Karlsruhe Institute of Technology

Institute for Applied Materials

Plasticity

Lecture for “Mechanical Engineering” and “Materials Science and Engineering”

Dr.-Ing. Alexander Kauffmann (Bldg. 10.91, R. 375)
Prof. Martin Heilmaier (Bldg. 10.91, R. 036)

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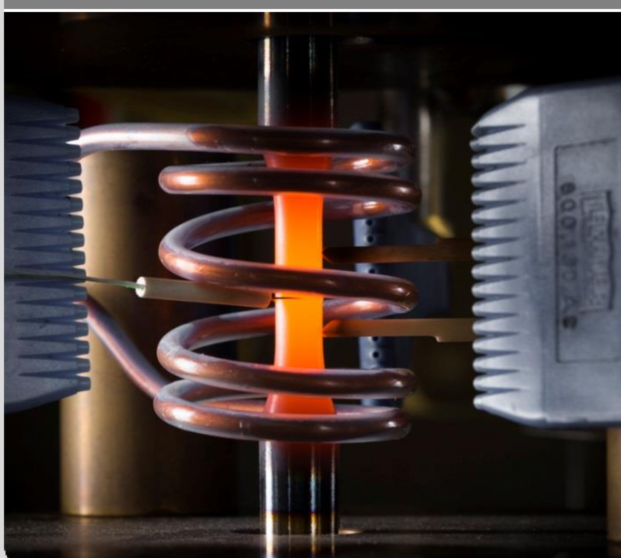
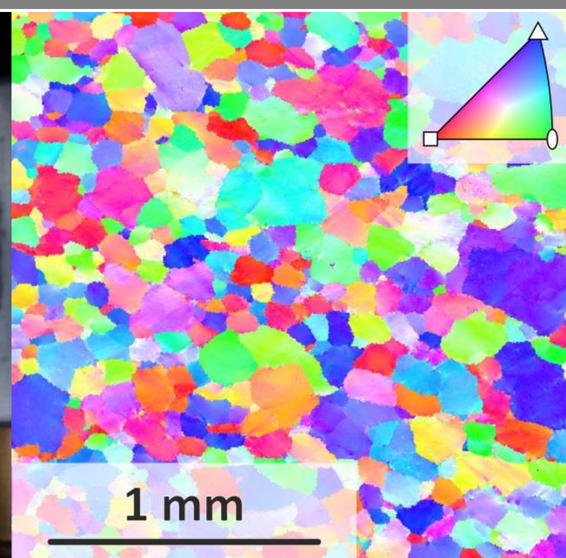
Institute for Applied Materials (IAM-WK)





Topics

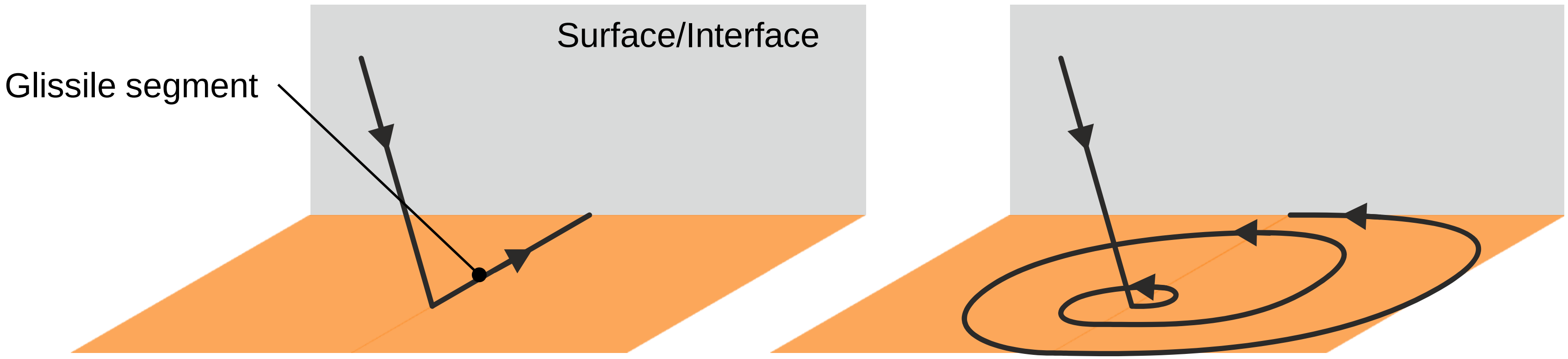
- Multiplication of Dislocations
  - Nucleation
  - Glide Sources (Single-Arm, Two-Arm)
  - Cross-Slip Sources
  - Climb Sources
  - Helical Source
  - Other Sources of Dislocations

Nucleation

- In Ch. 3d, we have seen that **vacancies spontaneously form at finite temperature** due to rather low formation enthalpy and gain in entropy by a configurational contribution.
- In Ch. 4c, we have seen that the **formation enthalpy of dislocation lines is orders of magnitude larger**. Hence, there is nothing like an equilibrium density.
- At $0\ \mathrm{K}$ (no thermal fluctuations), the critical stress to form a dislocation loop is in the order of $\tau_c \approx \frac{G}{10}$ **at a critical size of the loop of $3b$ (same meaning of critical as in typical nucleation theory for other subjects)** . If smaller **stresses in the order of ordinary yield strength of metals** are considered, the **critical sizes are in the order of several hundred b**. Even increasing temperatures, cannot sufficiently support the nucleation of loops.

Single-Arm Sources

- In Ch. 4c, we have seen that a **glissile dislocation segment can rotate** about a sessile segment.
- **By every revolution**, the **slip step** on the surface of the crystal **increases by one Burgers vector**.
- The **glissile segment tends to form a spiral** due to the relaxed structure of the dislocation at the surface/discontinuity of the crystal. The length of the dislocation line increases:



Single-arm glide source: initial state and after several revolutions

Single-Arm Sources

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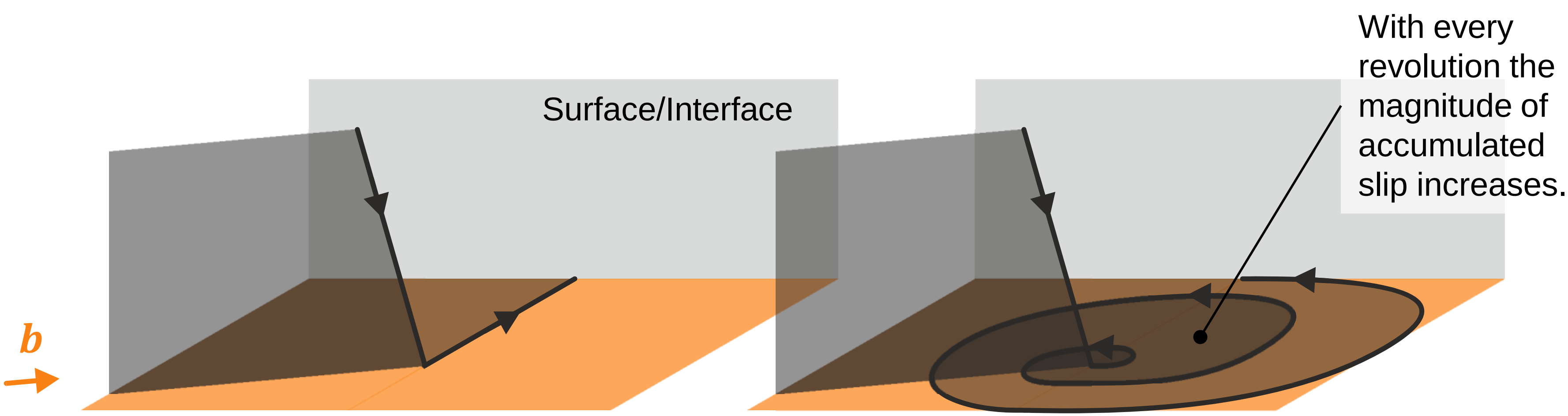
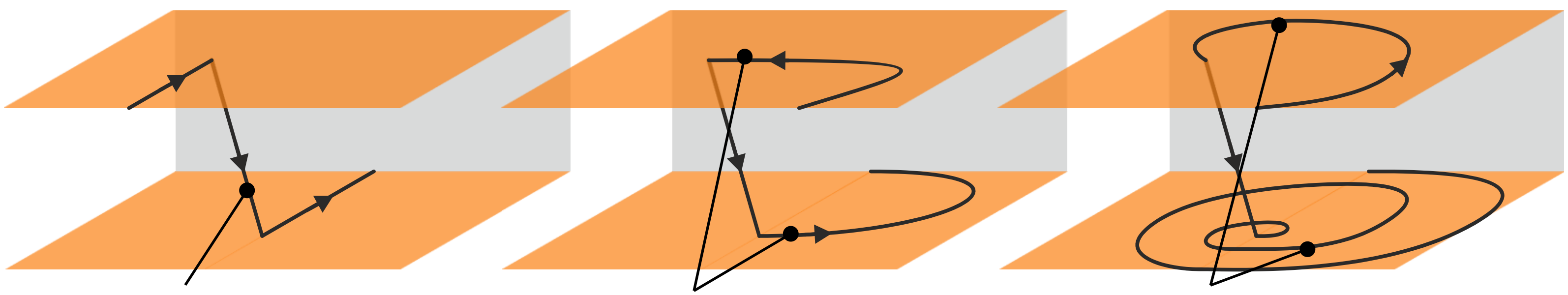


Slipped regions of the same.

Single-Arm Sources

- In case of a formation of the single arm by dislocation intersection (**jog**), the **stress for passing has to be considered**. As seen in Ch. 4d, the stress for passing by significantly increases for closely spaced dislocations. Hence, many intersection processes have to produce sufficiently large jogs:



jog of l in length

formation of an extended dipole (anti-parallel line sense!) in case of a blocked source

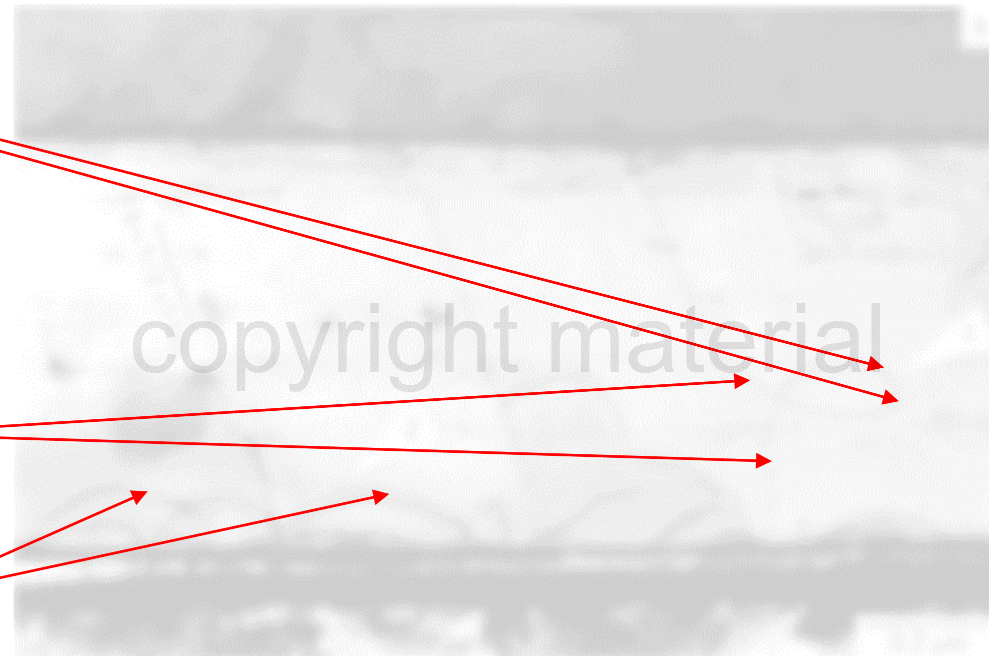
in case the stress is sufficient for passing, the source can be activated

Single-Arm Sources

jog

single-arm sources

emission of dislocations at interfaces



Dislocation multiplication TiAl (Ti-48Al-2Cr) after compression testing at 300 K to 3 % strain.

F. Appel et al.: “Recent Progress in the Development of Gamma Titanium Aluminide Alloys” in Adv. Eng. Mat. 2 (2000) 699-720

Single-Arm Sources

- Glide component of the interaction force between two anti-parallel edge dislocations (Ch. 4d):

$$\frac{F_x}{L} = -\frac{G\,b^2}{2\pi\,(1-\nu)}\frac{x\cdot(x^2-y^2)}{(x^2+y^2)^2}$$

- Maximum force:

$$\frac{F_x^{\max}}{L} = \frac{G\,b^2}{8\pi\,(1-\nu)}\frac{1}{l}$$

- Glide component of the interaction force between two anti-parallel screw dislocations (Ch. 4d):

$$\frac{F_x}{L} = -\frac{G\,b^2}{2\pi}\frac{x}{x^2+y^2}$$

- Maximum force:

$$\frac{F_x^{\max}}{L} = \frac{G\,b^2}{4\pi}\frac{1}{l}$$

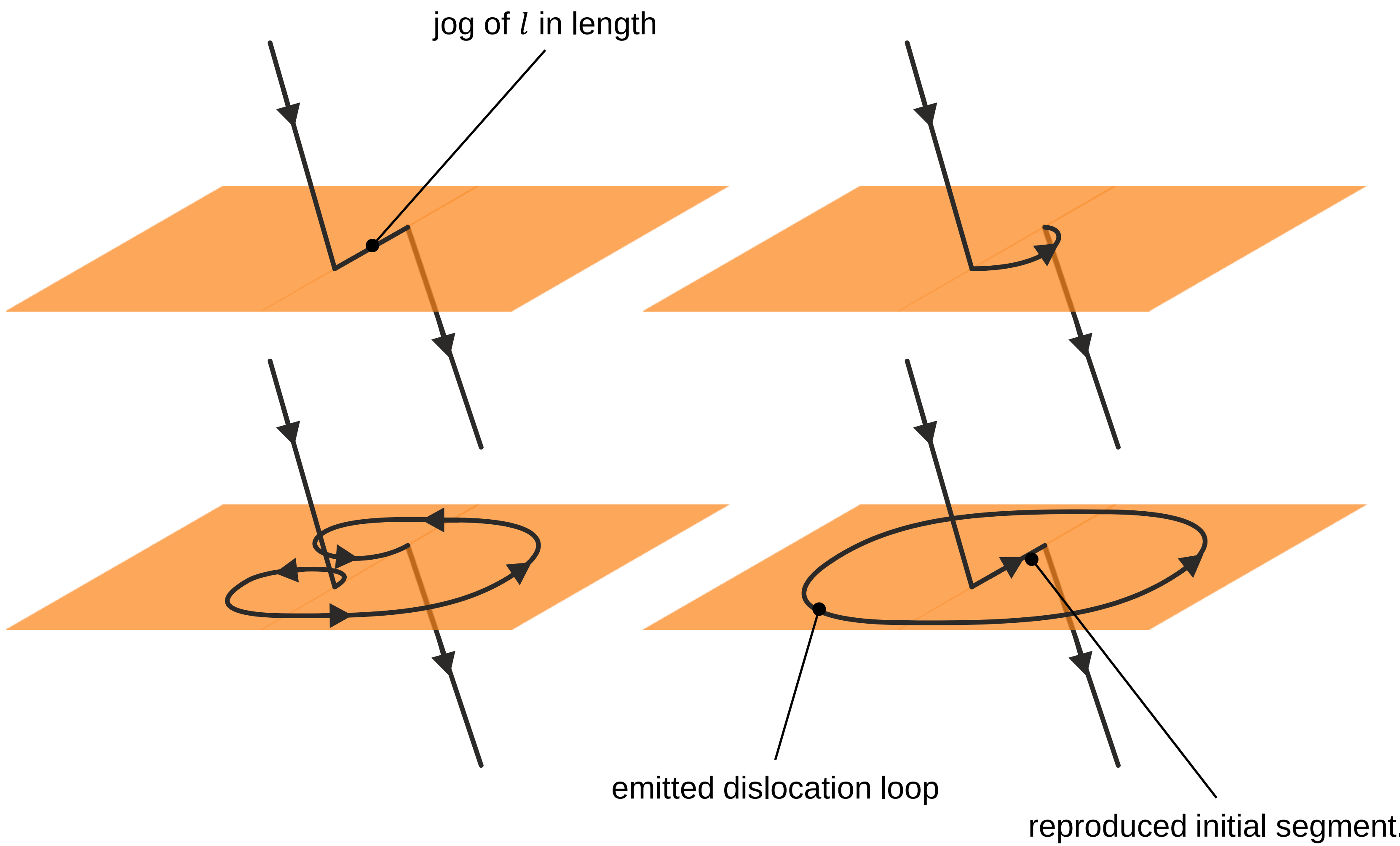
Single-Arm Sources

- The **critical stress for the activation of the source** is: $\tau_\mathrm{c}\, b = \frac{F_x^\mathrm{max}}{L}$.
- Hence, for the single-arm source $\tau_\mathrm{c} = \left(\frac{G\, b}{8\pi\,(1-\nu)} \dots \frac{G\, b}{4\pi}\right)\frac{1}{l}$ follows.
- For **reasonable yield strengths** of roughly $\tau \approx \frac{G}{1000} \dots \frac{G}{100}$, the jog must be of $l \approx 6 \dots 60\ b$ **or** $8 \dots 80\ b$ **in length** for edge or screw dislocations, respectively.
- The lower bound is realistic for jog-forming intersections.

Two-Arm Source: Frank-Read Source

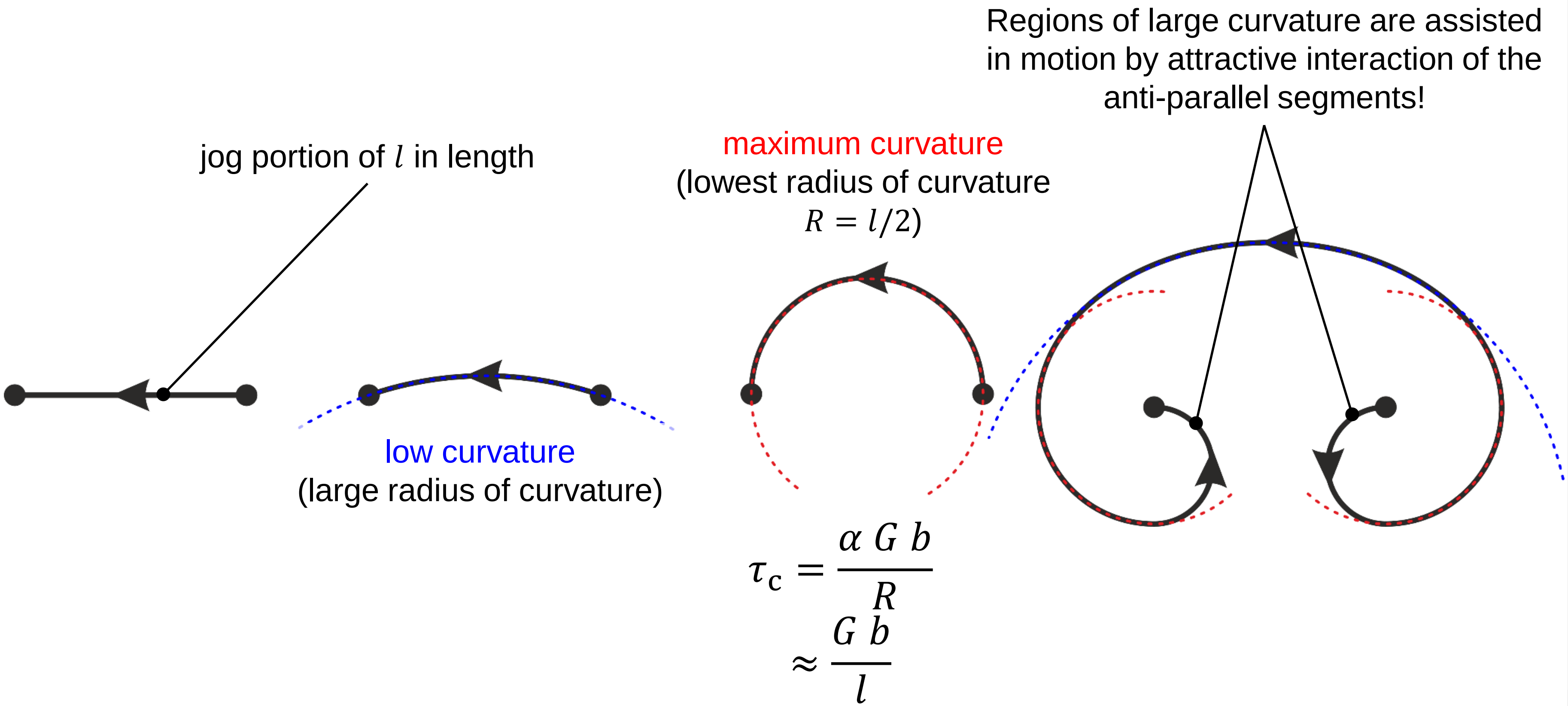
- By assuming **another pinning point**, the single-arm source converts into a **two-arm source**, well-known as the **Frank-Read source**.
- Due to the **rotation** about the pinning points, the **initially straight dislocation line transforms into an arc**.
- The **back stress** by the arcs increases **until the semi-circle configuration** is achieved with $\tau_c = \frac{G\,b}{l}$ **at** $R = \frac{l}{2}$. Every other configuration needs lower stress to be achieved.
- **By spiraling around**, **anti-parallel dislocation segments meet and annihilate**. A **new dislocation loop around the original dislocation segment** is formed. The **process continues** in the same way.
- Due to the large size of the loops, the loops expand under the applied stress.
- Any pile-up of the dislocation loops results in back stress and potential de-activation of the source.

Two-Arm Source: Frank-Read Source



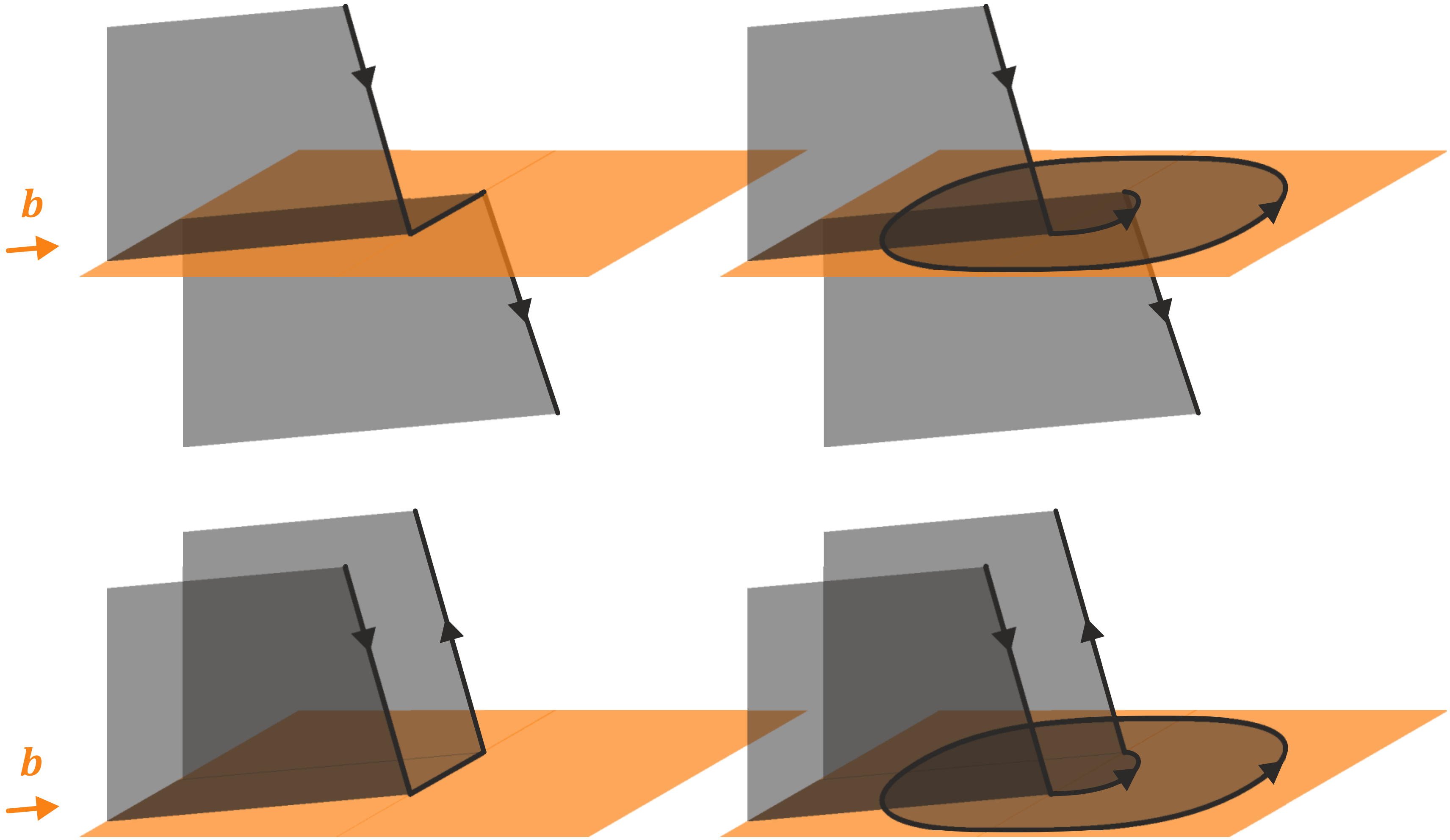
Two-arm source: various states.

Two-Arm Source: Frank-Read Source



Two-arm source: various states with tangent circles to visualize curvature.

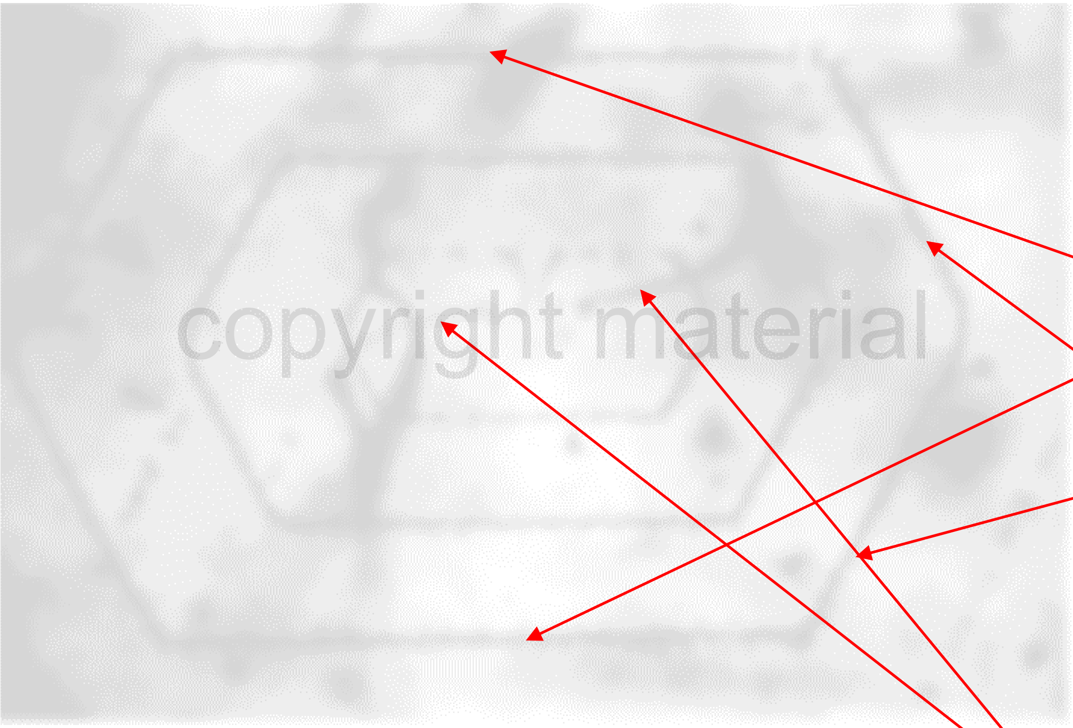
Two-Arm Source: Frank-Read Source



Slipped regions of the same: (upper part) Z shape, (lower part) U shape.

Two-Arm Source: Frank-Read Source

$$\boldsymbol{b} = \frac{1}{2}[1\bar{1}0]$$



(111)

screw segments

60°-segments along $[10\bar{1}]$ and $[01\bar{1}]$

Frank-Read source in Si. Decoration by Cu atoms.

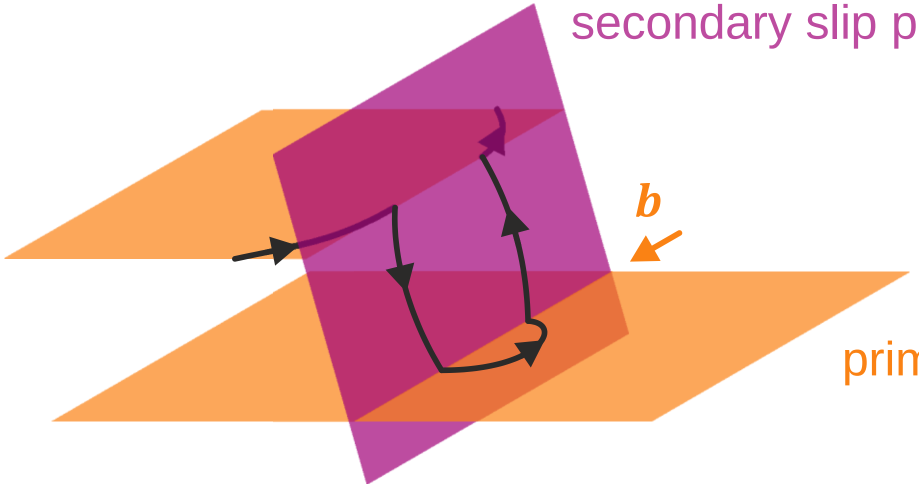
two segments not in the slip plane (and out-of-focus)

D. Hull, D. J. Bacon: “Introduction to Dislocations”, Amsterdam, etc.: Elsevier (2011)

Two-Arm Source: Frank-Read Source

- Assuming again stresses in the order of $\tau \approx \frac{G}{1000} \ldots \frac{G}{100}$, the segment enclosed by the pinning points must be of $l \approx 10^3 \ldots 10^2\ b$ in length.
- For the **Z shape**, the segment could be a result of intersection processes. It would be **a jog**. Since every intersection process only leads to an increase of $1b$, a total length of $10^3 \ldots 10^2\ b$ is only reasonable when another source on the same slip plane is already operating.
- The **U shape** might **originate from cross-slip** processes when the points of slip plane change are effeciently pinned (for example by dissociation). This process is conceivable.

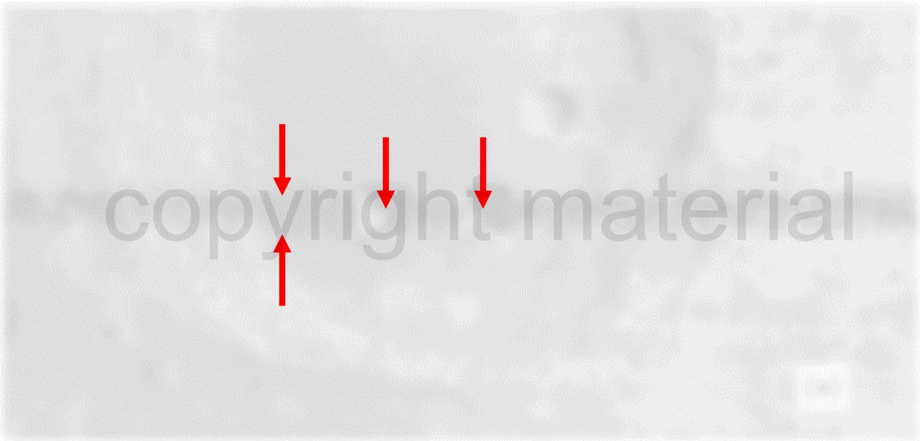
Two-Arm Source: Frank-Read Source



Two-arm source by cross-slip.

Significance in Experimental Results

- The increase of the width of deformation bands during plastic deformation indicates the significance of cross-slip in dislocation multiplication.
- The deformation bands gradually increase their thickness during plastic deformation by cross-slip.



Development of a single dislocation segment (large etch pits) to a deformation band (multiple small etch pits) by cross-slip and dislocation multiplication in LiF.

D. Hull, D. J. Bacon: “Introduction to Dislocations”, Amsterdam, etc.: Elsevier (2011)

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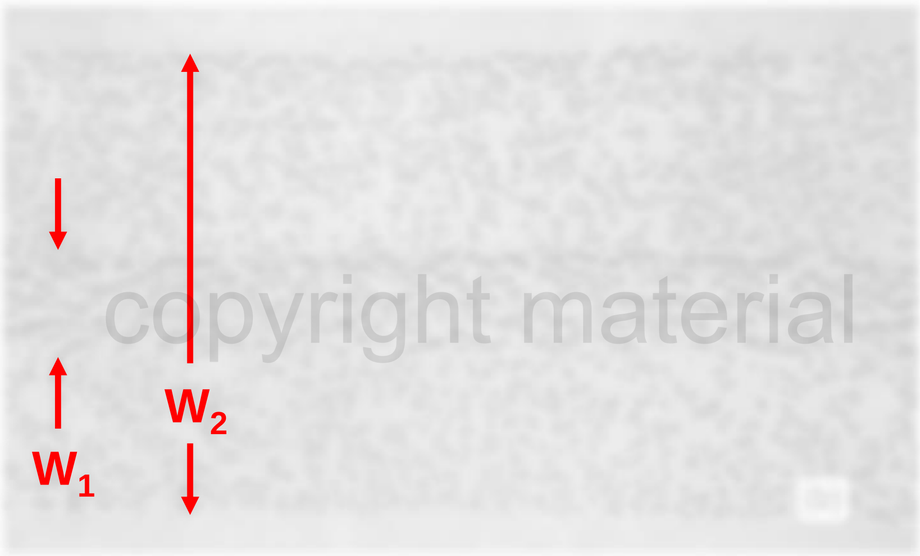
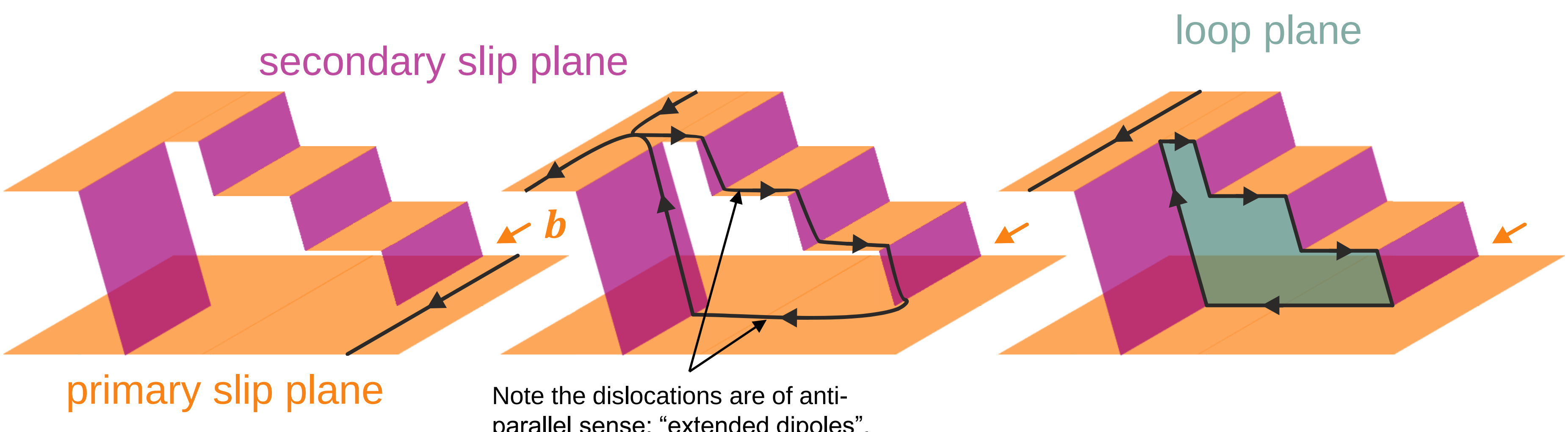


Development of deformation band thickness in LiF.

D. Hull, D. J. Bacon: “Introduction to Dislocations”, Amsterdam, etc.: Elsevier (2011)

Loop Formation by Cross Slip

- In materials with **high Peierls barriers and preferred screw character of dislocations** (A2 metals and intermetallics), **segment-wise cross-slip** can lead to the **formation of prismatic loops** (see Ch. 3a, only edge character along the dislocation loop).
- It is also observed in particle strengthened materials, since the induced cross-slip by the obstacle can lead to this configuration (one kind of bypassing the particle, Orowan mechanism).



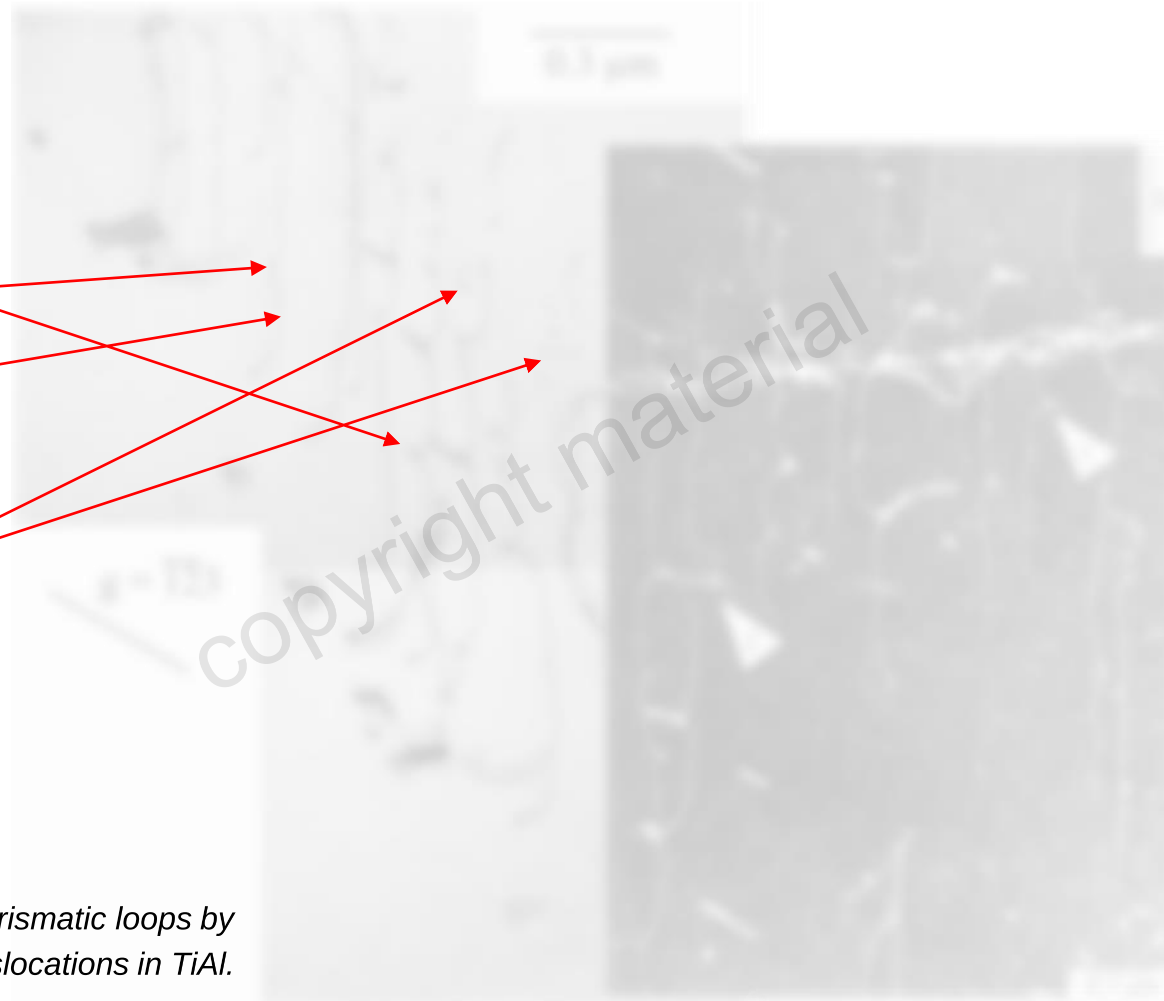
Formation of prismatic loops by segment-wise cross-slip.

Loop Formation by Cross Slip

extended dipoles

screw segments

first loops



Formation of extended dipoles and prismatic loops by cross-slip of $\frac{1}{2}\langle 110\rangle$ *screw dislocations in TiAl.*

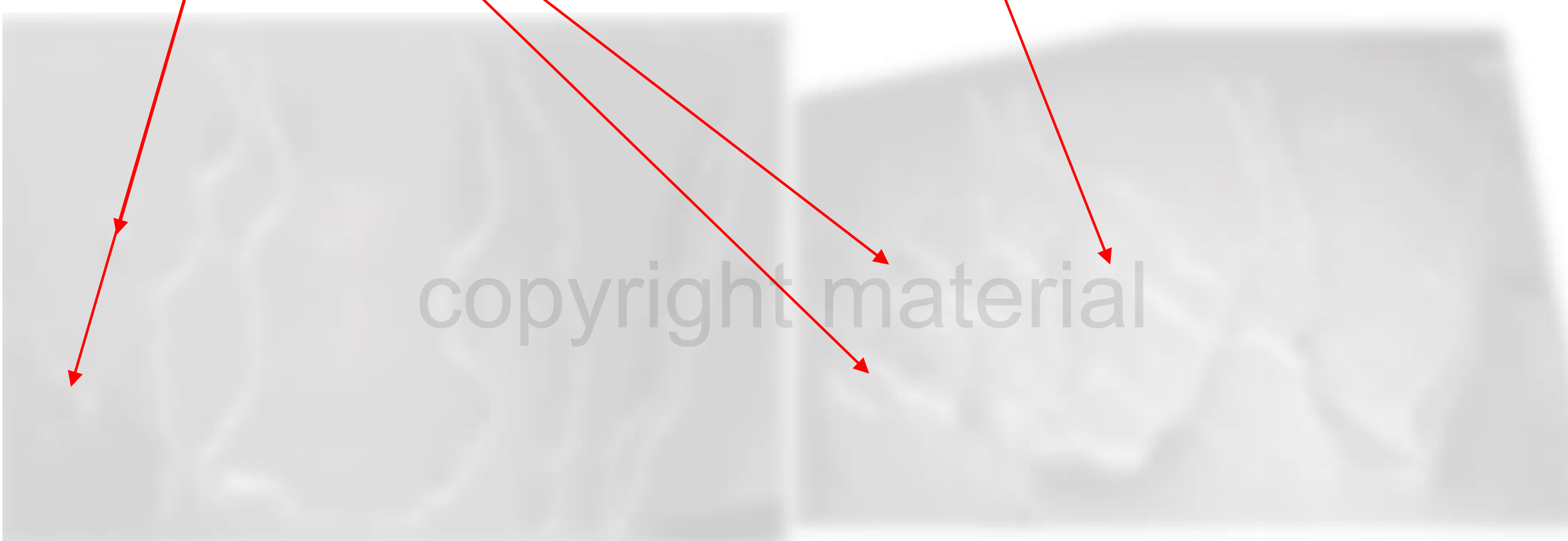
D. Caillard , M. Legros & A. Couret: “Extrinsic obstacles and loop formation in deformed metals and alloys” in Phil. Mag. 93 (2013) 203-221
F. Appel et al.: “Recent Progress in the Development of Gamma Titanium Aluminide Alloys” in Adv. Eng. Mat. 2 (2000) 699-720

Loop Formation by Cross Slip

same positions

loops

$a + c$ super dislocations on pyramidal planes of Ti_3Al during an in-situ deformation experiment: (left) initial state, (right) after plastic deformation.

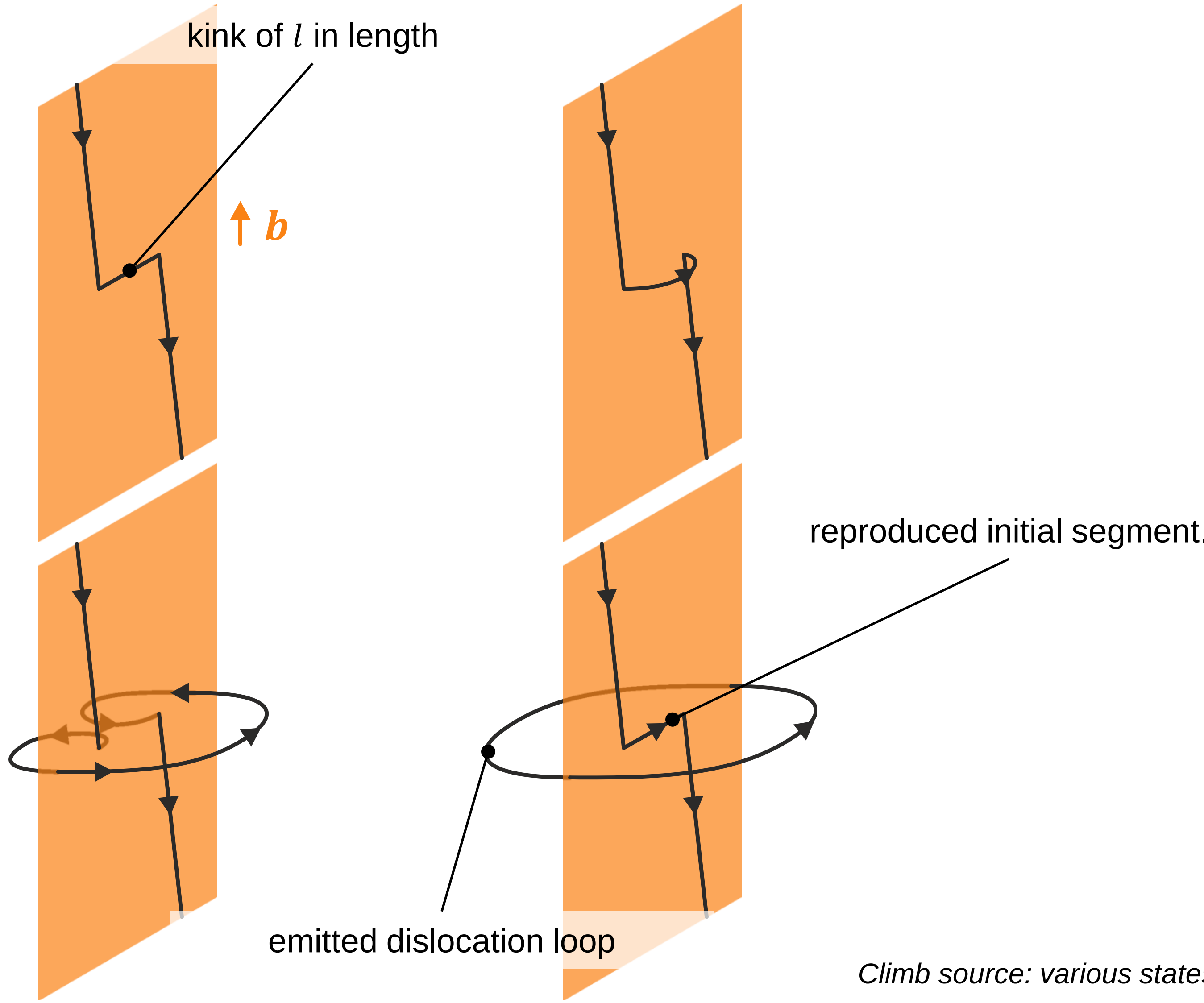


D. Caillard , M. Legros & A. Couret: “Extrinsic obstacles and loop formation in deformed metals and alloys” in Phil. Mag. 93 (2013) 203-221

Climb Source: Bardeen-Herring Source

- For the activation of climb sources, a **supersaturation in vacancies** is necessary. This can be achieved for example **by quenching from higher temperatures or irradiation**.
- Similar to what is observed for the Frank-Read source, the critical point is the semi-circle configuration $R = \frac{l}{2}$. The amount of supersaturation can, therefore, be estimated by $\ln(\frac{x}{x_0}) = \frac{G\, b\, \Omega}{l\, k_{\mathrm{B}}\, T}$ (see Ch. 3c for the chemical force on dislocation lines).
- Since we assume a climb process, the **Z shape** is not formed from a jog but from a **kink**.

Climb Source: Bardeen-Herring Source



Climb source: various states.

Climb Source: Bardeen-Herring Source

- Hence, $\Omega \approx b^3$ at $T = 300\ \mathrm{K}$ and with $G \approx 25\ \mathrm{GPa}$ as well as $b \approx 2.9\ \text{Å}$ lead to (Al):

$$l \approx \frac{150}{\ln\left(\frac{x}{x_0}\right)}\ b$$

- A supersaturation of only $\ln\left(\frac{x}{x_0}\right) = 25$ leads to a small kink of only $l \approx 6\ b$.
- Compare this with following quenching process
  - $x_0 = e^{\frac{\Delta S_\mathrm{V}^\mathrm{P}}{k_\mathrm{B}}} e^{-\frac{\Delta H_\mathrm{V}^\mathrm{F}}{k_\mathrm{B}\,T}}$ with $\Delta S_\mathrm{V}^\mathrm{P} = 2.4\ k_\mathrm{B}$, $\Delta H_\mathrm{V}^\mathrm{F} = 1.27\ \mathrm{eV}$ (again Al)
  - yields $x_0 = 5 \cdot 10^{-21}$ and $x = 2 \cdot 10^{-10}$ at $T = 300$ and $600\ \mathrm{K}$, respectively
  - **the supersaturation for quenching from 600 to $300\ \mathrm{K}$ is then**

$$\ln\left(\frac{x}{x_0}\right) \approx 25$$

Climb Source: Bardeen-Herring Source

- In many **intermetallic materials**, **anomalies of vacancies concentration** exist (note that there are thermal, constitutional, and structural vacancies possible)
- Due to the small excess vacancy concentration for activation of the climb source, climb sources are frequently found.

Bardeen-Herring source

growth of prismatic loops

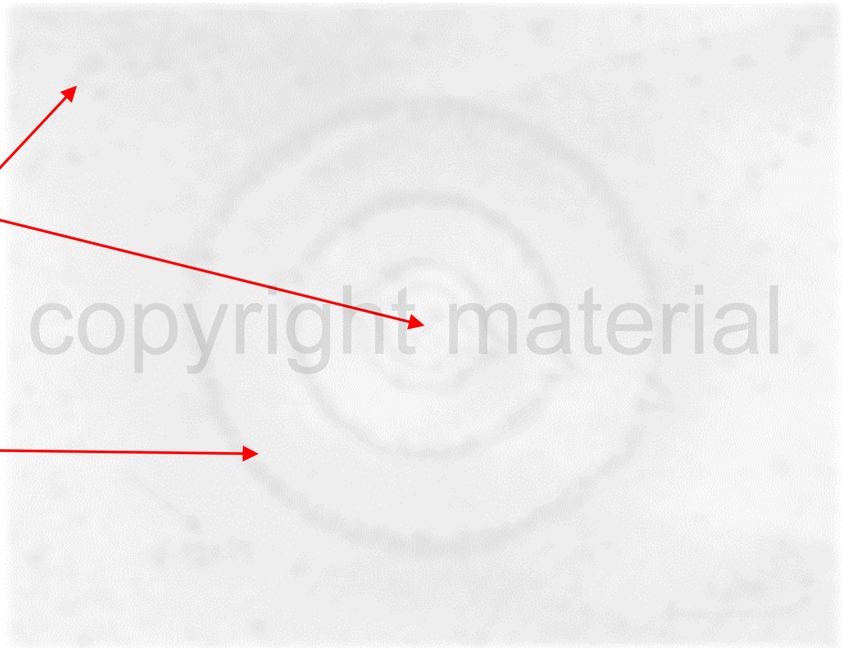
B-H source and growth of prismatic loops by vacancy agglomeration in TiAl at 820 K.

F. Appel et al.: “Recent Progress in the Development of Gamma Titanium Aluminide Alloys” in Adv. Eng. Mat. 2 (2000) 699-720

Climb Source: Bardeen-Herring Source

Bardeen-Herring source

many small prismatic loops



The vacancy agglomeration of the small loops retards the expansion of large loops. At the bulge, less small loops are formed due to cross-slip and, hence, the larger loops can expand further. Note that not all loops are visible for this diffraction condition!

Prismatic loops in an Al-4Cu single crystal subsequent to $793\ \mathrm{K}/2\ \mathrm{h}$*/quenching and ageing at* $266\ \mathrm{K}$ *with compression load along* $[011]$ *at* $30\ \mathrm{MPa}$*. Viewing direction is along* $[110]$*.*

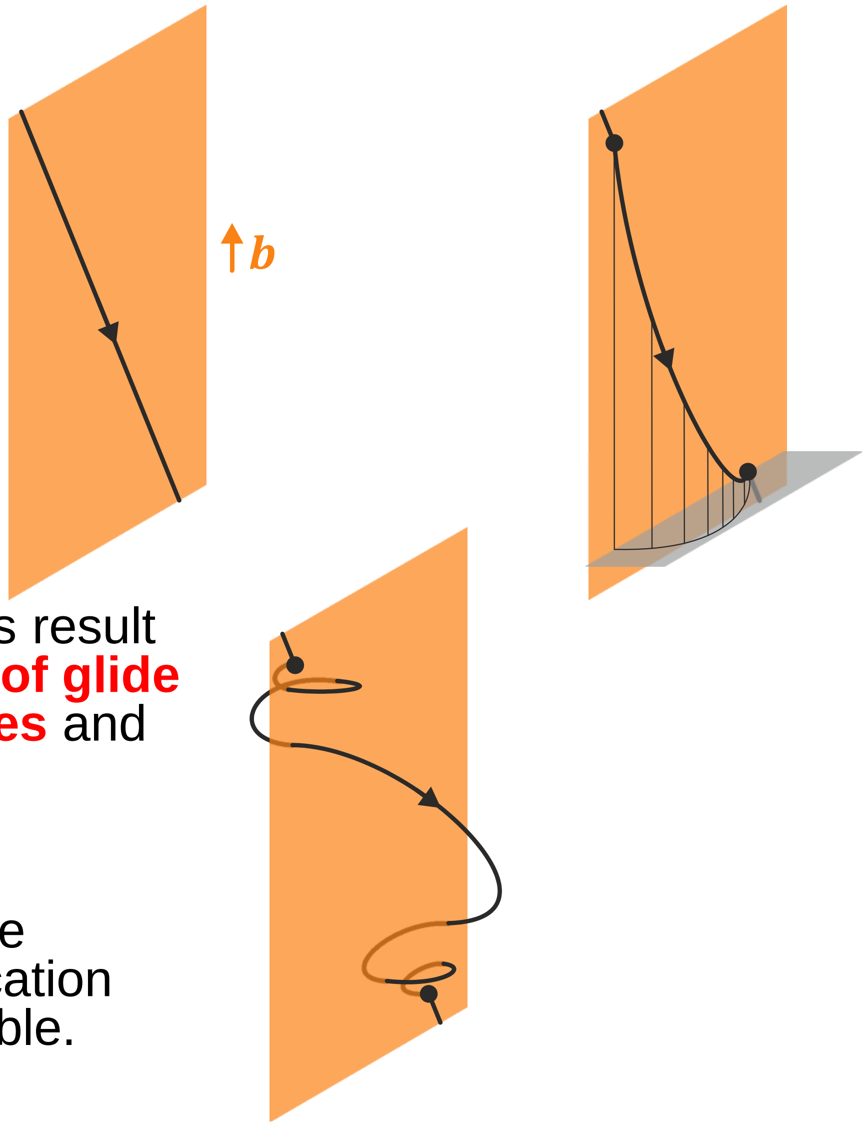
A. Sato , Y. Sugisaki & T. Mori: “Observation of dislocation sources in a quenched and stress-aged Al-4 wt% Cu alloy single crystal” in Phil. Mag. 51 (1985) 133-143

Climb Source: Bardeen-Herring Source

Prismatic loops in Al-3.5Mg subsequent to quenching from 550 °C.

D. Hull, D. J. Bacon: “Introduction to Dislocations”, Amsterdam, etc.: Elsevier (2011)

Helical Source



- Mixed dislocations result in **combinations of glide and climb sources** and the **formation of spirals/helices**.
- Under certain circumstances, the emission of dislocation loops is still possible.

Helical source: various states.

Helical Source

helix

Formation of the helix during in-situ heating in the TEM after deformation at room temperature.

F. Appel et al.: “Recent Progress in the Development of Gamma Titanium Aluminide Alloys” in Adv. Eng. Mat. 2 (2000) 699-720

Other Sources

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- **Heterogonous nucleation at interfaces** is also an important contribution to dislocation multiplication.
- Following issues during assessment have to be considered:
  - Interfaces are usually points of stress concentrations. Hence, stresses might be significantly higher than expected from external load.
  - High interface density leads to high nucleation rates. In contrast very narrow interface spacing can restrict the free motion of the arms of dislocation sources and block the dislocation multiplication.

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Summary

- **Dislocation do not form spontaneously.**
- **Dislocations nucleate at surfaces and interfaces** in heterogonous materials.
- Dislocations sources can be grouped by the major contribution to dislocation motion: **glide, cross-slip and climb**. In general, **single-arm** and **two-arm** sources are possible.