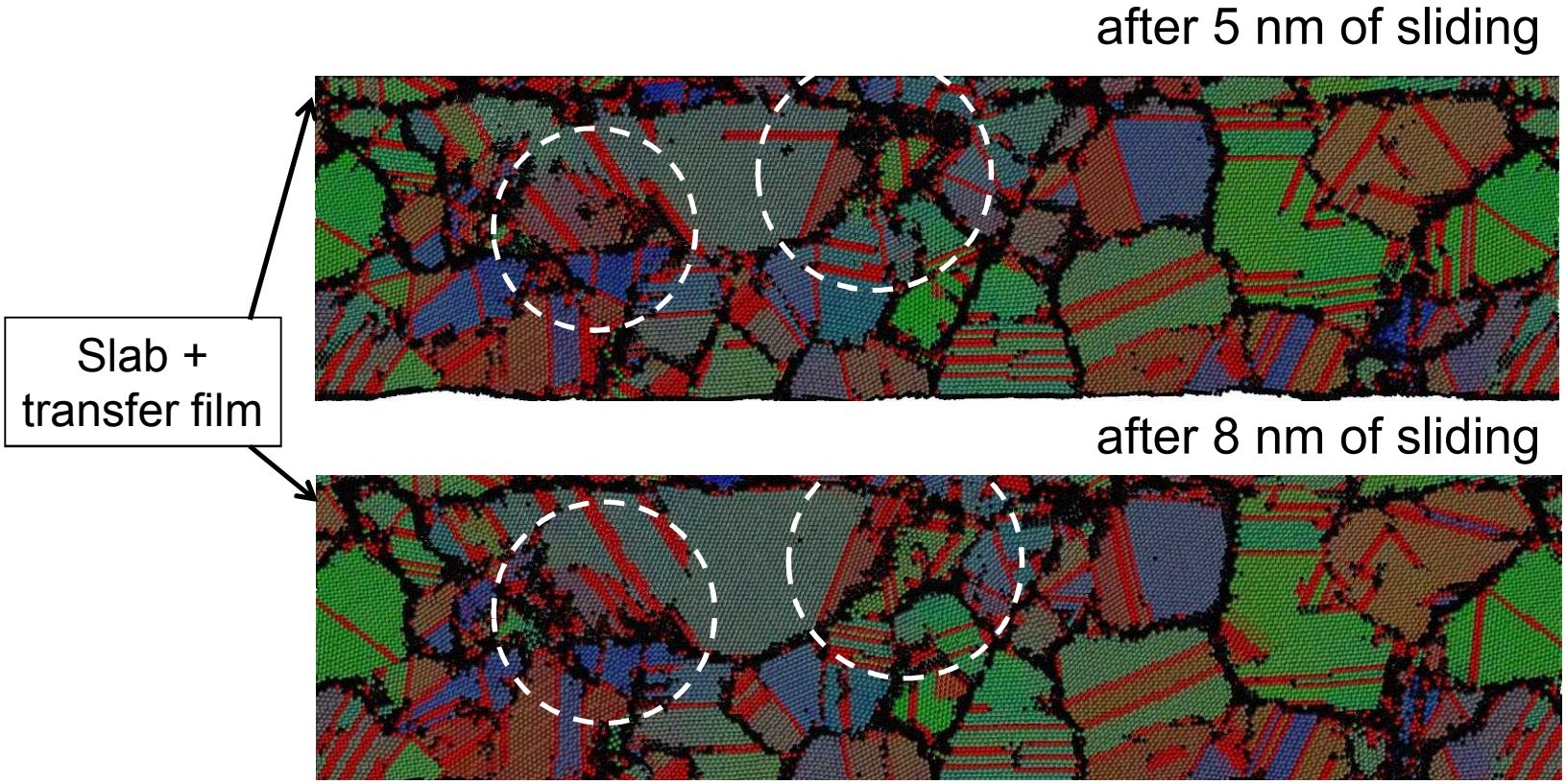
Substrate: nanocrystalline Ag
Constraint 1: constant velocity
Constraint 2: constant separation ***or*** normal force

Slab-on-slab sliding contact simulations remove wear, enable friction quantification



- Rigid slabs suppress grain growth
- No plowing is possible/reduced contact stress

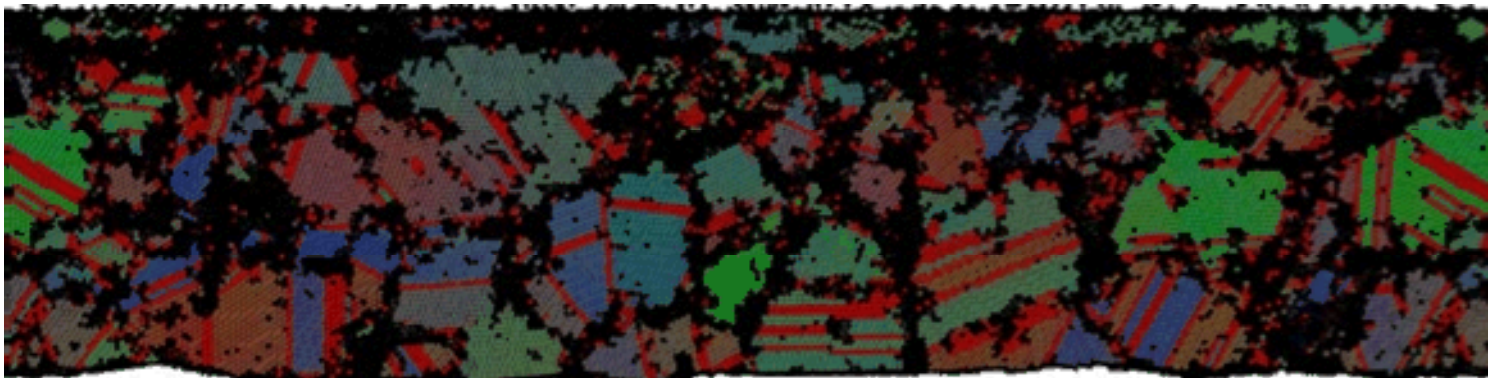
Sliding of pure Ag slabs



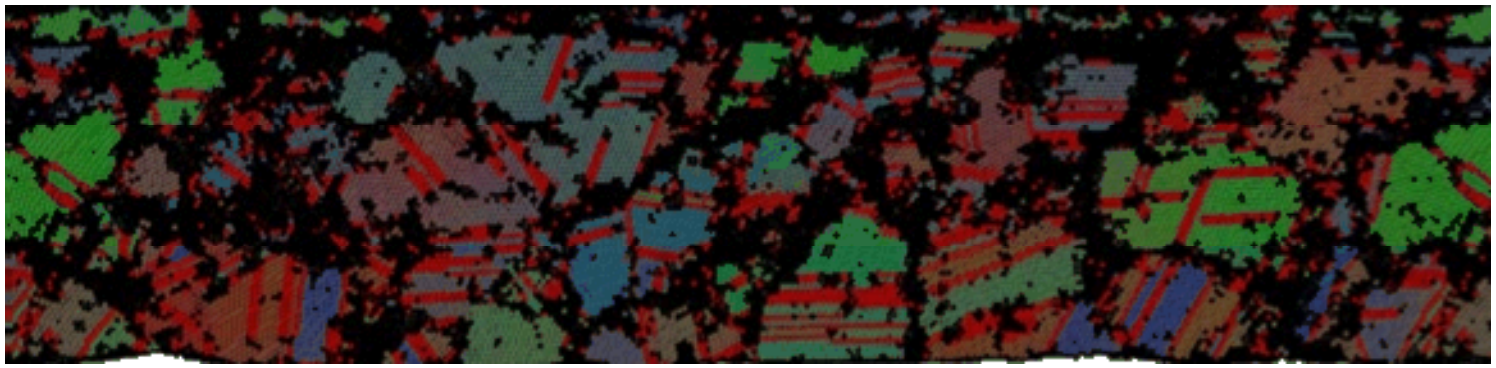
- Slight grain growth, forms transfer film
- Slides along transfer film grain boundaries or nearby stacking faults depending on availability

Sliding of Ag alloy (10% Cu) contact

after 6 nm of sliding



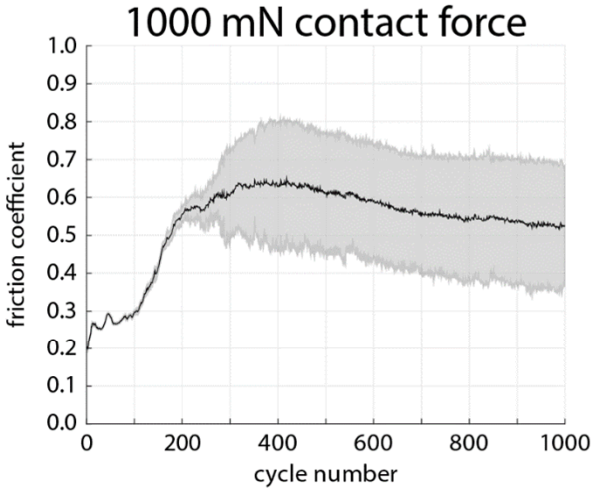
after 16 nm of sliding

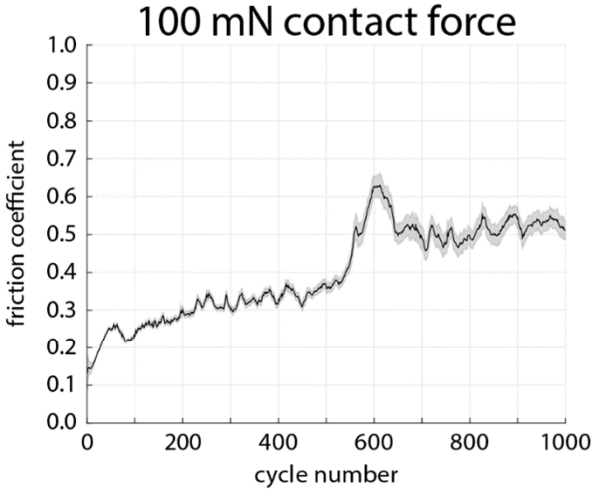


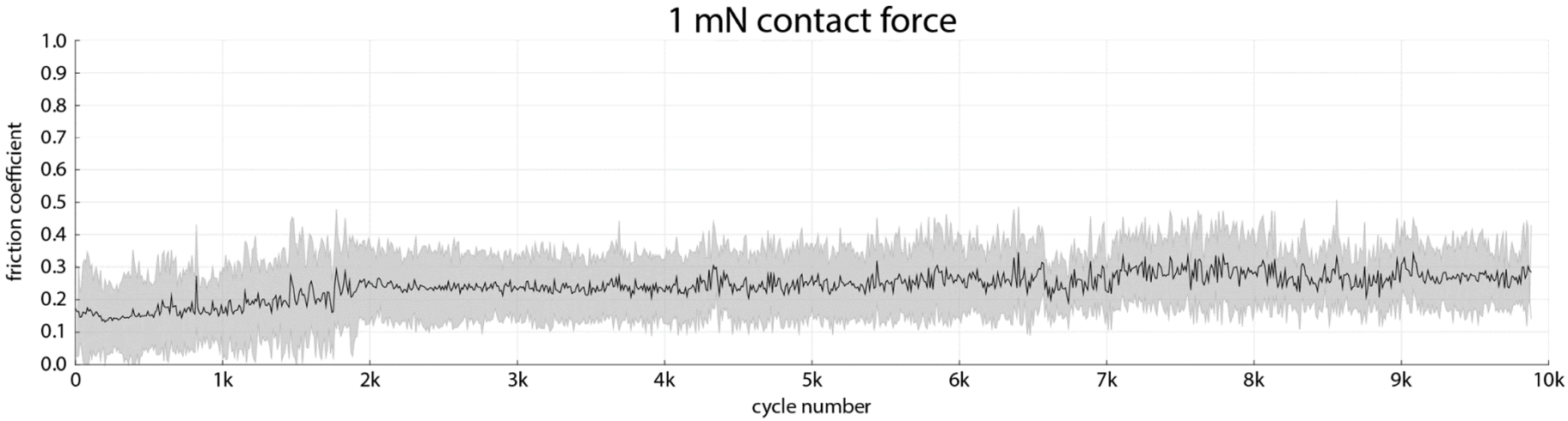
- Alloy slides at transfer film boundary, but also throughout substrate
- The pure Ag slabs on previous slide started with the exact same microstructure (lots of coarsening on the pure Ag slabs simulation!)

Three regimes observed for 60Ni-40W at.% vs sapphire in oxidizing environment

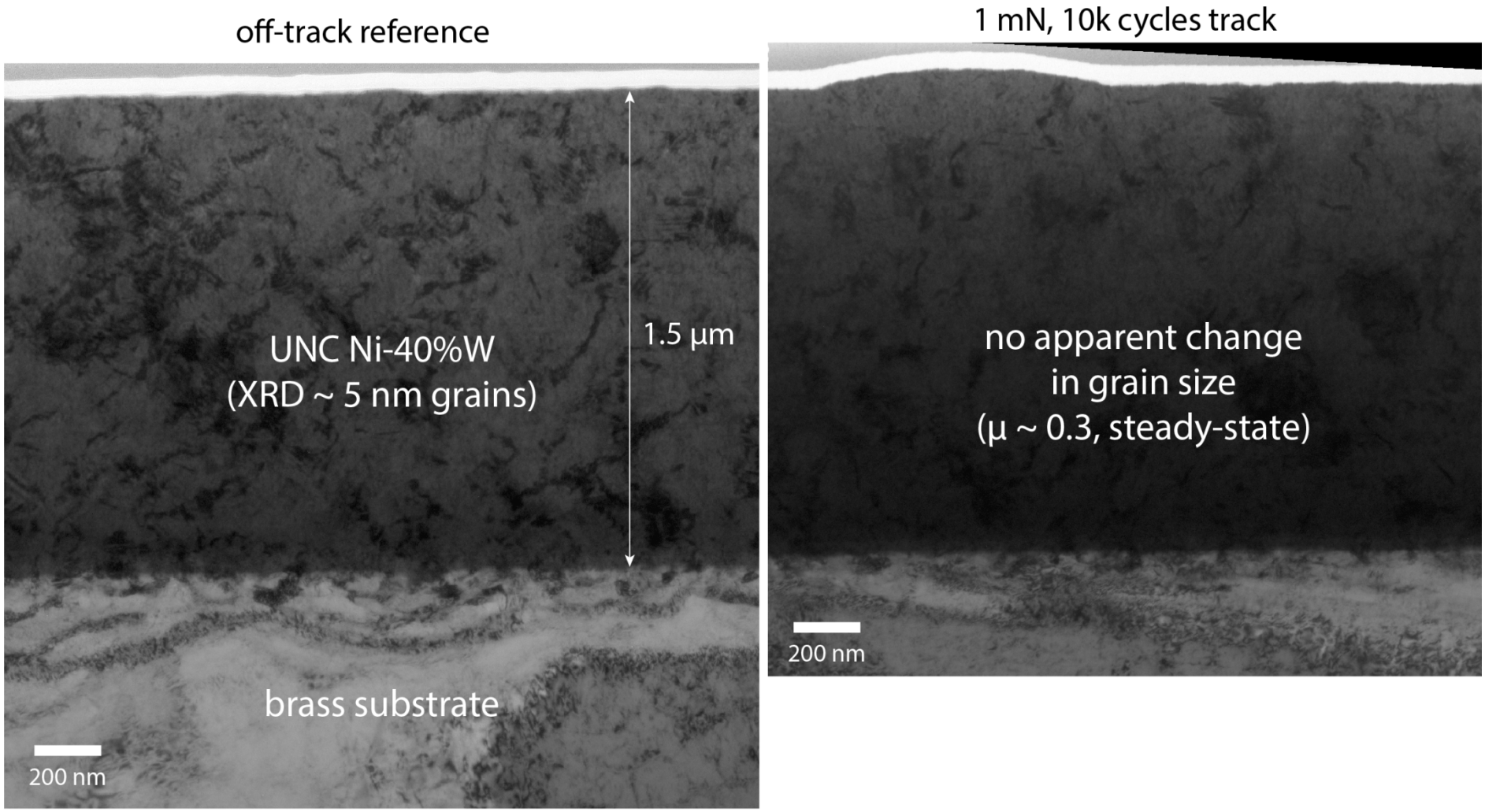
1 mm/s sliding speed
three contact forces used
bidirectional sliding
2mm long track
sapphire ball 1.6 mm diameter
sliding in ***lab air***



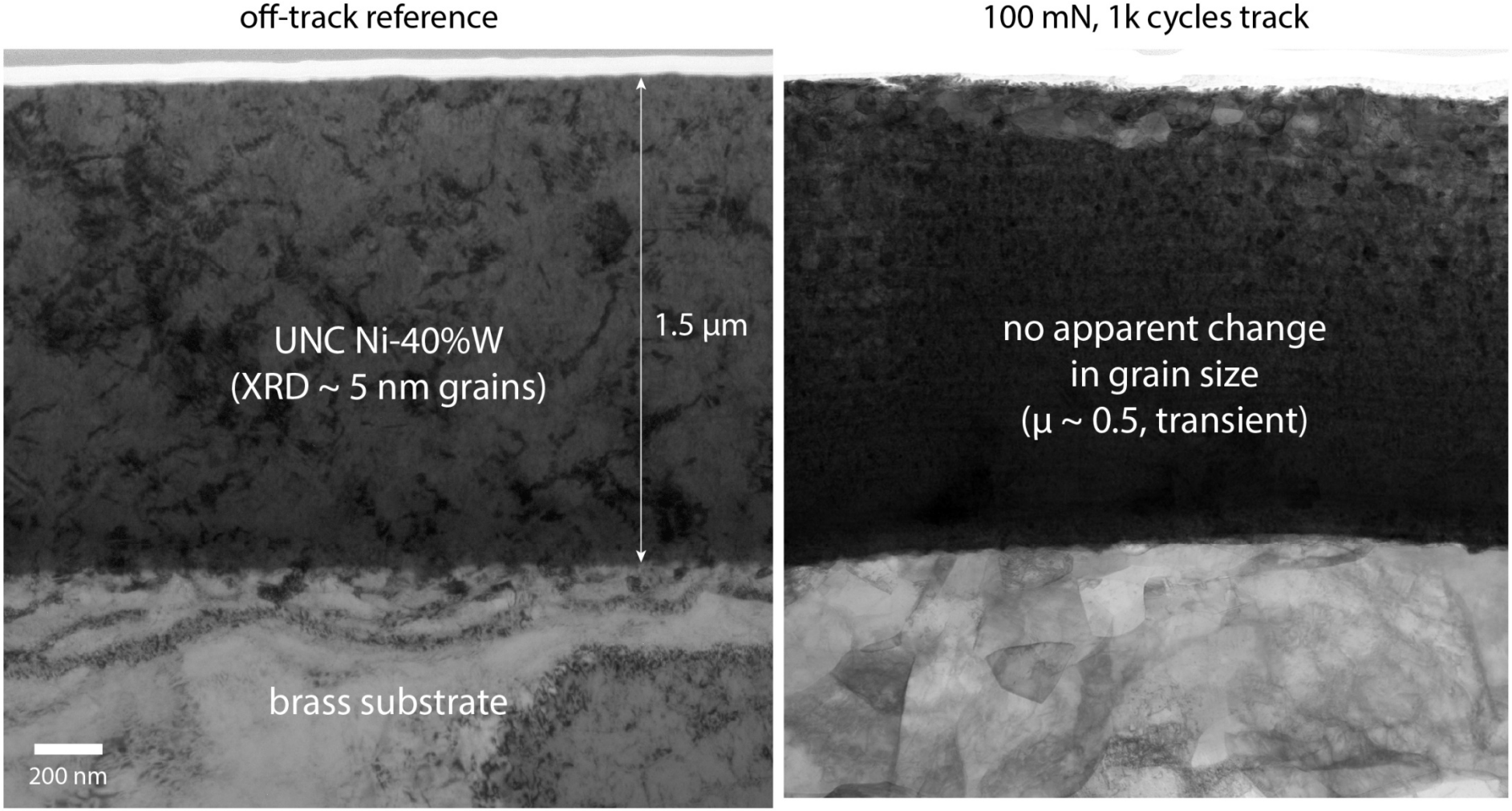




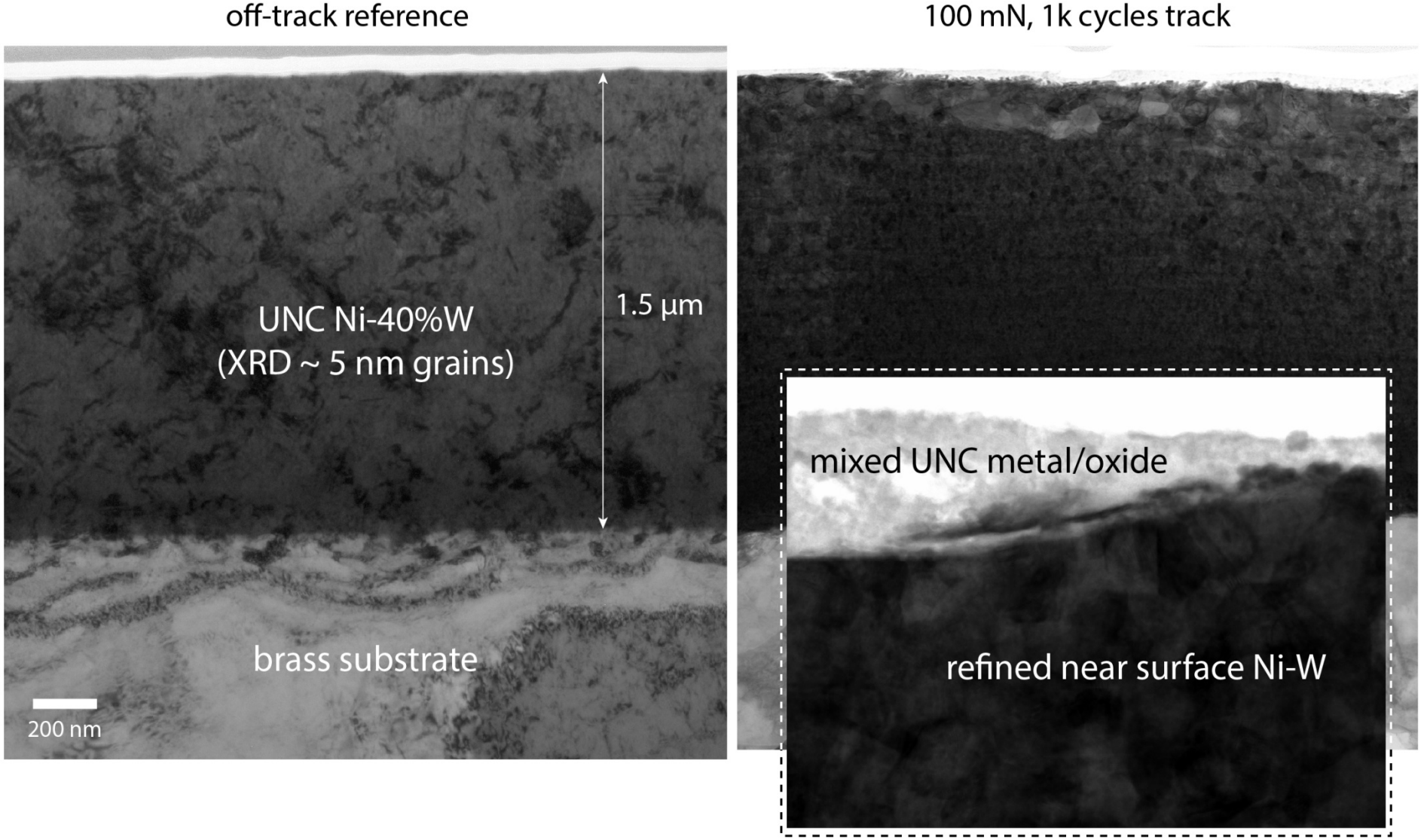
FIB-TEM wear track cross-section of 1 mN normal force / 10k cycle test



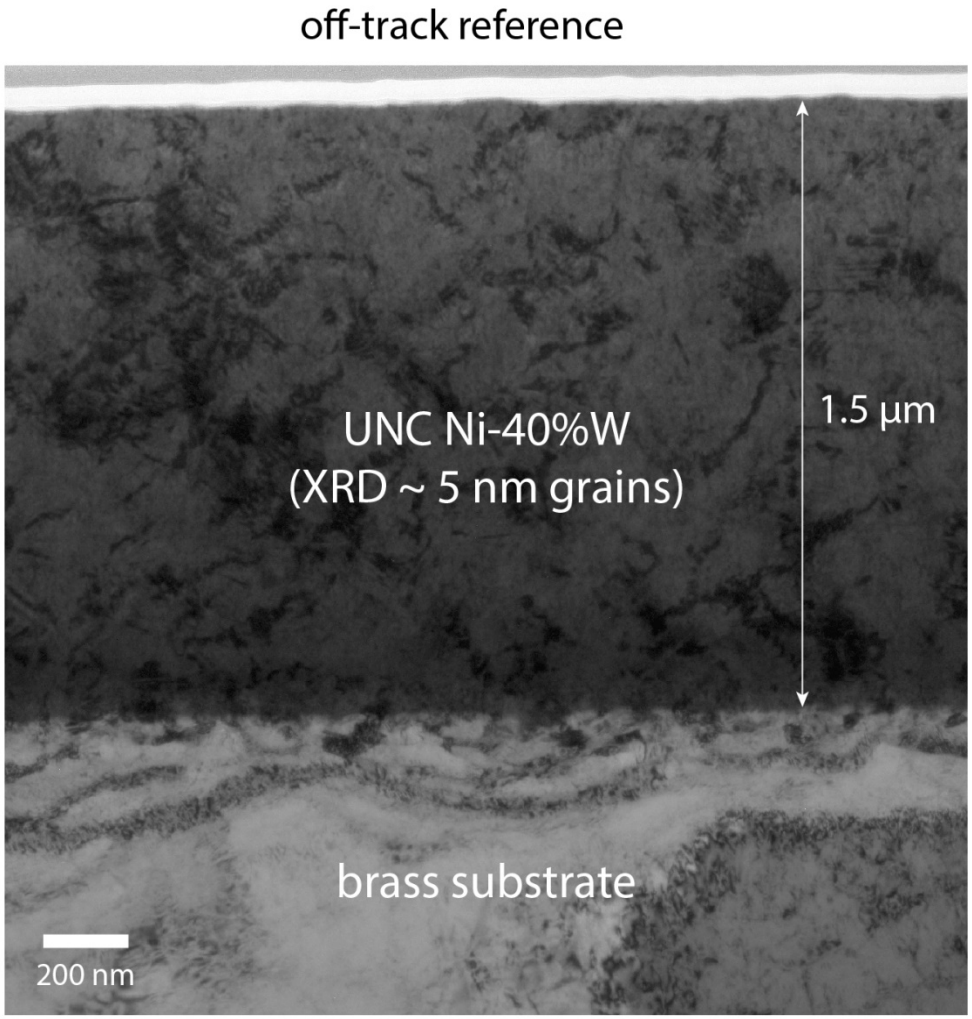
FIB-TEM wear track cross-section of 100 mN normal force / 1k cycle test

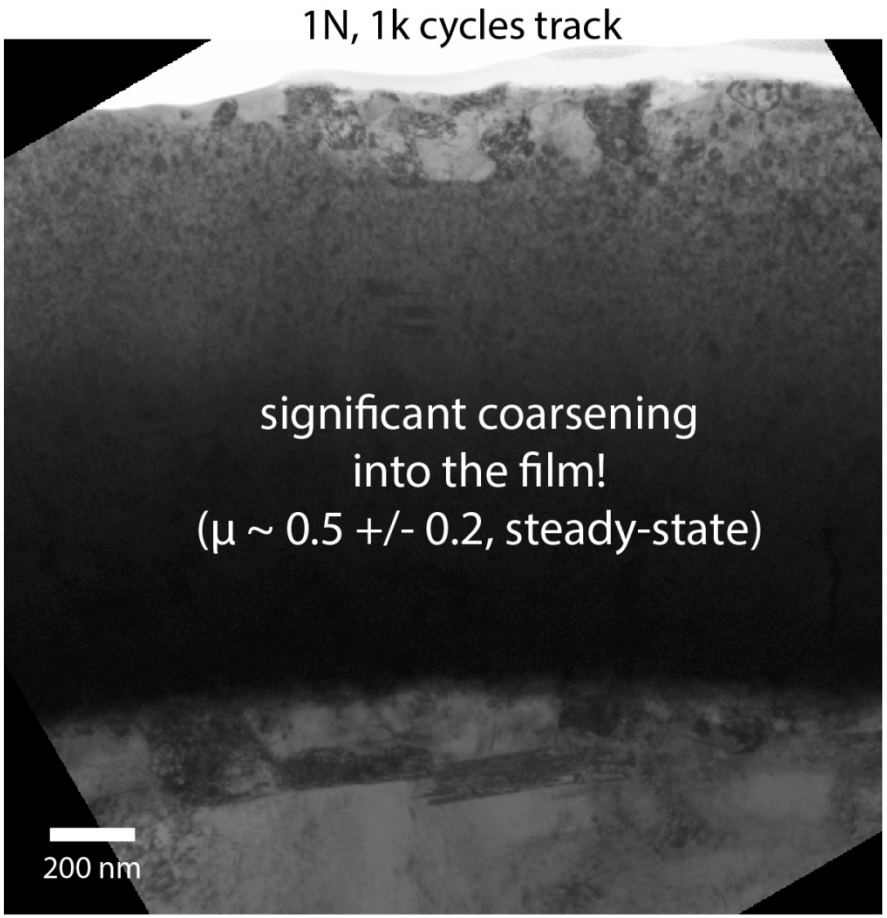


FIB-TEM wear track cross-section of 100 mN normal force / 1k cycle test



FIB-TEM wear track cross-section of 1000 mN normal force / 1k cycle test





Disruptive breakthrough in 2012: *intrinsic thermal stability* possible with NC alloys!

Regular Nanocrystalline Solution (RNS) model

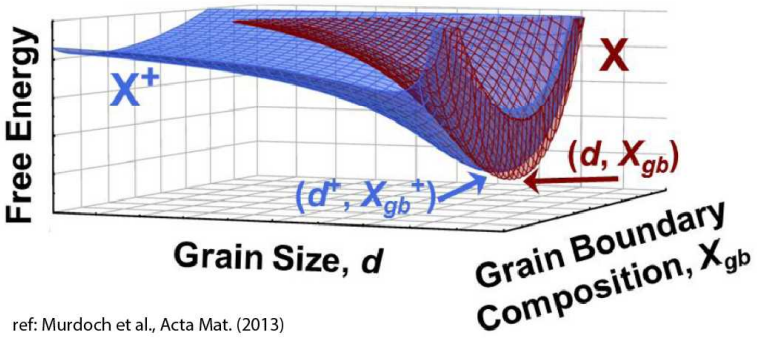
ref: Chookajorn et al., Science, 2012

$$\Delta G^{mix} = (1 - f_{gb})\Delta G_c^{mix} + f_{gb}\Delta G_{gb}^{mix} + z\nu f_{gb}(X_{gb} - X_c)\left[(2X_{gb} - 1)\omega_{gb} - \frac{1}{zt}(\Omega^B\gamma^B - \Omega^A\gamma^A)\right]$$

change in Gibbs free energy is positive, but ***local minimas exist***!

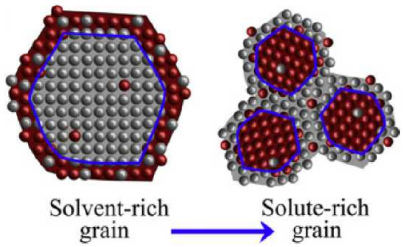
Implications:
Will not drive toward fine grain size, but will remain there

Two examples of predicted nanocrystalline intrinsic stability for global solute concentrations (X and X+) for a W-based binary alloy:



ref: Murdoch et al., Acta Mat. (2013)

Grain structure model: segregated 2-phase metal system:



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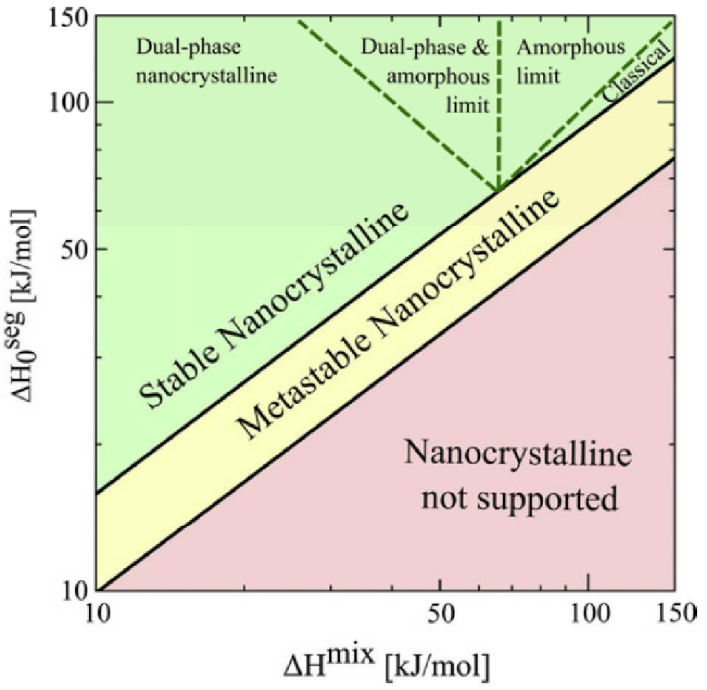
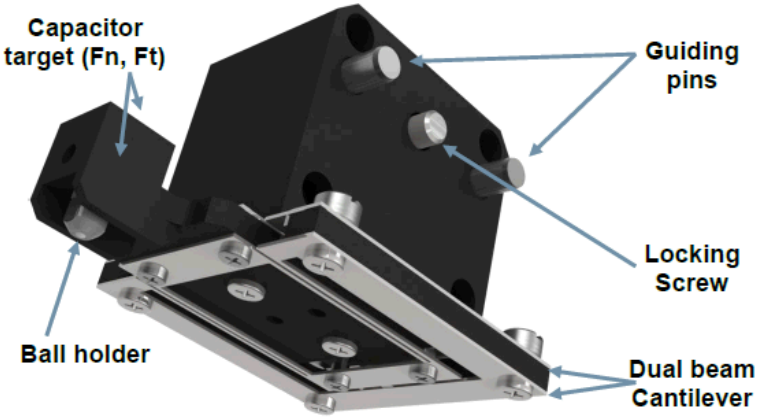


$$\Delta H_o^{seg} = z\left(\omega_c - \frac{\omega_{gb}}{2}\right) \qquad \Delta H^{mix} = z\omega_c X(1-X)$$

General condition for stability: $\Delta H_o^{seg} > \Delta H^{mix}$

Modified CSM Nanotribometer – friction and wear testing platform



CSM nanotribometer modified for 4-wire ECR measurement

- DC power supply
- nano-ohm meter

Test parameters:

- Fn = 100 µN to 1000 mN
- pin radius = 1.6 mm
- track length = 0.1 to 10 mm
- v = 0.01 to 10 mm/s