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Final Proceedings of
The EOARD/IRC-sponsored
International Workshop on Gamma
Aluminide Alloy Technology

held from 1 to 3 May 1996
at The IRC in Materials for High Performance
Applications
The University of Birmingham

SECTION ONE

The organisers wish to thank the United States Air Force European Office of Aerospace Research and Development for its contributions to the success of this conference

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3. Presentation Material:

F Appel

P Bowen

L Christodoulou

C Dimitrov

A Dowson

Y-W Kim

C Lang

M H Loretto

S Naka

P Threadgill

Foreword/Summary

The Workshop was organised jointly by Dr Young-Won Kim (UES, USA) and Professor Paul Bowen (University of Birmingham, UK), and was attended by fifty-four delegates. These included international representatives from the USA, France, Germany and the UK. In addition, two representatives attended on behalf of EOARD (ONR). The attendance list was, of course, dominated by UK delegates, but invited technical contributions were split as follows: seven sessions from the USA; four sessions from the UK; three sessions from France and two sessions from Germany.

The Workshop was built primarily around major contributions from Dr Young-Won Kim, and from individual speakers representing European countries with significant interests in the development of gamma-based titanium aluminide alloys. It thus afforded a unique opportunity to assess the relative progress of such alloys in the USA and Europe. The Workshop appeared to achieve its stated primary objective of providing focused academic debate on issues of fundamental understanding, and on how these alloys are likely to be introduced into the market place. The Workshop was both stimulating and informative, and favourable feedback from delegates was received by the organisers. Most important, it appears that the future of gamma based alloys is secure for selected aerospace, power generation and automotive applications.

Enclosed for reference are copies of the overheads used in presentations by invited contributors.

P Bowen
April 1997

Joint EOARD/IRC-sponsored International Workshop on Gamma Aluminide Alloy Technology

Wednesday 1 May to Friday 3 May 1996

This will be a short, intensive three-day workshop on fundamental and technological aspects of gamma based titanium aluminide alloys. It is intended that the conference will have twelve formal invited papers: approximately six from the USA and six from Europe. In addition, shorter focused presentations from delegates will also be encouraged. Attendance will be by invitation only from industrial companies, academic institutions and research organisations. It is anticipated that the audience will be limited to fifty delegates in order to facilitate useful discussion. The programme will allow ample time for discussion groups to meet on an informal basis, and there will be a focused summary discussion on the third day of the conference.

Topics to be included are:

- Fundamentals of behaviour
- Processing
- Microstructural development and control
- Microstructure - property relationships
- Damage tolerance and life prediction
- Alloy development design
- Property improvements
- Component-specific alloy design
- Joining
- Applications
- Future R and D directions

Confirmed speakers include Dr Young-Won Kim, UES Inc, USA; Professor P Bowen, The University of Birmingham, UK; Dr F Appel, GKSS, Germany; Dr S Naka, ONERA, France; Professor M H Loretto, The University of Birmingham, UK. Formal contributions are also expected from several industrial and other research organisations.

Formal proceedings will not be published, but lecture notes and handouts will be made available to delegates. In addition, a summary document will be prepared following the conference and this will be distributed to all delegates. There will be no registration fee. The central aim of the workshop will be to encourage focused debate between academics and industrialists with a view to expediting the introduction of gamma based titanium aluminides into the market place. It is also anticipated that the workshop will contribute to underlying issues of fundamental understanding still required for this emerging class of engineering structural alloys.

We wish to thank the United States Air Force European Office of Aerospace Research and Development for its contribution to the success of this conference.

Joint EOARD/IRC International Workshop
on
Gamma Aluminide Alloy Technology

Wednesday 1 May to Friday 3 May 1996

REGISTRATION FORM (Photocopies/facsimile copies are acceptable)

Name: ....

Organisation: .... **Address**: ....

....

Telephone No: .... **Fax No**: ....

Short contribution requested (yes/no): ....

Title of contribution (if applicable): ....

....

Accommodation

(A limited number of rooms (twenty) are available at our Conference Park on a "first-come-first-served" basis at an approximate cost of £44.50 per night (single room) or £60.00 per night (double room), including breakfast. Preference will be given to international delegates, provided that requests are received by 31 March 1996.)

Please reserve accommodation at (*Please tick appropriate boxes, if you would like us to reserve accommodation for you*):

		30 April	1 May	2 May
i)	**The University Conference Park**	☐	☐	☐
	single/double..................................			
ii)	**A local hotel**	☐	☐	☐
	single/double..................................			
	(*please indicate price limit, if applicable*)			

Meals

Lunch and dinner will be provided for a total cost of £20.00 per delegate. (*Please tick box if you would like to take advantage of this offer.*) ☐

Please return this form to Professor P Bowen, IRC/School of Metallurgy and Materials, The University of Birmingham, Edgbaston, Birmingham B15 2TT, UK (Fax: 0121 414 5232)

Gamma Workshop

Wednesday 1 May to Friday 3 May 1996

PROGRAMME

Wednesday 1 May

10.30 Registration and Coffee

10.50 Welcome and Introduction to the Workshop

11.00 **Session One - Chairman: P Bowen**

- 11.00 Fundamentals of behaviour - Y W Kim
- 12.00 Processing of gamma based aluminides - Y W Kim

13.00 Lunch

14.30 **Session Two - Chairman: T Khan**

- 14.30 Alloy development and microstructural behaviour - M H Loretto
- 15.15 Recent activities and future directions in the study of microstructures of gamma titanium aluminides - S Naka

16.15 Tea

16.45 **Session Three - Chairman: S Naka**

- 16.45 Microstructural development and control - Y W Kim
- 17.45 Microstructual development in gamma TiAl alloys containing dispersoids of Ti B_2 - L Christodoulou
- 18.30 Discussion

Thursday 2 May

10.00 **Session One - Chairman: I P Jones**

- 10.00 Segregation in cast alloys - A Dowson
- 10.45 Coffee
- 11.15 Structural instabilities in the TiAl alloys with B2 structure: first principle approach - Dr Nguyen-Manh
- 11.45 Microstructure-property relationships - Y W Kim

13.00 Lunch

14.15 **Session Two - Chairman: Y W Kim**

- 14.15 Overview of joining of gamma alloys - P Threadgill
- 15.15 Mechanical behaviour of extruded gamma alloys - M Winstone

16.00 Tea

16.30 **Session Three - Chairman: J Petit**

- 16.30 Fatigue crack propagation in gamma aluminide alloys - C Mabru
- 16.50 Fatigue crack growth behaviour in titanium alloys and titanium aluminides - S Listerin
- 17.10 Design against fracture and fatigue failure in gamma based alloys - P Bowen
- 18.00 Discussion

Friday 3 May

9.15 **Session One - Chairman: Y W Kim**

- 9.15 High temperature deformation mechanisms in solution and and precipitation hardened two-phase titanium aluminide alloys - F Appel
- 10.00 Preliminary results on point defects, atomic mobility and creep in model TiAl compounds - C Dimitrov
- 10.25 The role of the initial steps of oxidation for high temperature oxidation resistance - C Lang
- 10.50 Discussion

11.10 Coffee

11.30 **Session Two - Chairman: M H Loretto**

- 11.30 Industrial applications, component specific design, and future research and development directions - Y W Kim

13.00 Lunch

14.30 General discussion and close of conference

High - Temperature Deformation Mechanisms in Solution and Precipitation Hardened Two-Phase Titanium Aluminides

F. Appel, U. Christoph, M. Oehring, and R. Wagner

Institute for Materials Research
GKSS-Forschungszentrum Geesthacht
D-21502 Geesthacht

Birmingham, May 1996

High-temperature applications of titanium aluminides

- design requirements at intended service temperatures of about 700 °C:
  - high strength
    reasonable toughness
  - microstructural stability
  - creep resistance
    corrosion resistance

- problems at elevated temperatures:
  - degradation of strength properties,
  - rate dependend deformation processes become important

$$\dot{\varepsilon} = f \cdot \rho_m v_d = f \rho_m v_0 \exp - \Delta G/kT$$

present study

- mobilities and multiplication of dislocations
- structural stability
- metallurgical factors affecting high-temperature strength

Alloy compositions and microstructure

Composition (at.%)	Microstructure
Ti-48Al-2Cr	nearly-lamellar
Ti-47Al-2Cr-02Si	near gamma nearly-lamellar
Ti-47Al-1Cr + + Nb, Mn, Si, B	fully-lamellar
Ti-49Al + + (60 - 1200) wt.ppm C	duplex and nearly lamellar, depending on C-conc

concentration of interstitial elements in the starting material

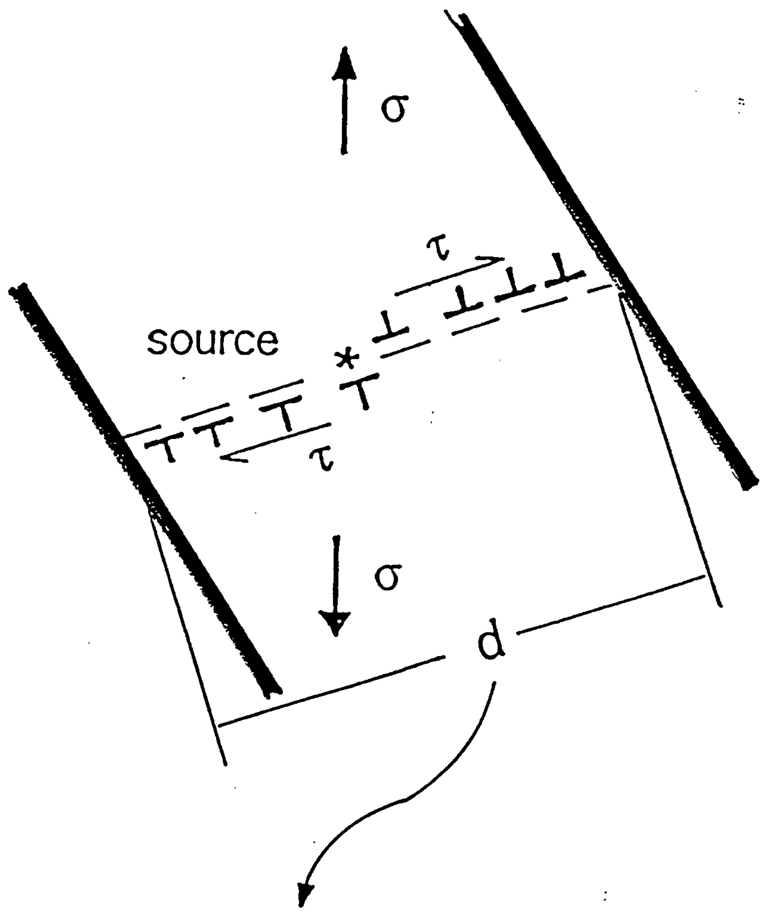
N_2 : 100 - 200 wt. ppm
0_2 : 500 - 700 wt. ppm
C : 100 - 200 wt. ppm

- thermal treatments for solution and precipitation of carbon and nitrogen

Strength properties of titanium aluminides

- Structure/property relationships usually described by Hall-Petch equations

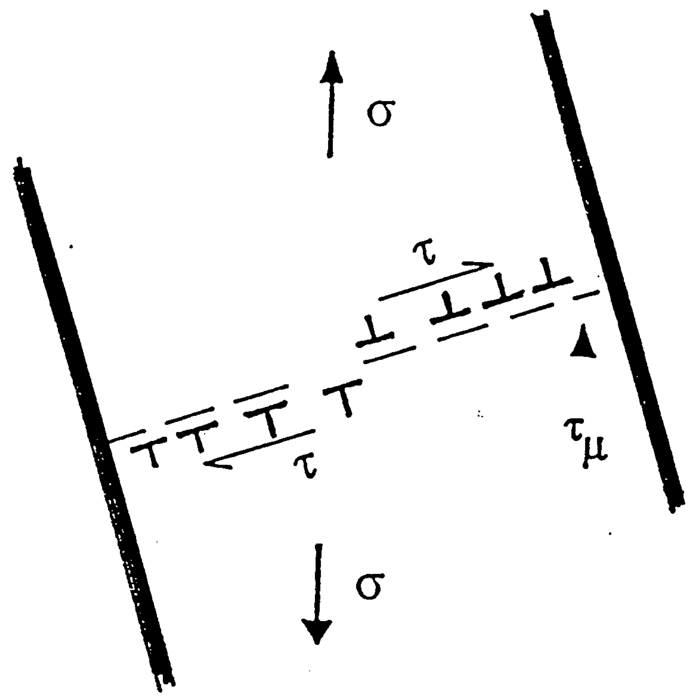
- model bases on dislocation pileup theories

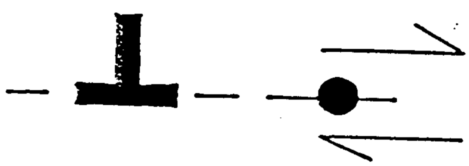


$$\sigma_{HP} = \sigma_o + K_y d^{-1/2} \quad , K_y = 0.9...1.5\,\text{MPa m}^{1/2}$$

- athermal stress contribution that is independent of temperature and strain rate

Factors governing the dislocation velocity





$\tau^* = \tau - \tau_\mu$

- friction forces due to
  - localized obstacles
  - lattice resistance
  - jog dragging
  - dislocation climb, etc.

- dislocation velocity

$$v_D = v_o \exp\left[- \Delta G (\tau^*) / kT\right]$$

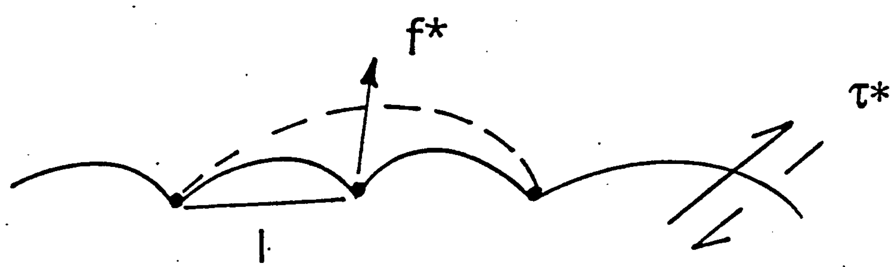
- strain rate

$$\dot{a} = \rho_D\, b\, v_D = \dot{a}_o \exp\left[-\Delta G (\tau^*) / kT\right]$$

Thermal stress parts

- Overcoming of glide obstacles with the aid of thermal activation

model



- dislocation velocity

$v_D = v_o \exp [- \Delta G (\tau^*) / kT]$

- strain rate

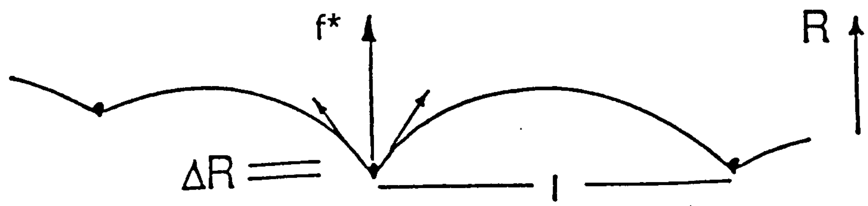
$\dot{a} = \rho_D \, b \, v_D = \dot{a}_o \exp [-\Delta G (\tau^*) / kT]$

⬇

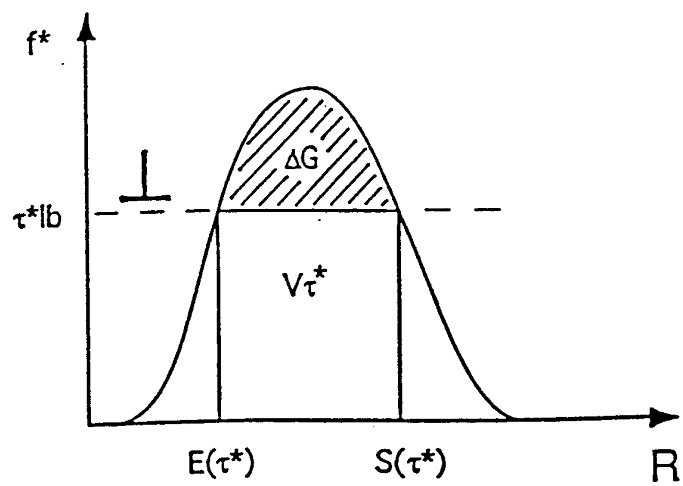
τ^* described in terms of activation parameters

Dislocation mobilities

- dislocation velocity: $v_d = v_0 \exp - \Delta G(\tau^*)/kT$, described in terms of activation parameters



Energy profile characterizes obstacle strength



- activation parameters:

$V = l\, b\, \Delta R$
$\Delta F^* = \Delta G + V\tau^*$
$\tau^* = (1/V)\ (\Delta F^* + k\, T \ln \dot{a}/\dot{a}_0)$

- Identification of relatively small and weak glide obstacles (solute atoms, jogs etc.)

Experimental methods

- Changes of the mobile dislocation density and of the obstacle structure should be avoided

incremental changes of strain rate and temperature:

- $(\Delta\sigma/\Delta \ln \dot{\varepsilon})_T$
  strain rate cycling tests

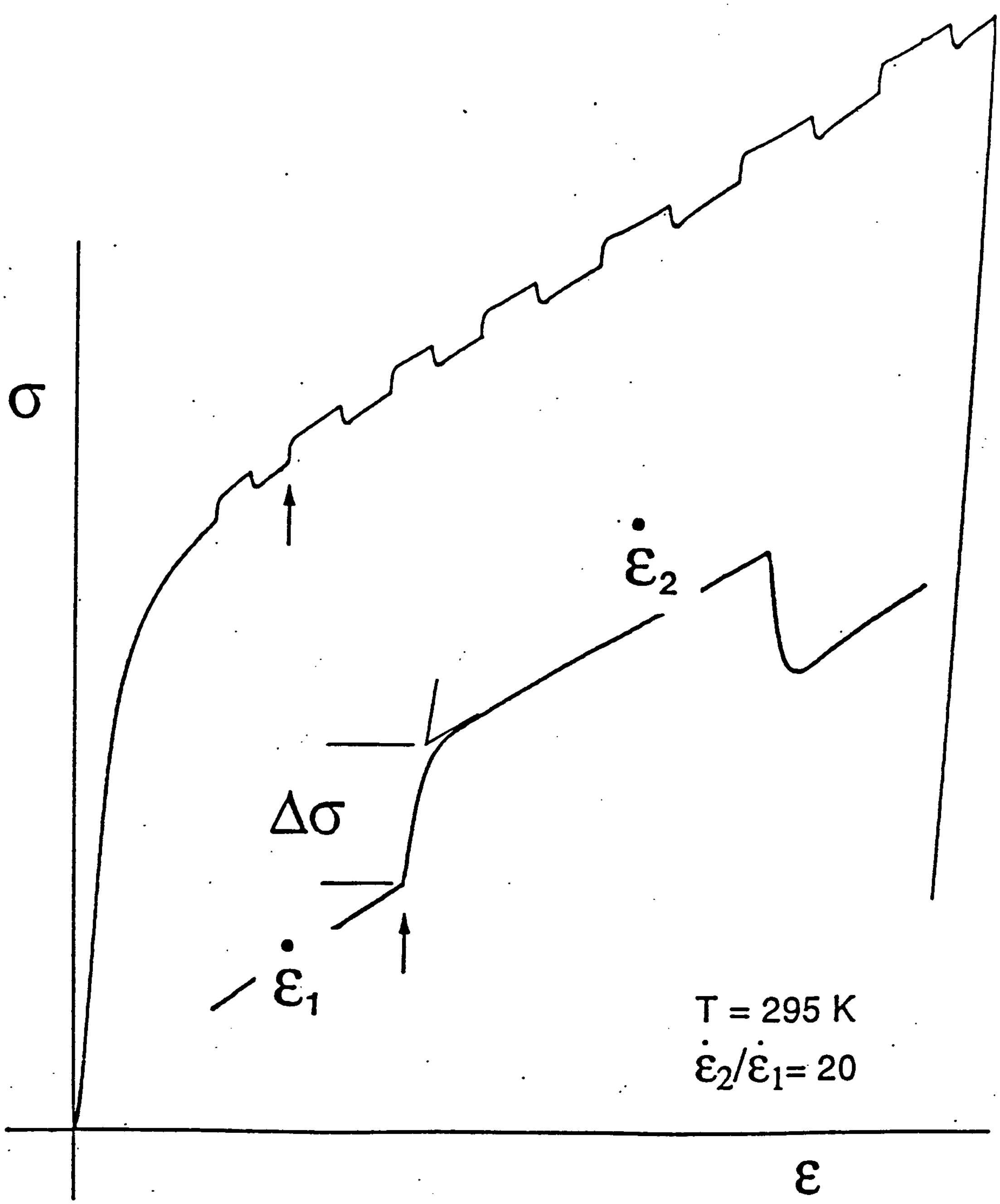
- $\ln(-\dot{\sigma}) = f(\sigma)$
  stress relaxation test

- $(\Delta\sigma/\Delta T)_{\dot{\varepsilon}}$
  temperature cycling tests

determination of these parameters as function of σ, T and ε

example: load elongation trace of a strain rate cycling test performed on $(\alpha_2 + \gamma)$ TiAl

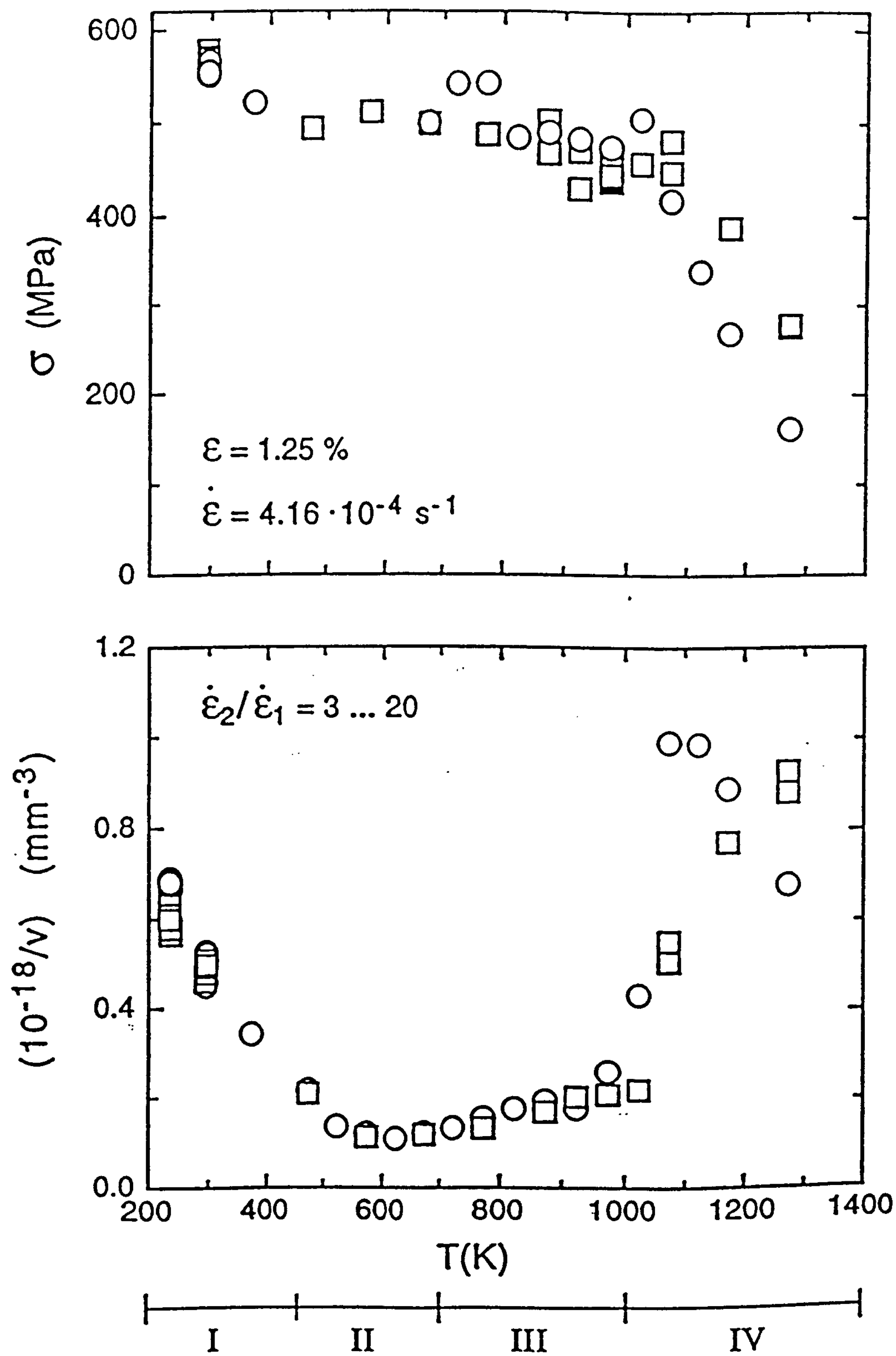
Estimation of activation parameters



Load elongation trace of a strain rate cycling test performed on an $(\alpha_2 + \gamma)$ TiAl alloy

Flow stresses and activation volumes

Microstructures: O near gamma ☐ nearly-lamellar



Interpretation region I

Athermal stress parts

$$\sigma_\mu = \sigma_{dis} + \sigma_{HP}$$

- σ_{dis}: long-range interaction of dislocations

$$\sigma_{dis} = f\,\alpha\,\mu\,b\,\rho^{1/2} = 30 \text{ MPa},$$

$$\rho = 10^8 \text{ cm}^{-2},\ f = 3,\ \alpha = 0.5$$

- σ_{HP}: interaction of dislocations with grain boundaries and lamellar interfaces

$$\sigma_{HP} = K_y\, d^{-1/2} = 400 \text{ MPa},$$

$$d = 11.4\ \mu\text{m},\ K_y = 1.35 \text{ MPa m}^{1/2}$$

- athermal stress part arises mainly from interactions of dislocations with grain boundaries and lamellar interfaces

Activation parameters

near gamma microstructure

$T = 300\ K$, $\varepsilon = 1.25\%$, $\dot{\varepsilon} = 4.16 * 10^{-4}\ s^{-1}$

$\sigma\ = 550\ MPa$

$\sigma_{\mu} = 430\ MPa$

$\tau^* \quad = 40\ MPa$

$V \quad = 95\ b^3$, $\quad b = 1/2 <110]$

$V\tau^* \quad = 0.5\ eV$

$\Delta G \quad = 0.8\ eV$

$\Delta F^* \quad = \Delta G + V\tau^* = 1.3\ eV$

▶ low dislocation mobility

⬇

TEM-observations

Pinning of 1/2 <110] screw dislocations by localized obstacles and jogs. (CM 3938)

Conclusions

- The estimated activation parameters

  $\tau^* = 40$ MPa, $V = 95\ b^3$, $\Delta G = 0.8$ eV, $\Delta F^* = 1.3$ eV

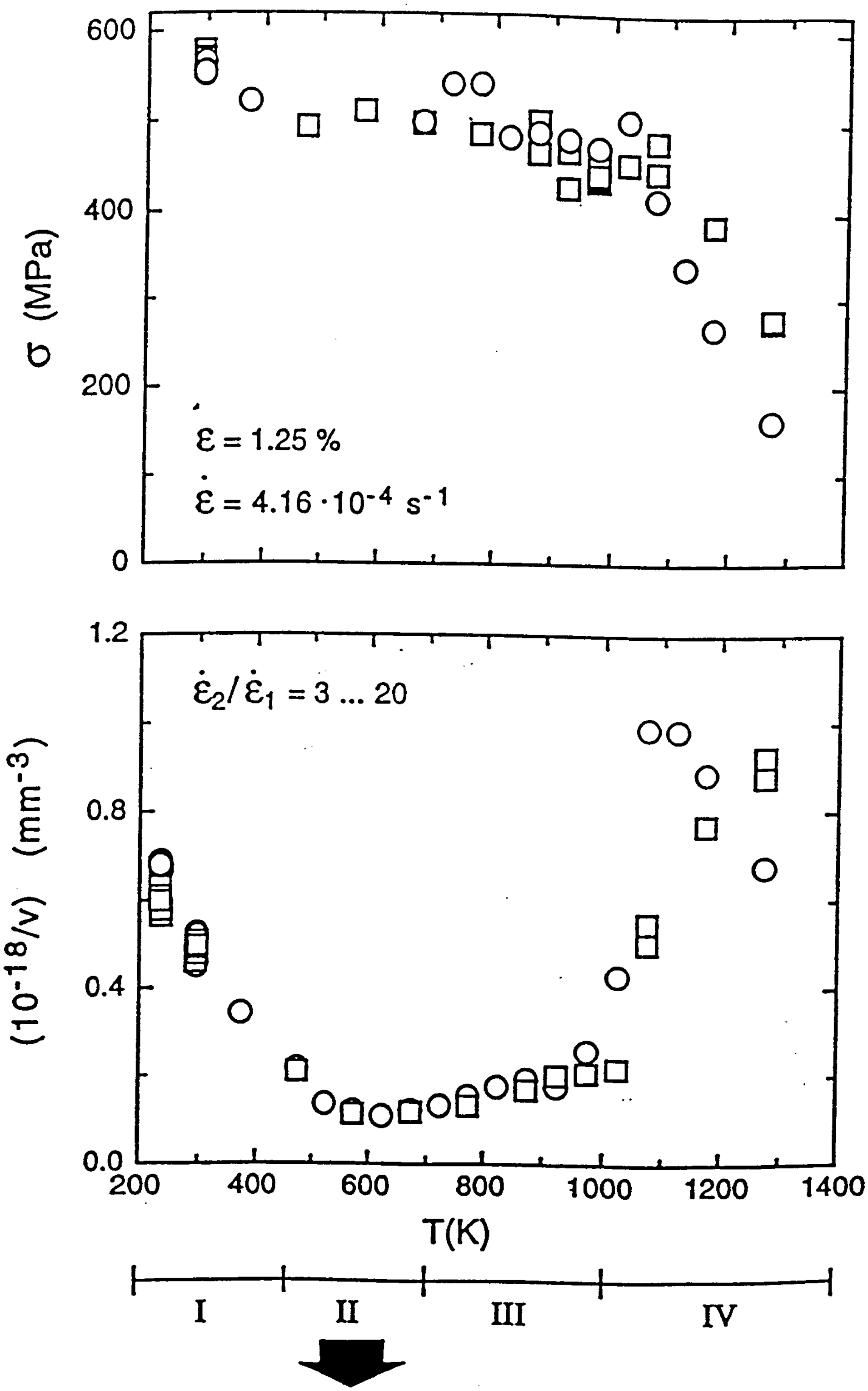
  suggest a relatively low dislocation mobility at room temperature.

- The mobility of ordinary dislocations is controlled by the combined operation of localized pinning and lattice friction.

- A thermal stress part contributes with 20% to the total stress at room temperature.

Flow stresses and activation volumes

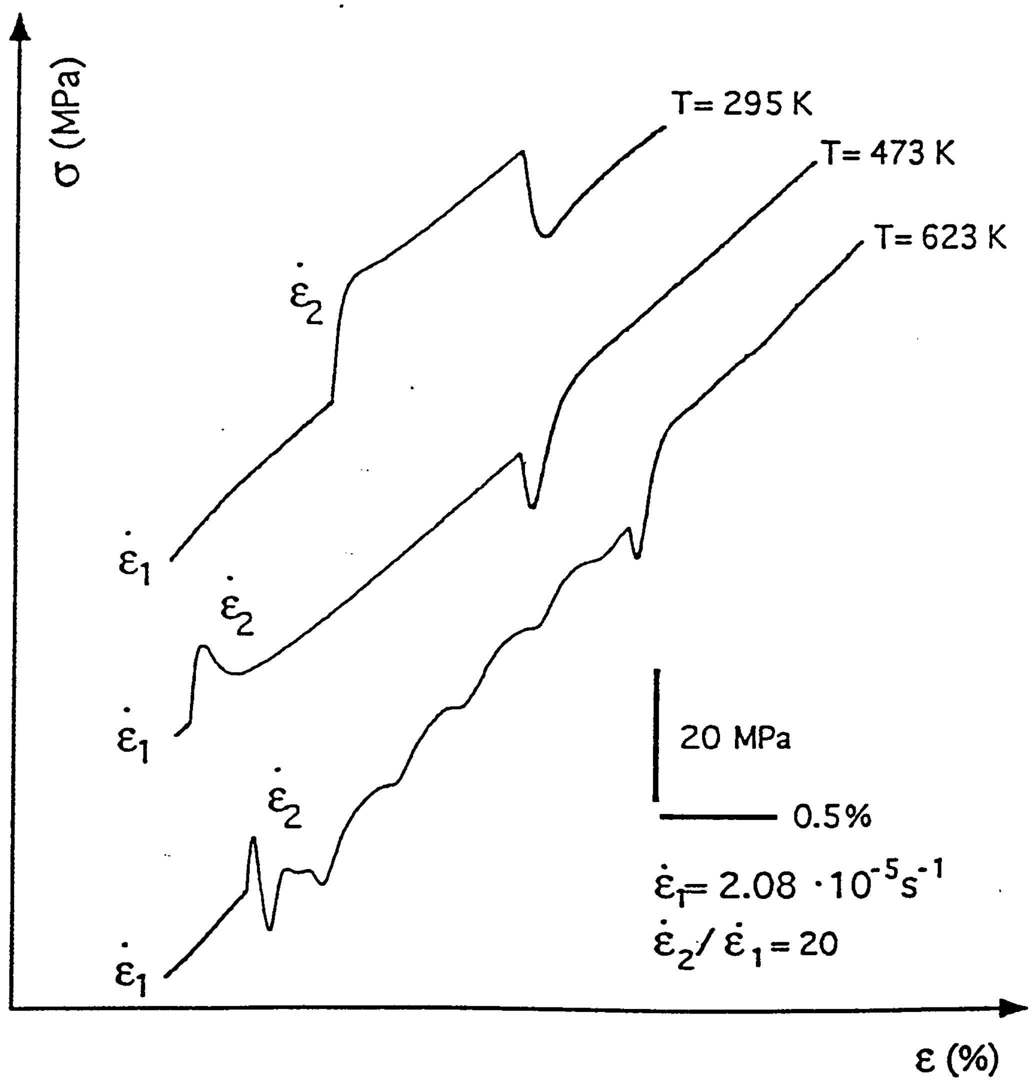
Microstructures: O near-gamma □ nearly lamellar



Temperature dependence of the flow stress

Region II: T = 450 ...700 K
Yield drop effects and serrations in a narrow temperature interval,

dependence on strain rate → load elongation traces

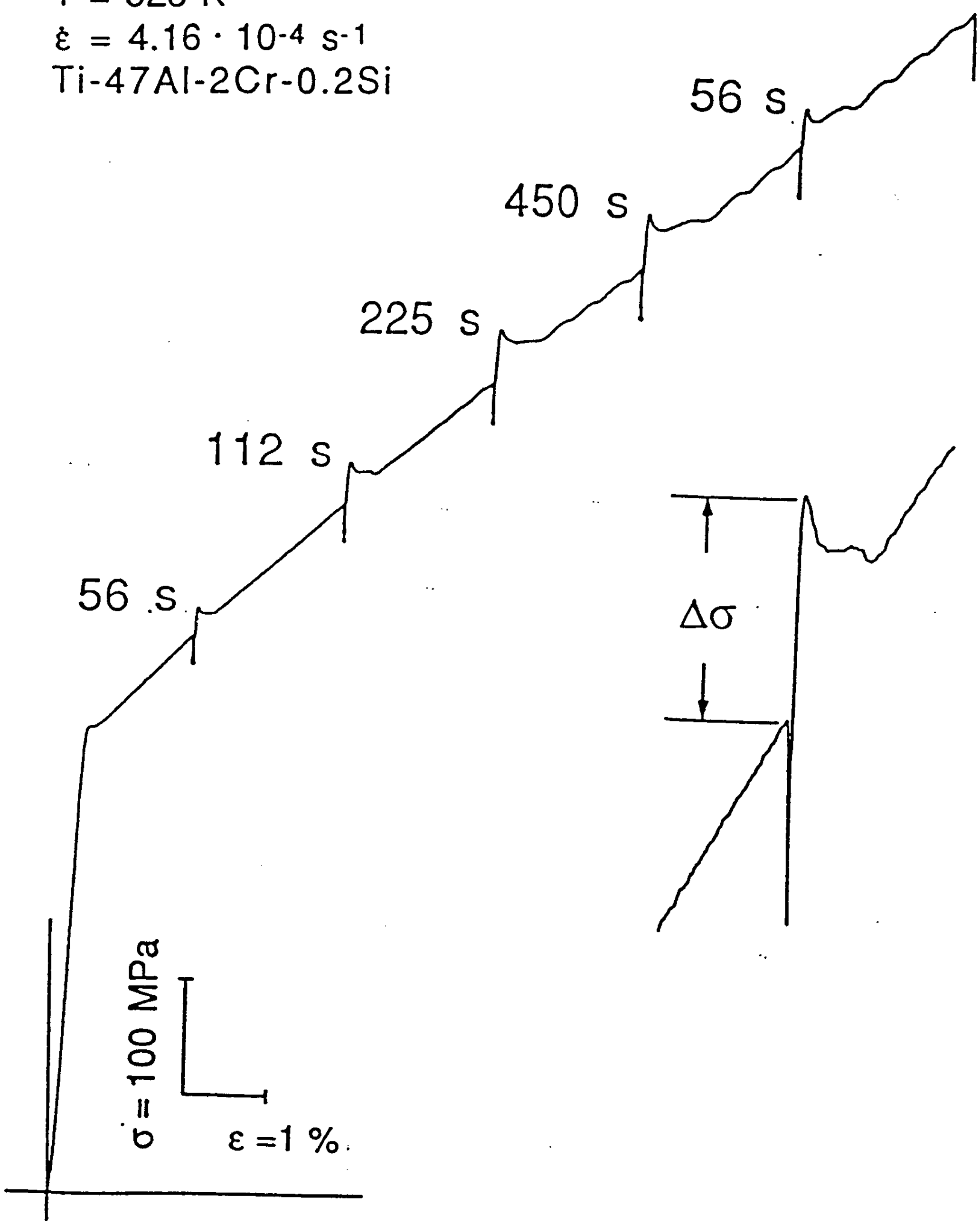


formation of impurity atmospheres,
further investigations

Static strain ageing

load - elongation trace

T = 523 K
$\dot{\varepsilon} = 4.16 \cdot 10^{-4}\ s^{-1}$
Ti-47Al-2Cr-0.2Si



Static strain ageing

experimental investigations

$T = 300, 423, 523, 623$ K

$\dot{\varepsilon} = 4.16 \cdot 10^{-4} s^{-1}$

$\Delta\sigma = f\ (T,\ t_a,\ \varepsilon,\ \sigma_a,\ c_i)$

$\Delta\sigma = g\ (\sigma_a)$

$\Delta\sigma = h\ (t_a)$

$\Delta\sigma = u\ (\varepsilon)$

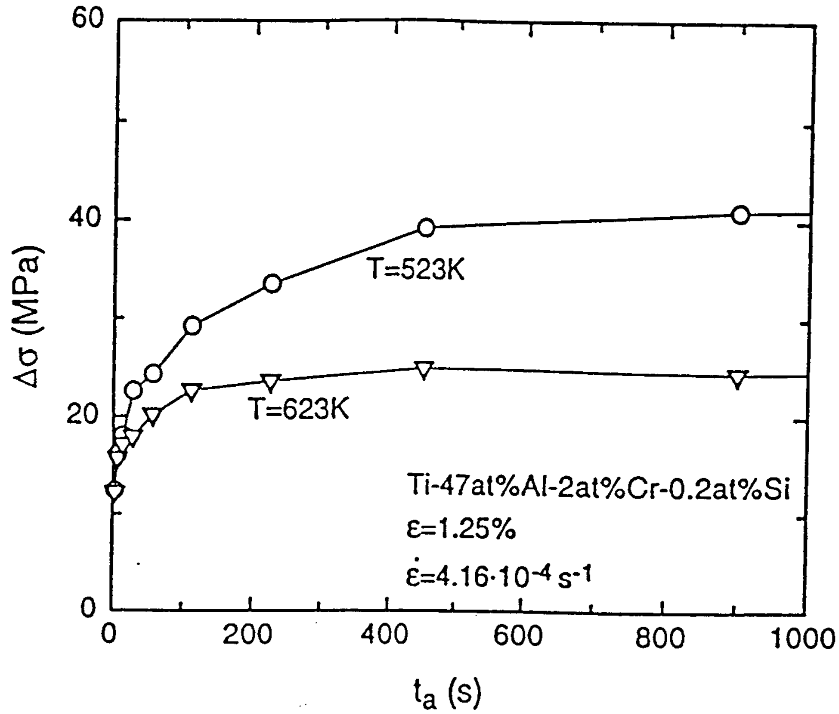
Ti-49 Al + 1200 wt. ppm C

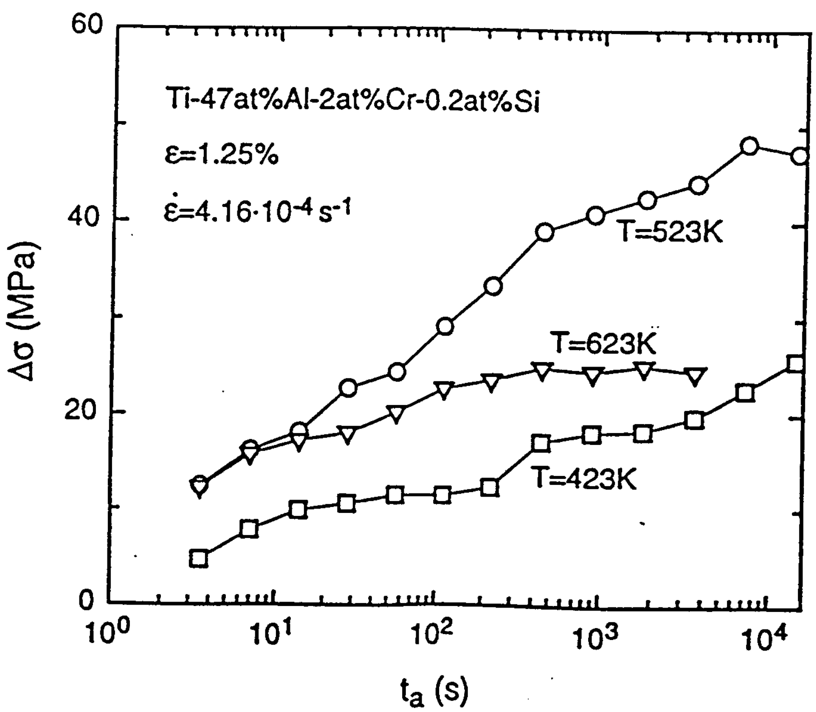
comparison with
Ti-47Al-2Cr-0.2Si

Static strain ageing

stress increments $\Delta\sigma$ due to strain ageing

- dependence on ageing time t_a

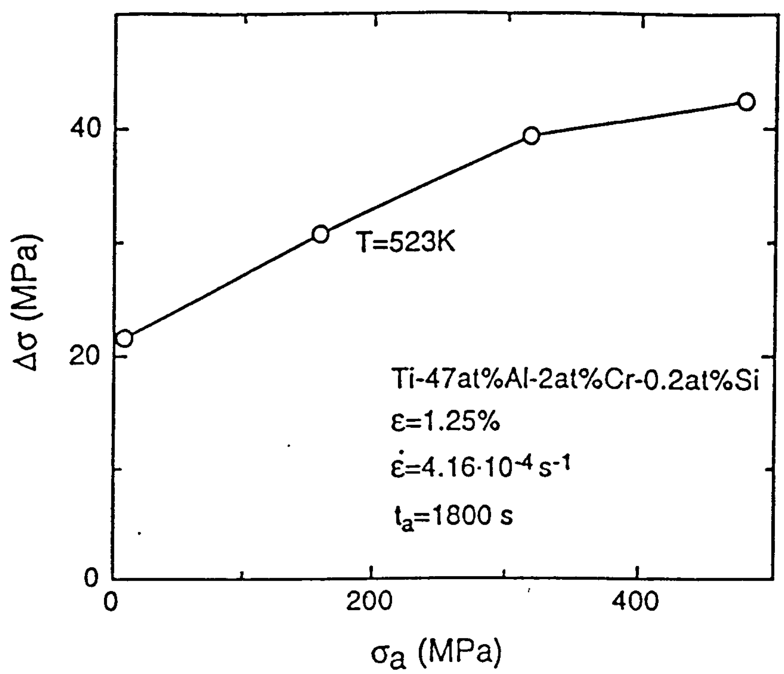




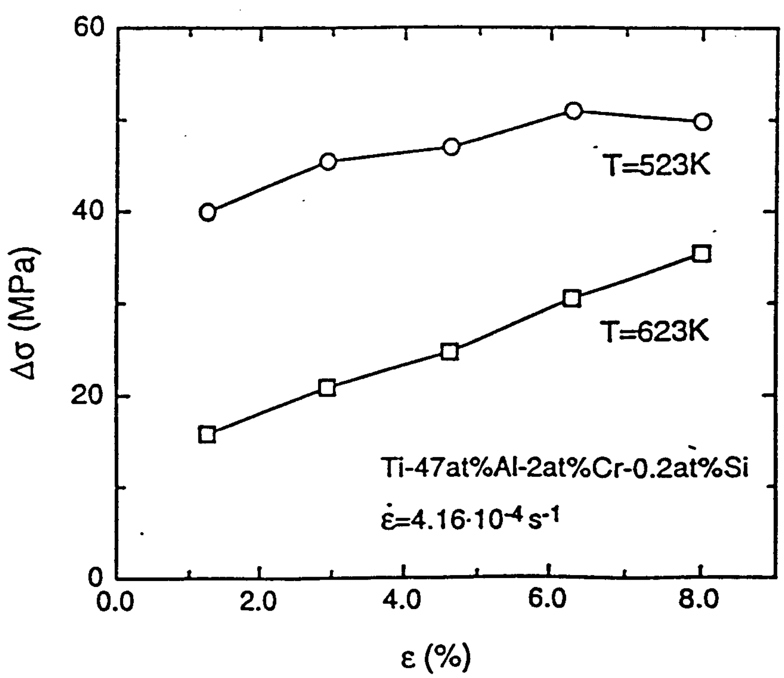
Static strain ageing

stress increments $\Delta\sigma$ due to strain ageing:

- dependence on ageing stress σ_a



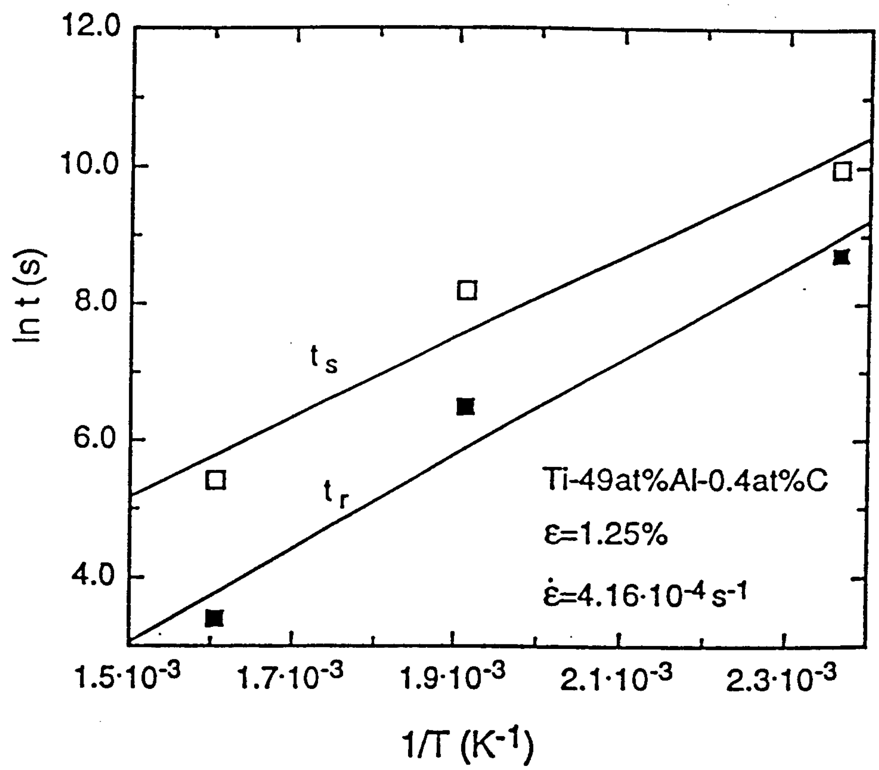
- dependence on strain ε



Static strain ageing

- evaluation of the kinetics $\Delta\sigma$ (t_a):

saturation values $\Delta\sigma_S = f(T)$



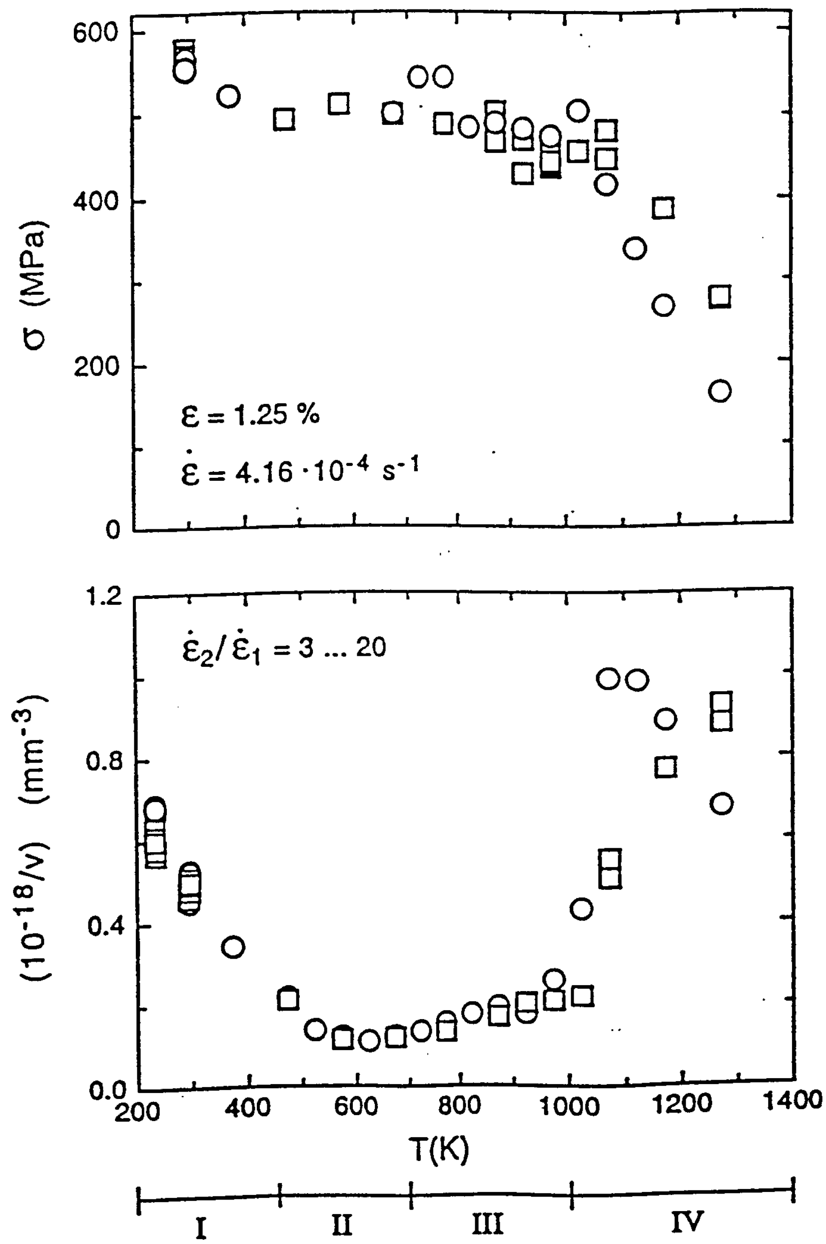
alloy	ΔG (eV)
Ti-47Al-2Cr-0.2Si	0.47
Ti-49Al-0.4C	0.55

Flow stresses and activation volumes

$$\tau^* = (1/V)\,(\Delta F^* + kT \ln \dot{a}/\dot{a}_0)$$

Ti-47 at.% Al-2 at.% Cr-0.2 at.% Si

○ near gamma, □ nearly-lamellar



activation parameters

Activation parameters

Region III: 700...1000 K

- strong increase of $1/V$ with T

$\tau^* = (1/V)\,(\Delta F^* + kT \ln \dot{a}/\dot{a}_o)$

- nearly lamellar microstructure:

$T \;=\; 900$ K

$\sigma \;=\; 480$ MPa

$V \;=\; 200\ b^3$

$\Delta G \;=\; 3.2$ eV

- Comparison: self-diffusion energy

$Q_{SD} = 3{,}01$ eV (Kroll et al., Brossmann et al.)

- diffusion controlled mechanisms at the transition from brittle to ductile material behaviour?

$\Downarrow$

implications on high-temperature strength

Activation parameters

Region III: 700...1000 K

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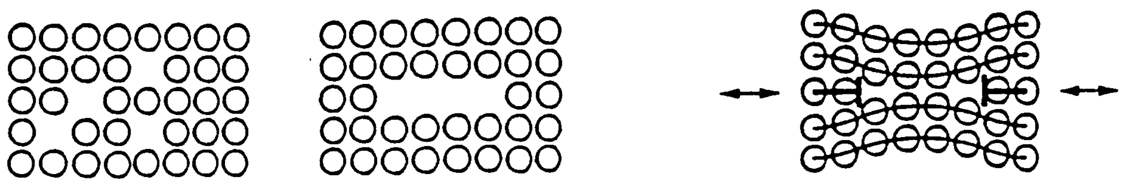
↓

implications on high-temperature strength

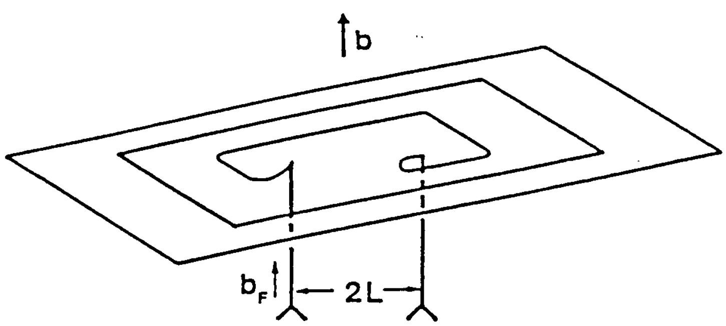
Implications on high-temperature strength

operation of Bardeen-Herring type dislocation climb sources

- nucleation and growth of prismatic loops:

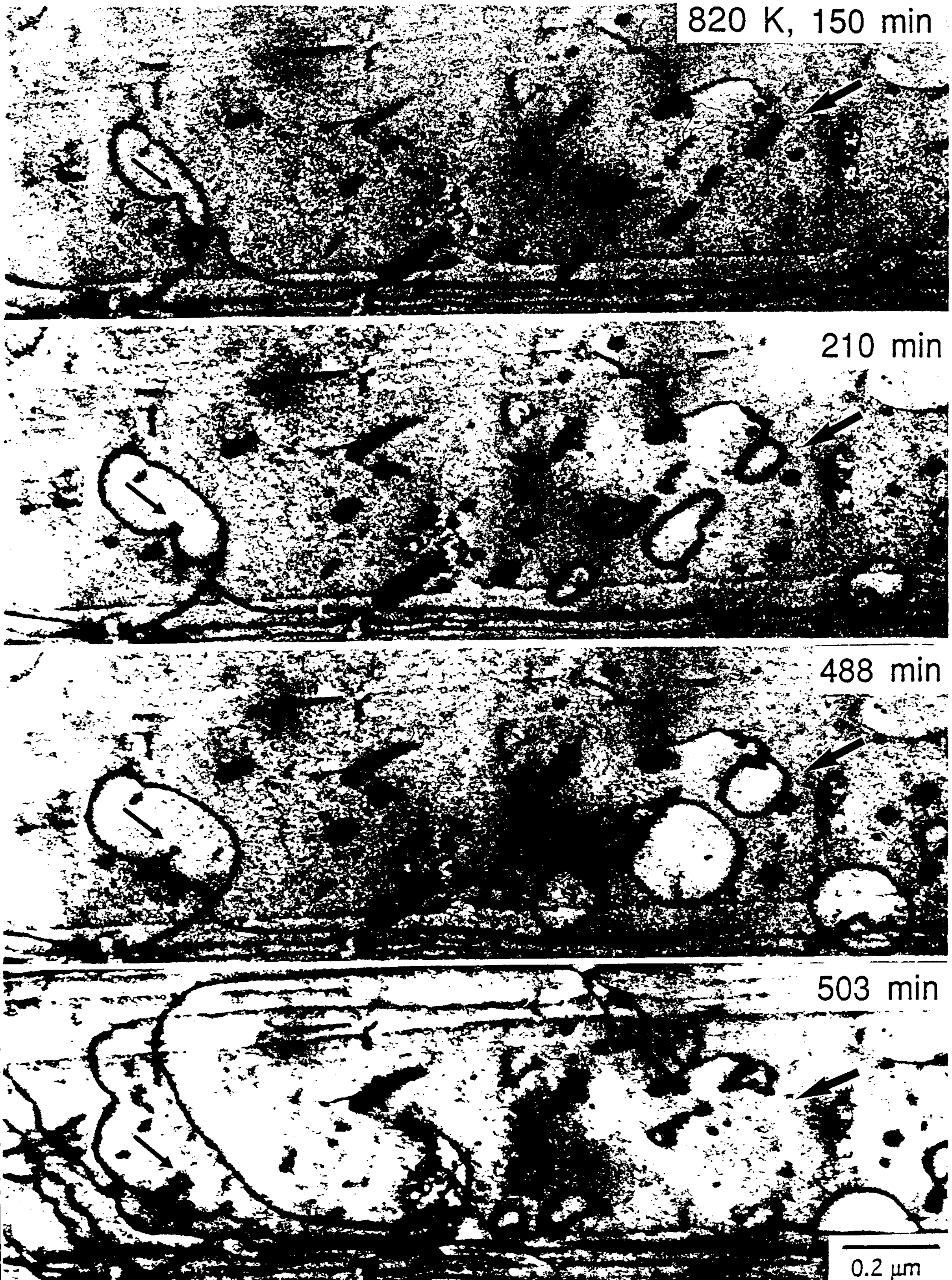


- regenerative climb sources (source attached to dislocations having screw components):



- formation of helical dislocations

TEM in situ observations

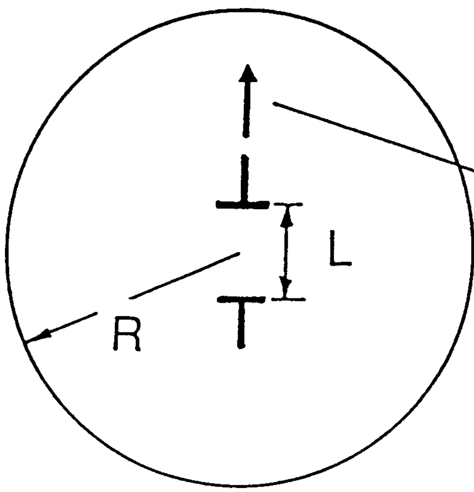


Operation of Barden-Herring type dislocation climb sources during in situ heating inside the TEM.
Ti-48Al-2Cr; acceleration voltage 120 kV, (400 T: 1937, 39, 47, 48)

Evaluation

vacancy source - and sink interaction
(Hirth/Lothe 1992)

creep velocity of either dislocation



$$v_c = \frac{D_s \, \Omega \, \mu}{L(1-\nu)\ln[R/(bL)^{1/2}]}$$

- present situation: $R = 10^4 b$, $L = 10^2 - 10^3 b$

T (K)	v (mm/s)	D_s (m^2/s)
820	2.4×10^{-9}	5×10^{-21}
900	2×10^{-7}	2×10^{-19}

- comparison: T = 1173 K

$D_s = 10^{-17} - 10^{-15}$ m^2/s
(Ouchi et al., Kroll et al.)

Evaluation

vacancies supersaturation required to operate a Bardeen-Herring source of length L (Hirth, Lothe 1992)

$$\ln(c/c_o) = \frac{\mu b \Omega}{L\, 2\pi (1-\nu)\, kT} \ln(L\alpha/1.8\, b)$$

- present situation: $\alpha = 4$

$\Omega = b^3$, $\mu b^3 = 9.5$ eV, $T = 820$ K,
$L = 150 - 350b$

$c/c_o = 3 - 1.7$

- small supersaturation in comparison with those met in rapid quenches ($c/c_o \sim 10^4$)

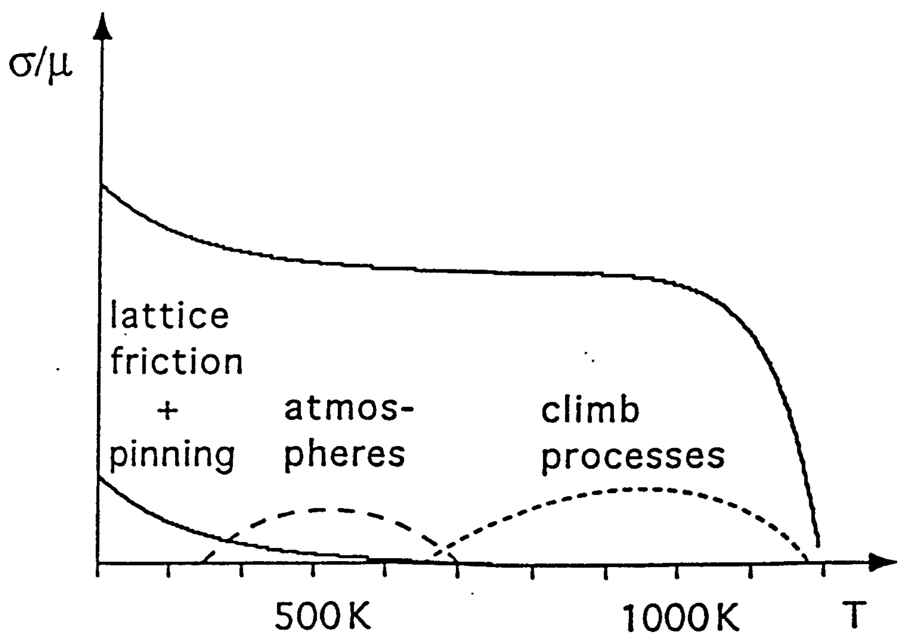
- Bardeen-Herring sources can probably operate throughout the period of fast cooling

Conclusions

Two-phase titanium aluminide alloys of technical significance contain a relatively high level of impurities, such as O, N and C, which impede dislocations due to solution and precipitation hardening.

At temperatures around 500 K dislocation locking occurs due to atmospheres of yet unknown defects.

Among different alloys the effects of these mechanisms are distinguished only quantitatively.



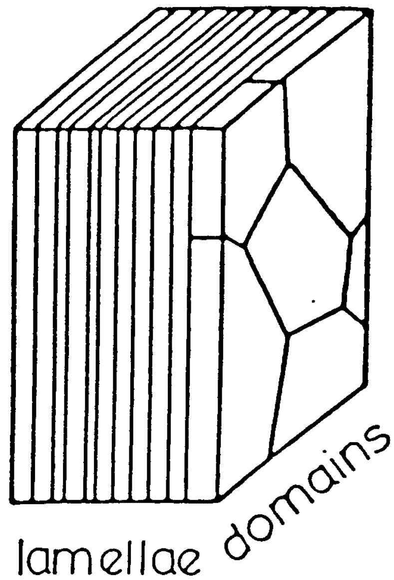
Interface - related deformation phenomena

- structure and stress state of lamellar interfaces
- translation of shear deformation across interfacial boundaries
- crack propagation

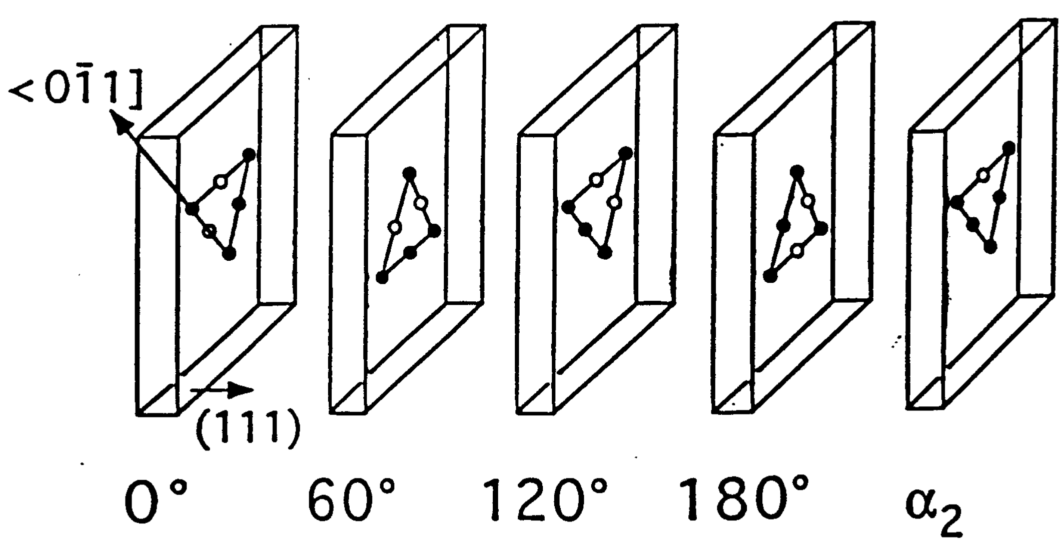
Related problems:

- poor ductility at ambient temperatures
- insufficient resistance against creep and recrystallization at high temperatures

Structural features of lamellar interfaces



- γ phase: domain structure of six ordered variants
- tetragonality of the γ phase c/a = 1.02



HREM Observation of lamellar interfaces

Interfaces types:

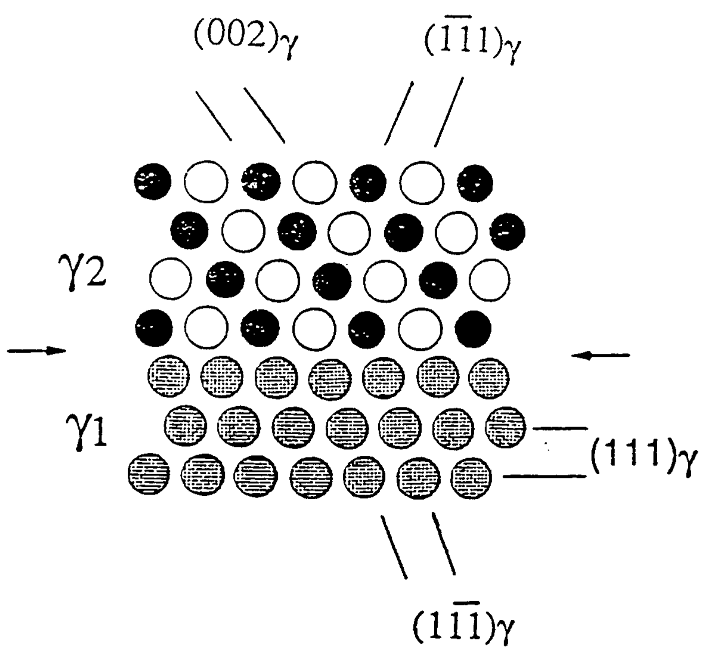
α_2/γ

γ/γ_T true twin

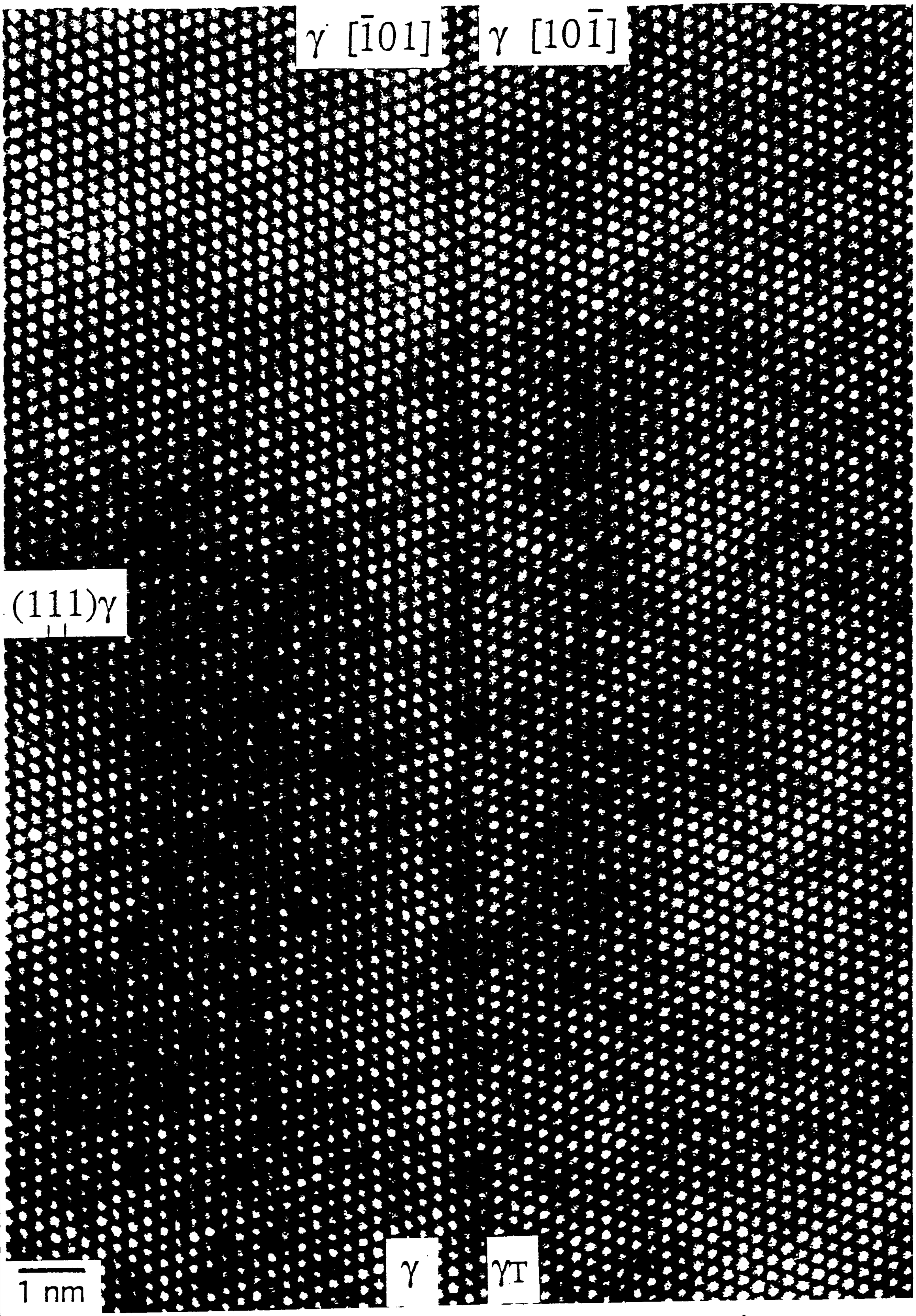
γ_1/γ_2 matrix/matrix

γ_1/γ_2 pseudo-twin

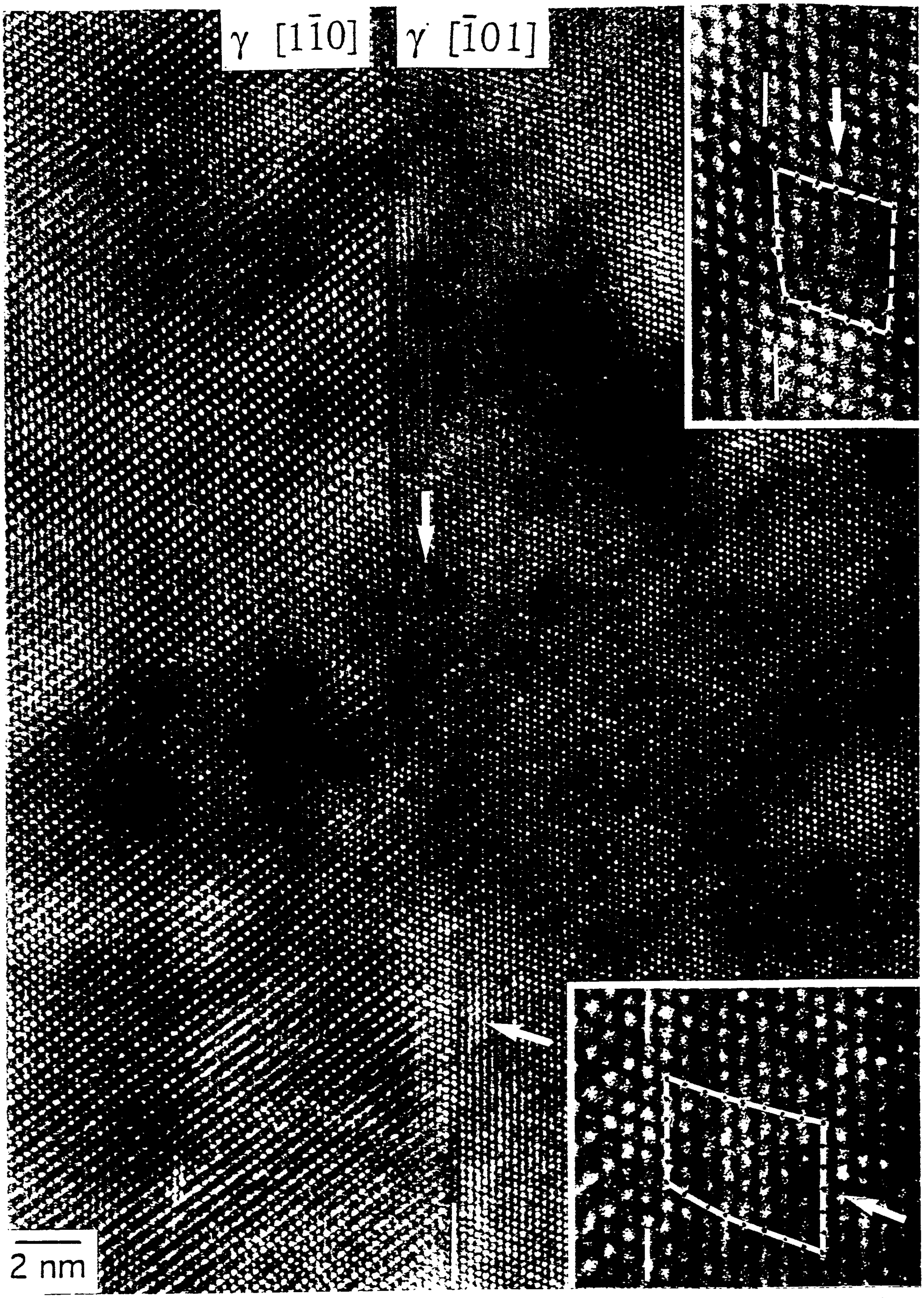
<101> projection of the pseudo-twin:



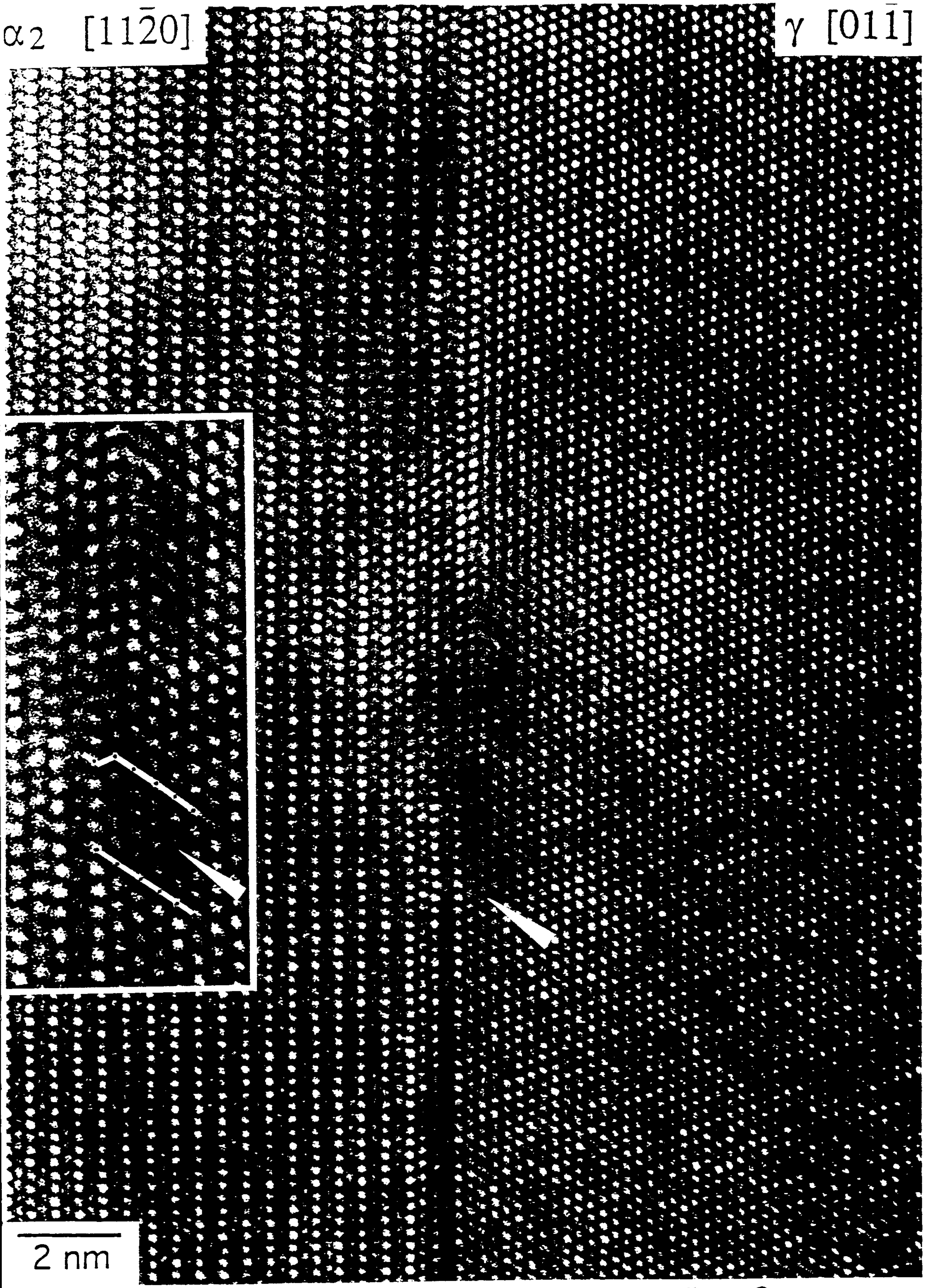
examples



Interface between γ variants with true twin orientation; Ti-48at.%Al-2at.%Cr. (CM 4204)



Interface between γ variants with matrix/matrix orientation; Ti48-at.%Al-2at./%Cr. (CM 4232)



α_2/γ interface in a Ti-48at.%Al-2at.%Cr alloy. Creep deformation: σ = 140 MPa, T = 700 °C, t = 6000 h, ε = 0.69%. (CM 4278)

Stress state of lamellar interfaces

- lattice mismatch largely accommodated by misfit dislocations

- residual homogeneous straining of adjacent lamellae

➡ long-range internal stresses τ at the interfaces

Origin of the residual stresses

- lattice mismatch at semicoherent interfaces
  $\Delta\varepsilon = 1 \dots 2\ \%$

- largely accommodated by misfit dislocations

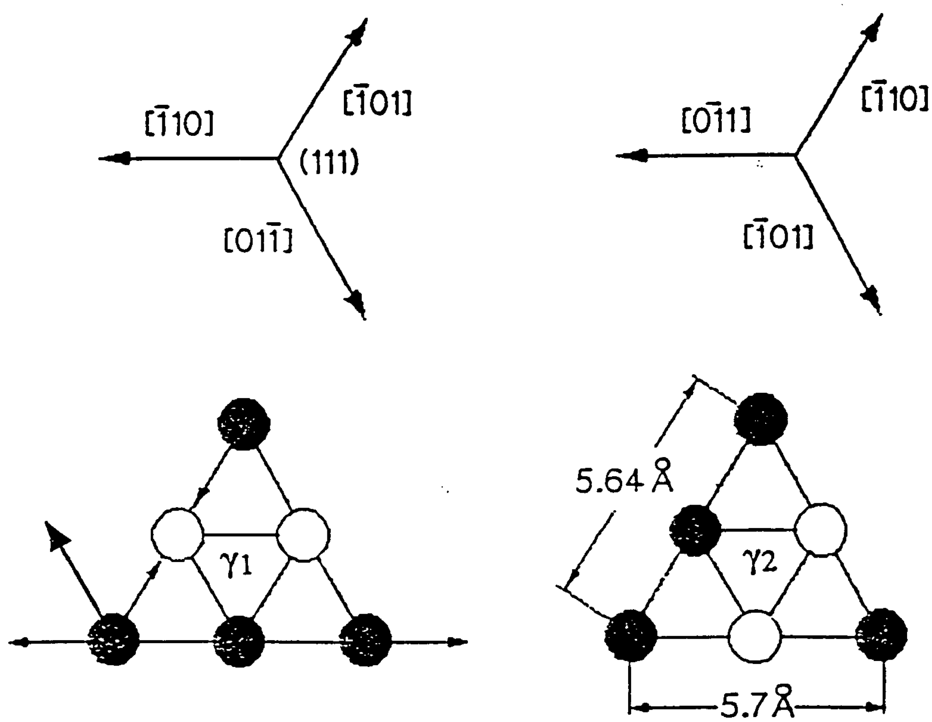
- residual homogeneous straining $\Delta\varepsilon_r$ of adjacent lamellae

- high elastic stiffness of γ-TiAl,
  $\mu = 4.3 \cdot 10^4$ MPa

long-range residual stresses at the interfaces: $\tau = \Delta\varepsilon_r \cdot \mu$

Origin of the residual stresses

Atomic arrangement of the $(111)\gamma$ planes at an interfaces between γ lamellae with pseudo-twin relation

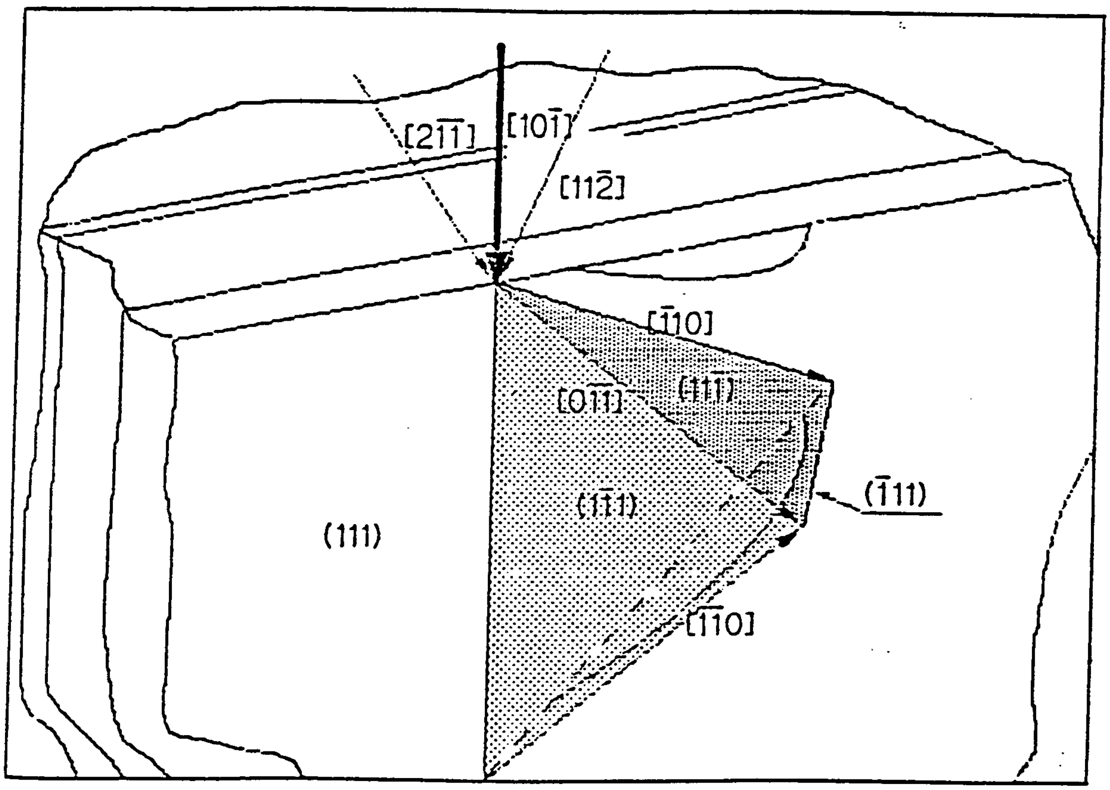


- pure shear deformation along $[0\bar{1}1]$,
- resolved into shear stresses acting on <110> $\{\bar{1}11\}$ slip systems of the adjacent lamellae

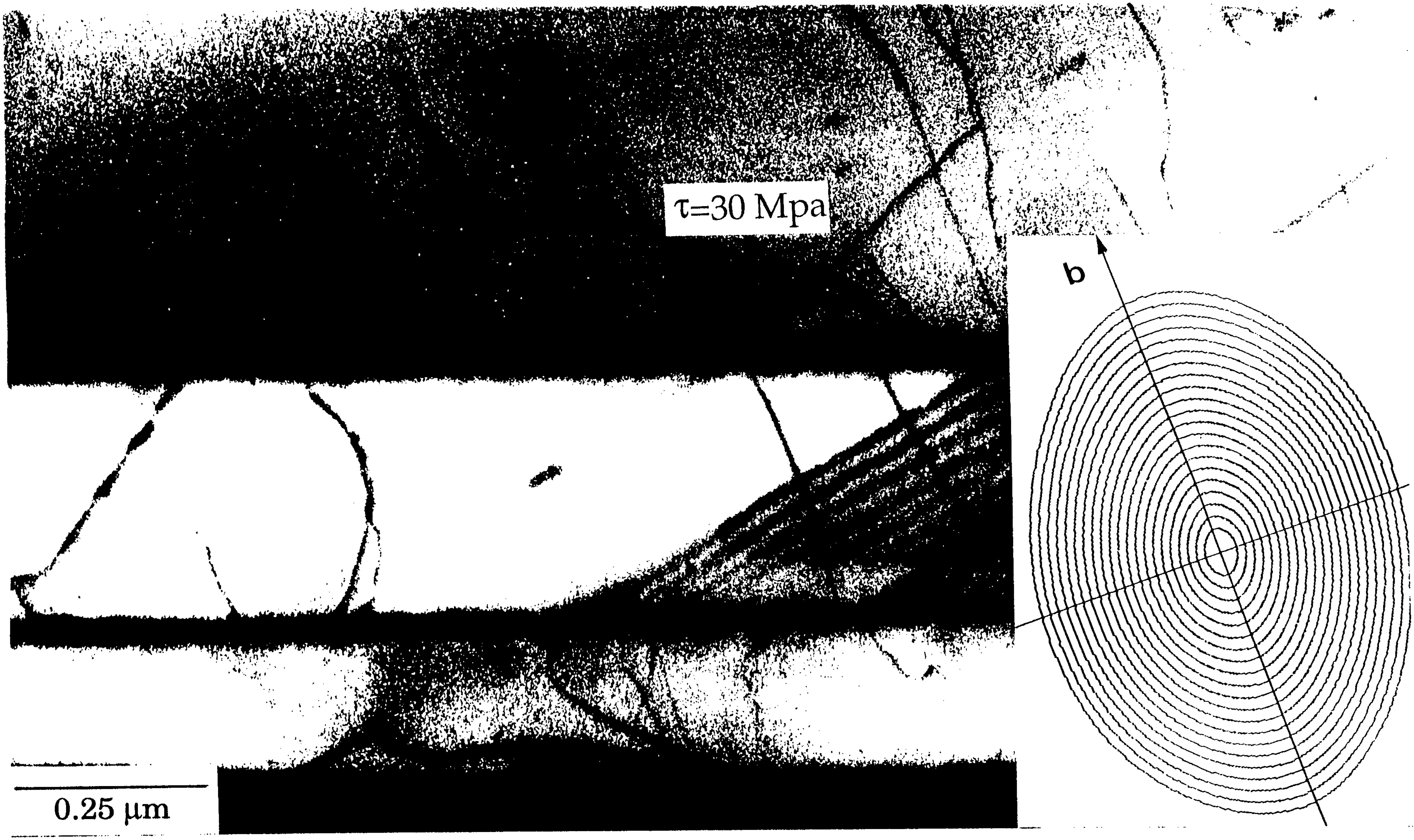
schematic drawing and example

Stress state of lamellar interfaces

Coherency stresses at semicoherent $(111)_\gamma$ interfaces, resolved into shear stresses τ acting on <110>{111} slip systems



experimental observation,
comparison with line tension configurations



Comparison of the loop shape with the configuration expected from the dislocation line tension model (CM 1472)

The dislocation line tension model

DE Witt and Koehler

$$T = [\mu b^2/4\pi(1-\nu)][\ln(R/r_o)+C(\Theta)]$$

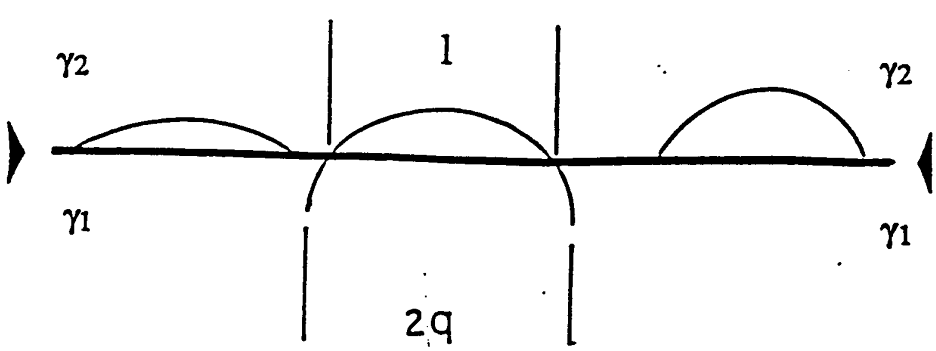
$R = l/5$, $r_o = b/8$ (Hirth, Lothe)

$$q = [\mu b/4\pi(1-\nu)\ \tau][\ln\ l + \ln\ (8/5b)]$$

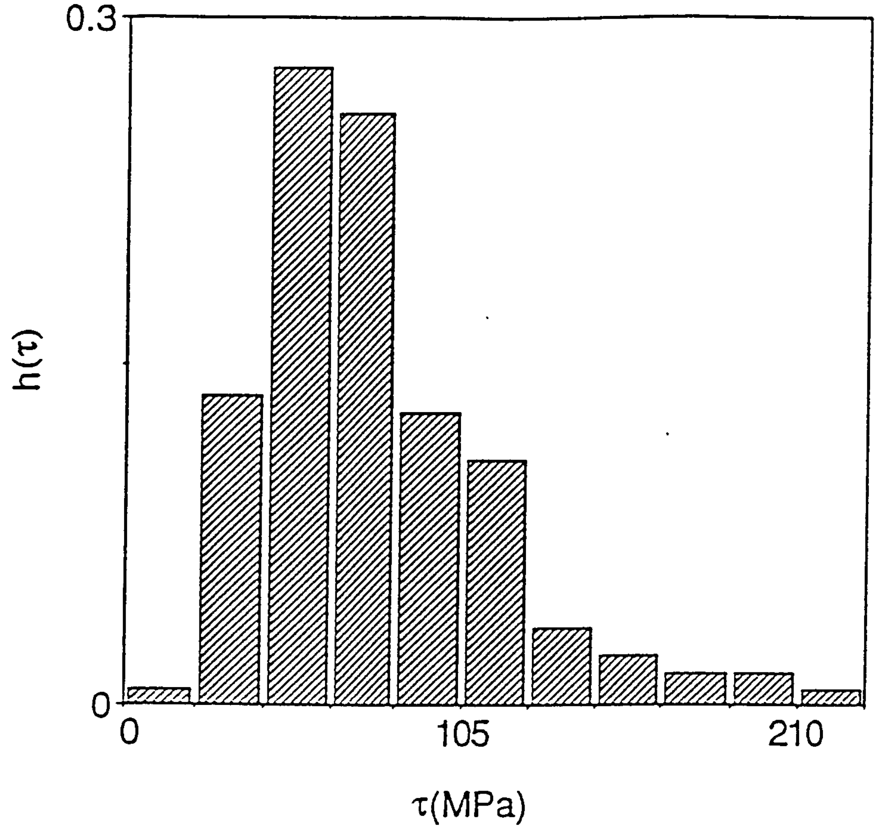
τ = const.

$q = E' \ln l + D$

Evaluation:



Stress state of lamellar interfaces



Distribution of the internal stresses τ acting on dislocation loops emitted from interfacial boundaries.

Comparison with deformation experiments:

$\sigma_a = 430$ MPa $\rightarrow$ $\tau_a = \sigma_a/3 = 140$ MPa

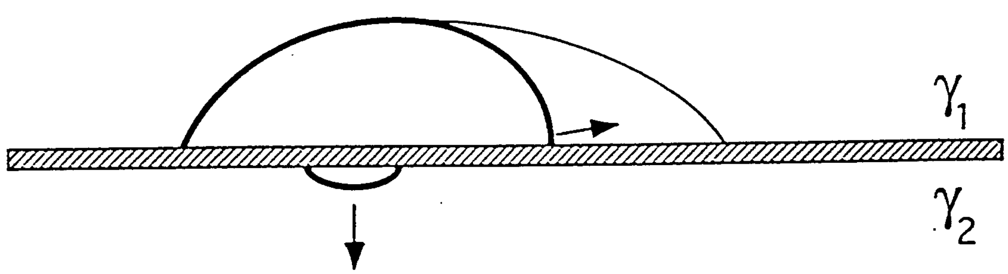
$\tau = 20 \ldots 220$ MPa

Consequences:

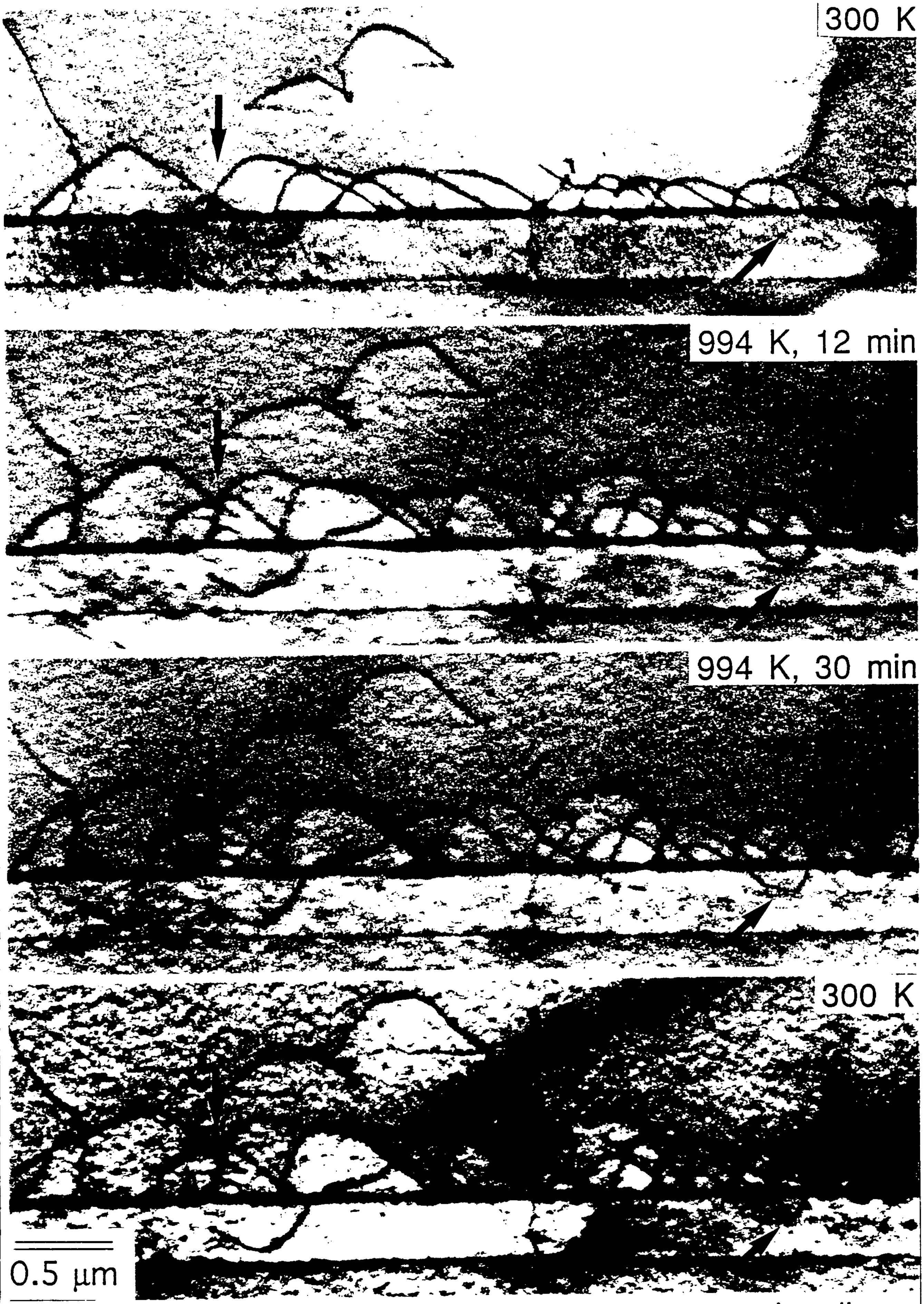
- high density of dislocation sources
- relaxation of local stress concentrations
- contribution to glide and climbing processes

Interfaces as dislocation sources

- Dislocation segments strongly bowed out due to coherency stresses and thermal stresses
- friction forces impede propagation
- unzipping and generation of new loops at elevated temperatures

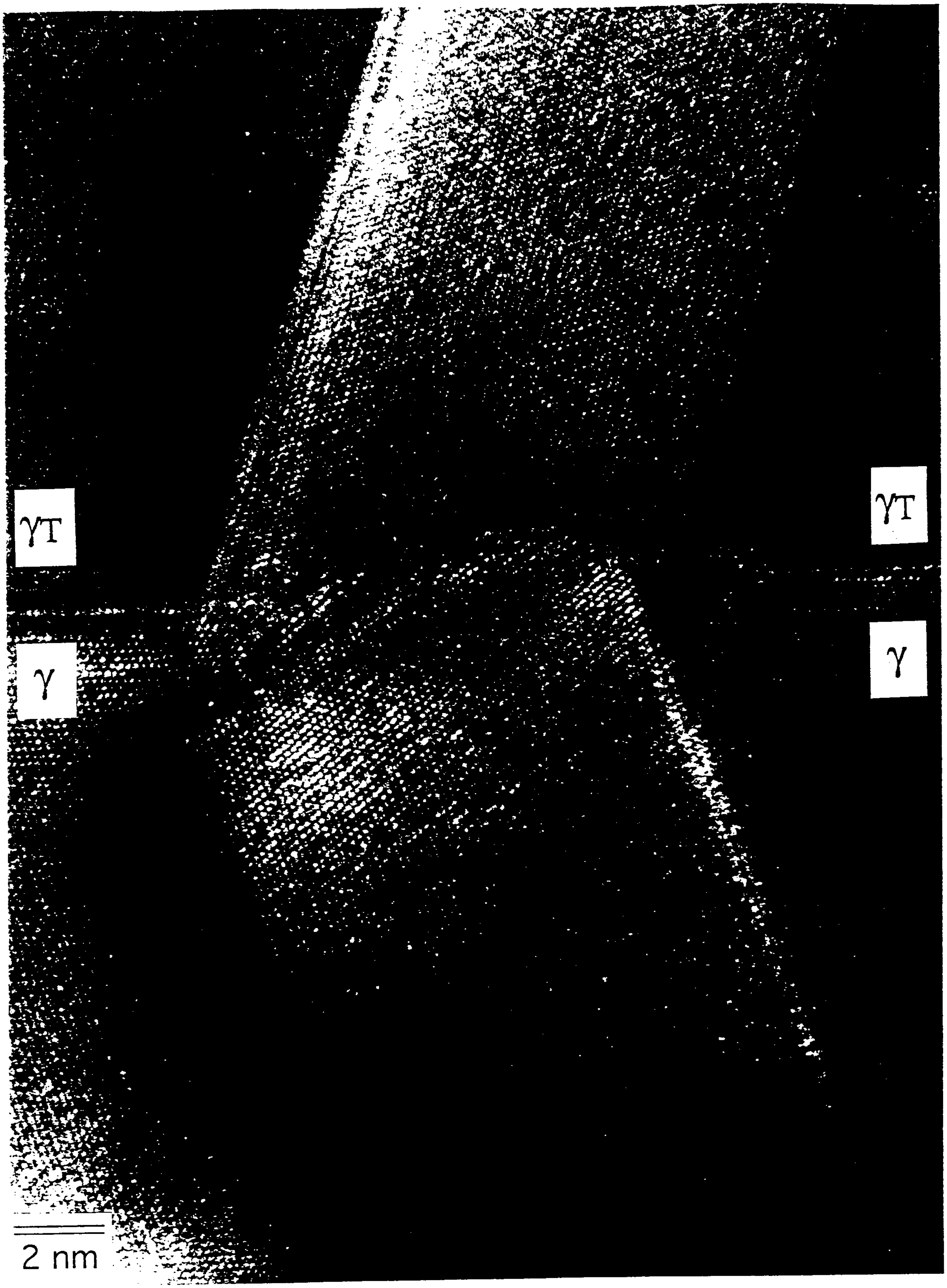


TEM in situ heating study



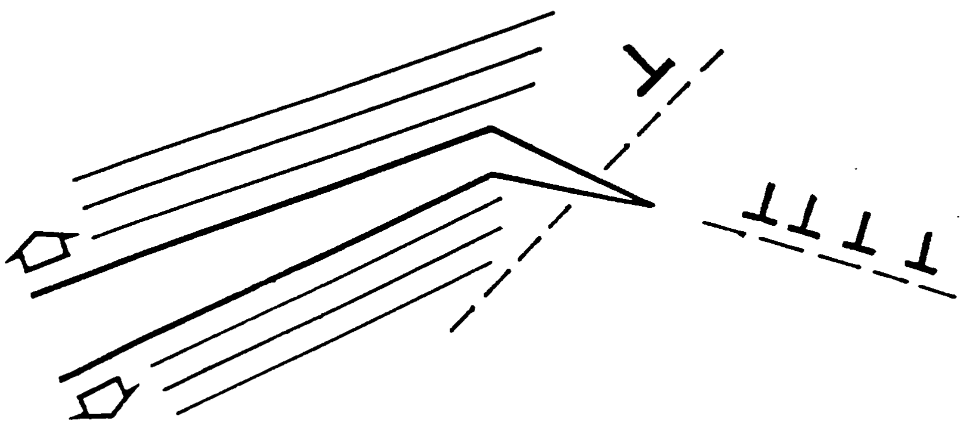
Evolution of the dislocation structure at lamellar interfaces during in situ heating;
Ti-48at.%Al-2at.%Cr. C 297, 301, 308

Dislocation glide processes initiated at semicoherent interfaces in a two-phase ($\alpha_2 + \gamma$) TiAl alloy. Deformation at room temperature (CM3569).



Translation of twinning deformation through an interfacial boundary γ/γ_T between lamellae with true twin relation. T = 300 K, ε_f = 0.2% (3A5743)

Crack Propagation



Possible Processes:

- lattice decohesion
- crack deflection
- crack tip blunting
- crack tip shielding
- formation of a plastic zone

TEM: Interaction of cracks with lamellar interfaces

Crack propagation

- Inverse correlation between ductility and fracture toughness in $(\alpha_2 + \gamma)$ titanium aluminide alloys

Kim and Dimiduk 1991, Chan and Kim 1992

Fully-lamellar microstructures:
Low ductility/high toughness

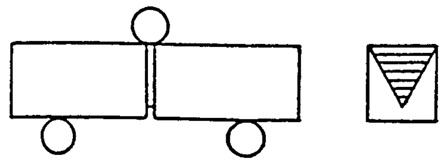
Duplex microstructures:
High ductility/low toughness

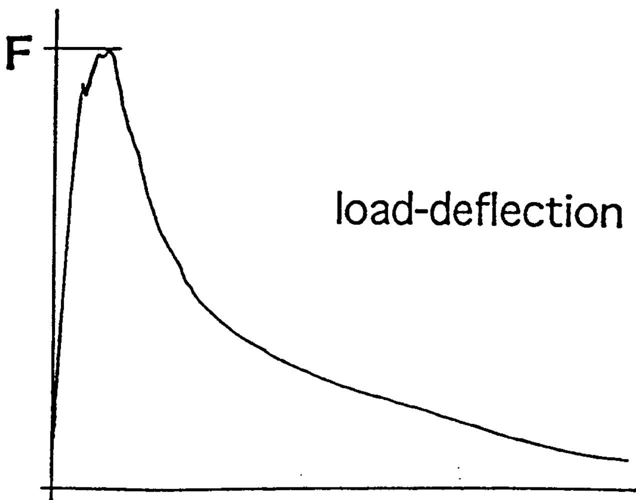
↓

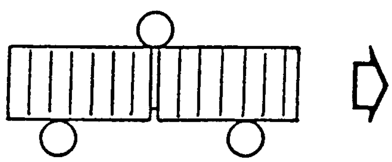
Fracture toughness of lamellar $(\alpha_2 + \gamma)$TiAl

Fracture toughness of $(\alpha_2 + \gamma)$ TiAl

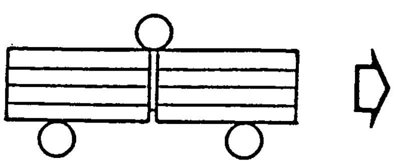
Chevron-notch





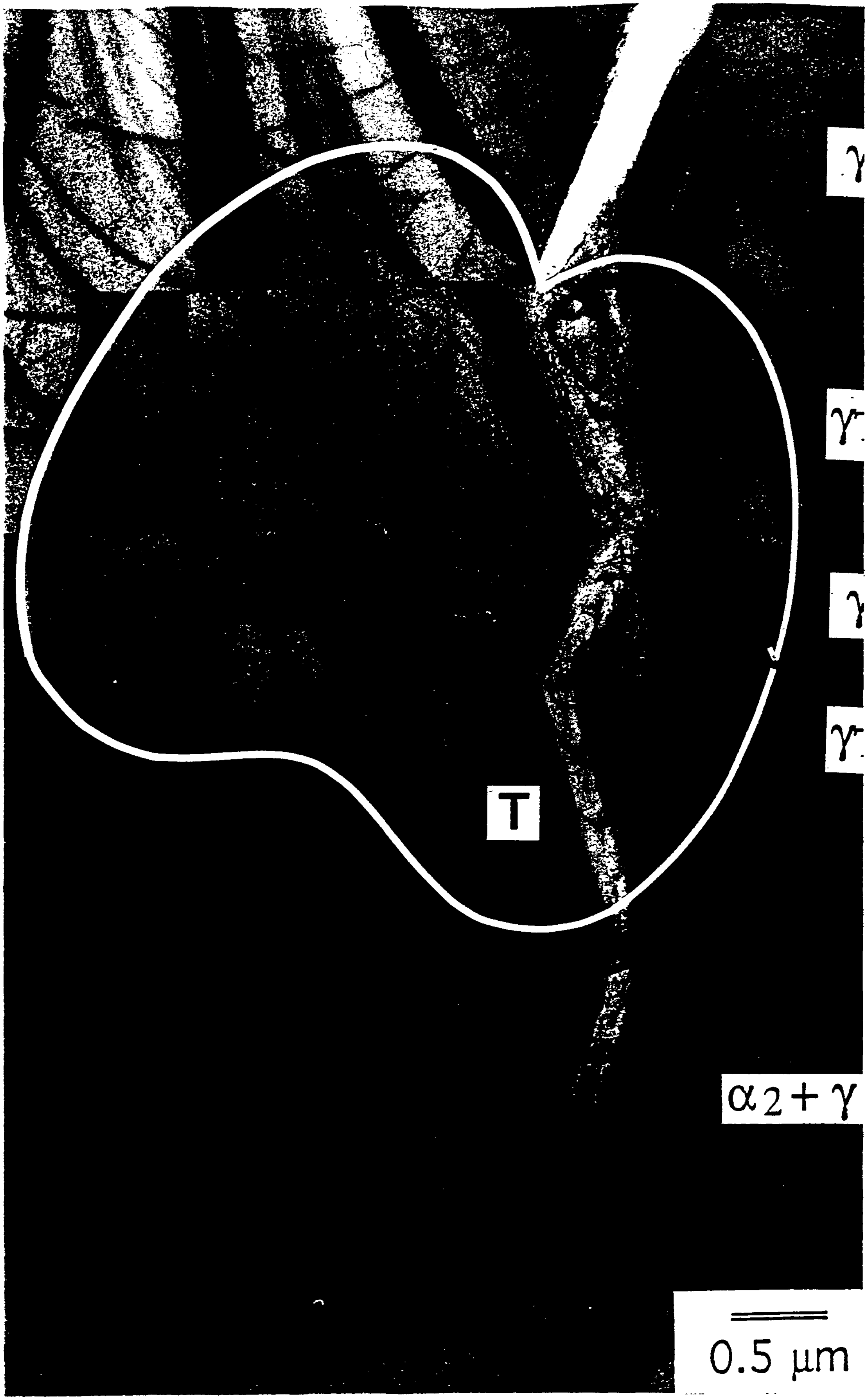


$K_{Ic} = 15.2 \pm 2.3$ MPa $m^{1/2}$



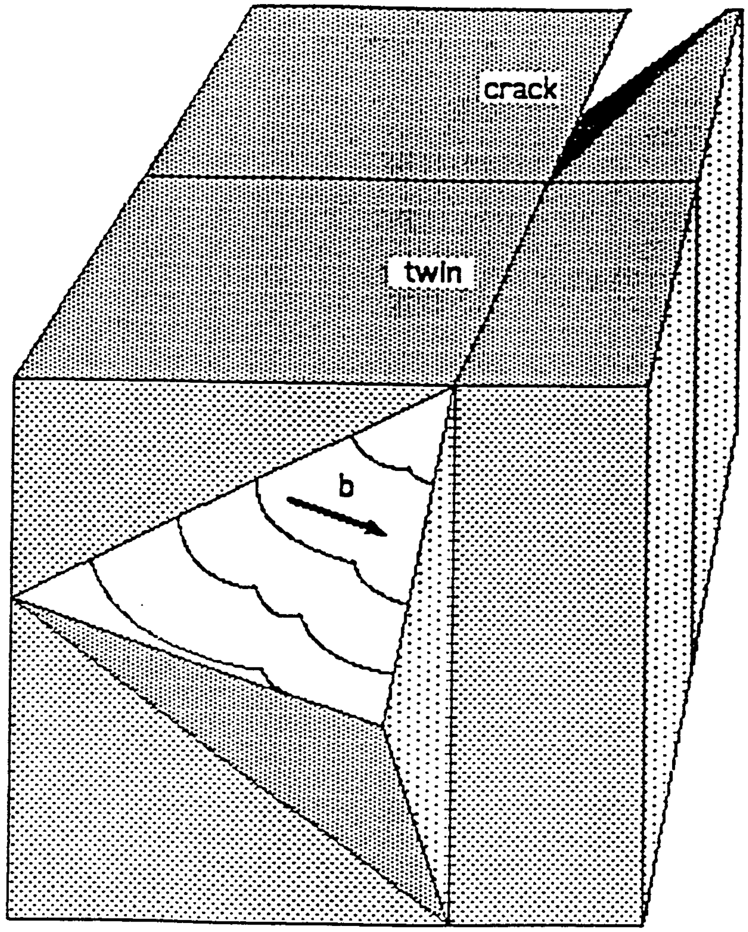
$K_{Ic} = 22.1 \pm 1.9$ MPa $m^{1/2}$

TEM: interactions of crack tips with lamellar interfaces

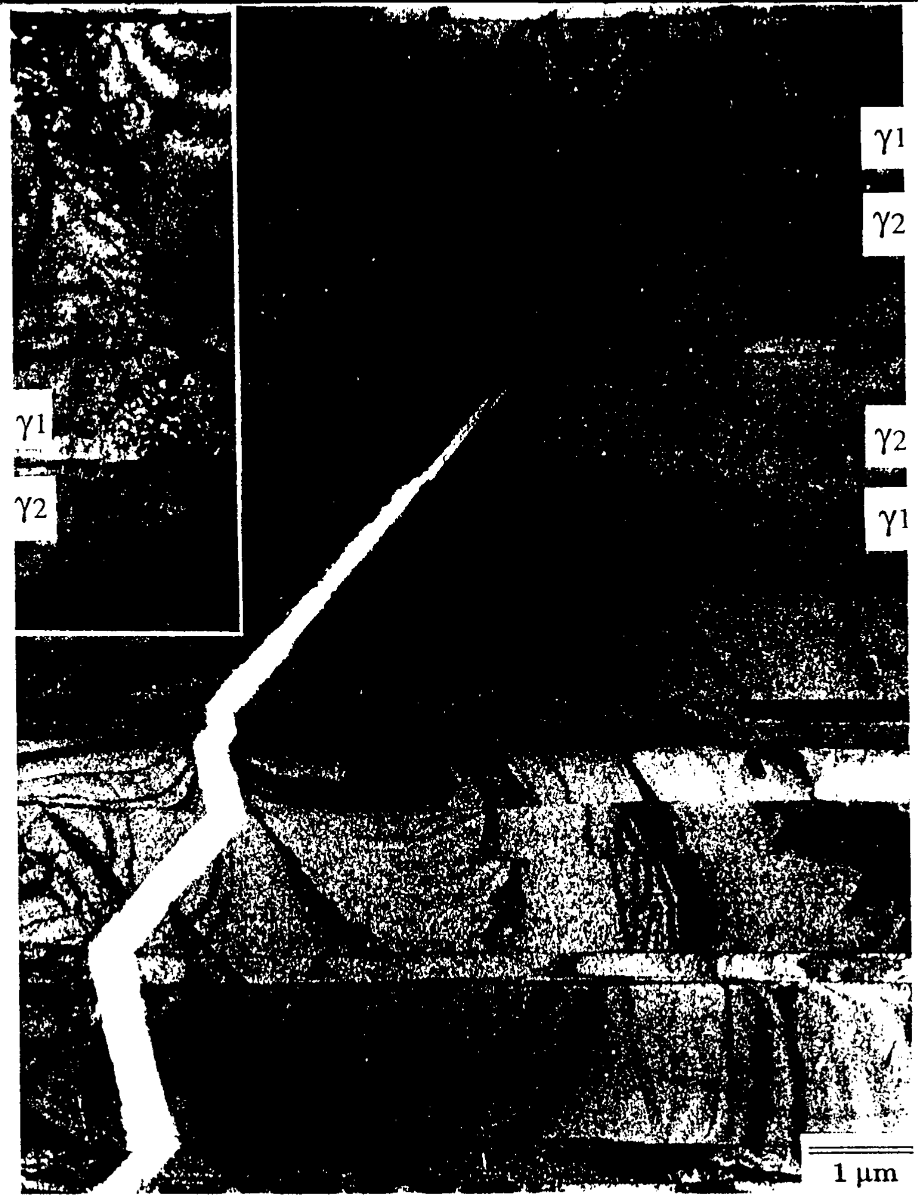


Crack propagation in a two-phase ($\alpha_2 + \gamma$) TiAl alloy with a duplex microstructure

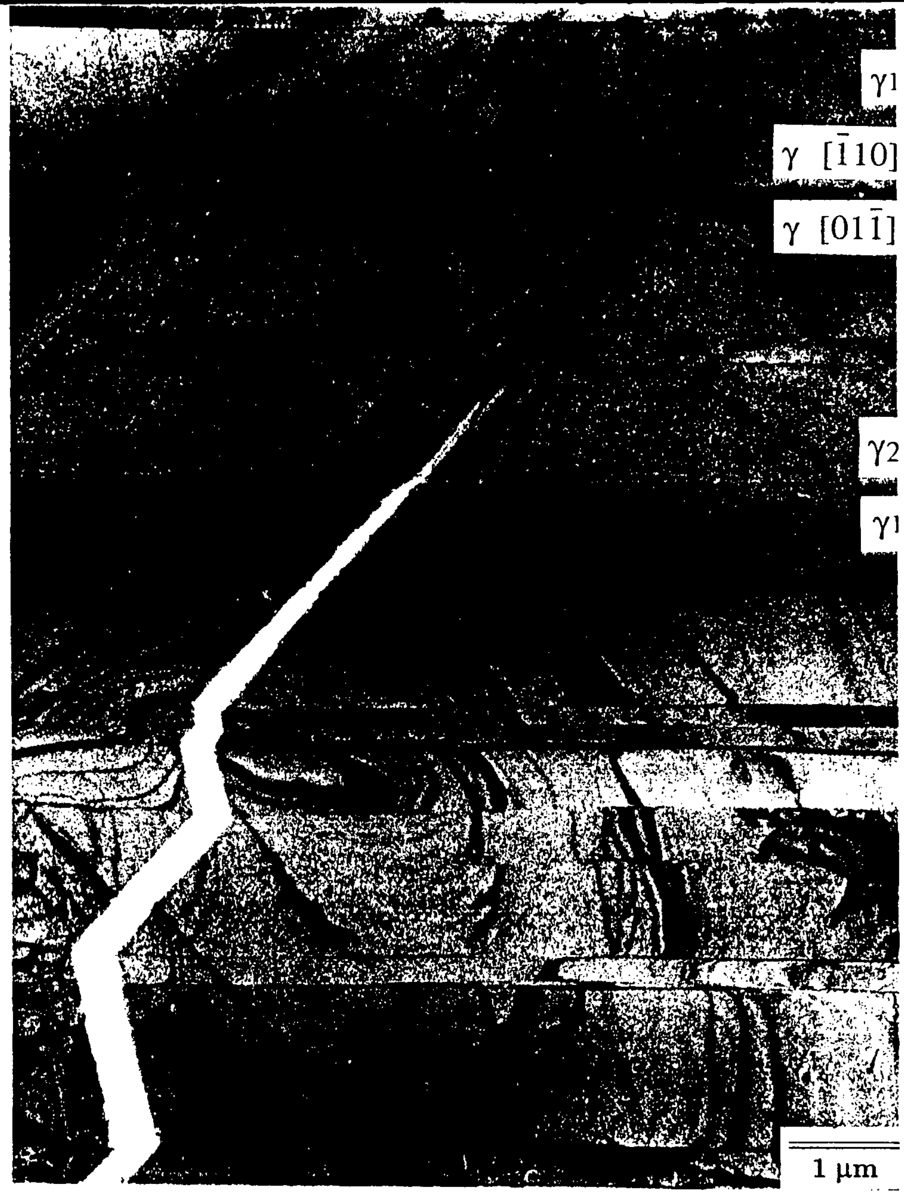
Crack Propagation



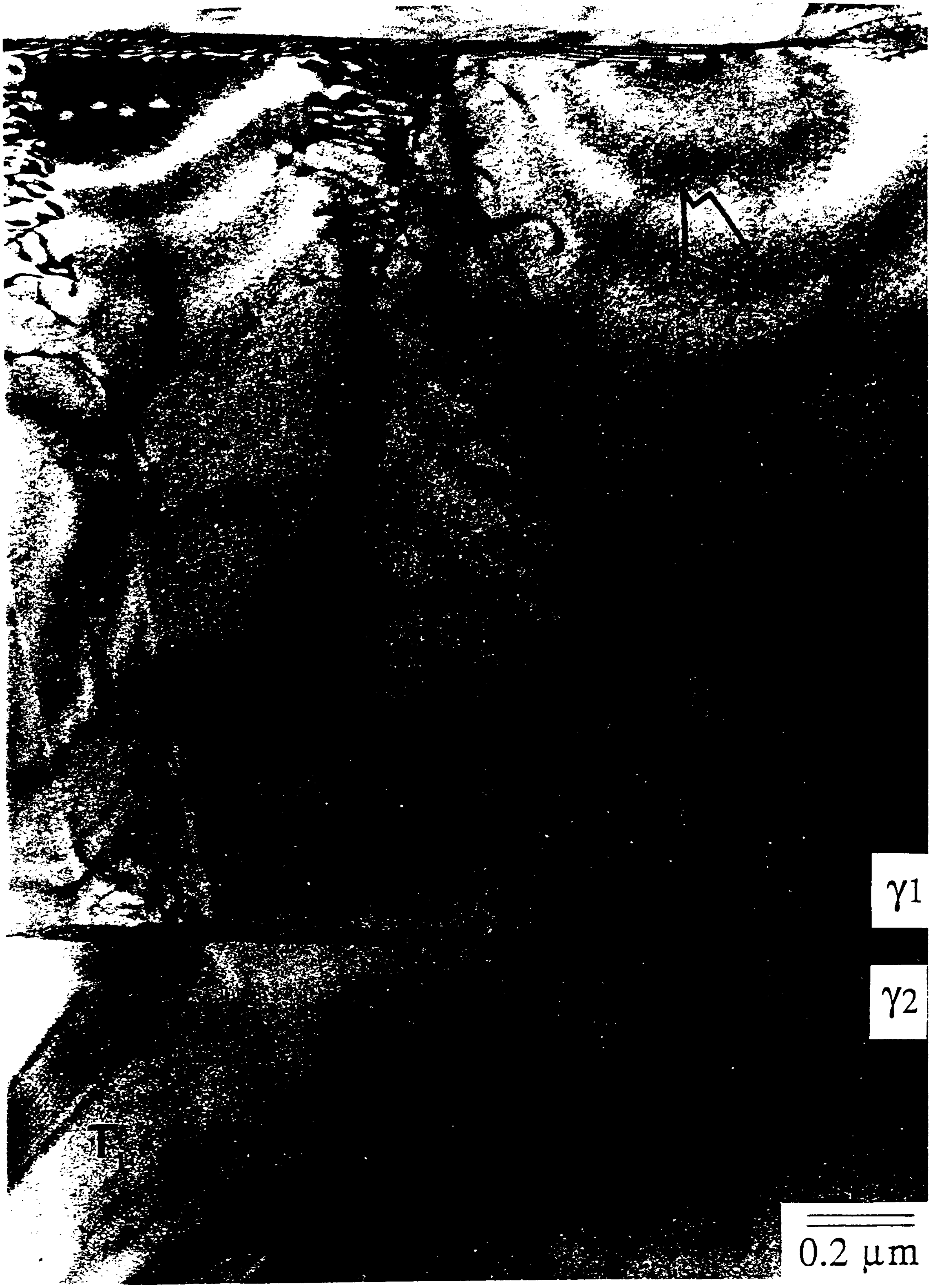
- $\{111\}$ cleavage planes
- twinning precedes crack propagation
- immobilisation at semicoherent interfaces
- shielding of the crack tip, τ = 70 ... 290 MPa



Crack propagation in a two-phase ($\alpha_2+\gamma$) titanium aluminide alloy (CM 3293, CM 3334).



Crack propagation in a two-phase ($\alpha_2+\gamma$) titanium aluminide alloy (CM 3293).



Shielding of a crack tip in a two-phase (α_2+γ) titanium aluminide alloy by deformation twins and (1/2) <110] dislocations (CM 3334).

Conclusions

The deformation behaviour of $(\alpha_2+\gamma)$ titanium aluminides is closely related to lamellar interfaces.

Semicoherent α_2/γ and γ/γ interfaces are characterized by a high density of misfit dislocations and residual coherency stresses.

These structural features support the generation of glissile dislocations and of a fine dispersion of deformation twins.

The generation mechanisms of dislocations and twins are involved in the translation of shear deformation across lamellar boundaries and contribute to stabilize crack propagation.

The low ductility of the material seems therefore not to result from a lack dislocations but from their insufficient mobility.

Implications on creep resistance

- design requirements regarding long term creep resistance:
  $T = 700\ °C$, $\sigma = 150$ MPa, $t = 10.000$ h $\rightarrow \varepsilon \leq 1\%$,
  nominal creep rate $\dot{\varepsilon} \leq 10^{-10} s^{-1}$
  not yet fulfilled

- problem: fast primary creep

- potential mechanisms:
  non-conservative dislocation processes,
  structural changes

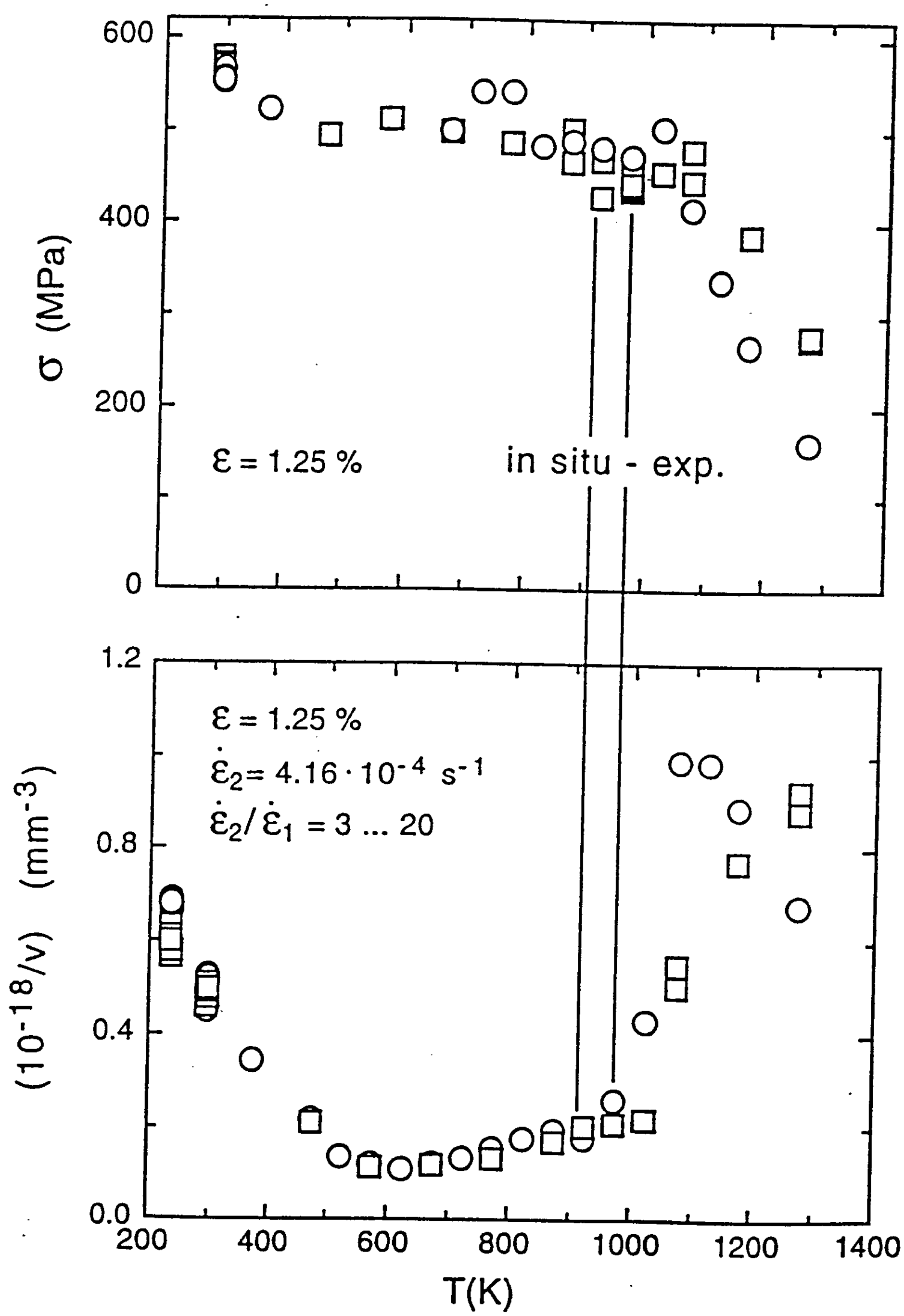
TEM-observations on lamellar
Ti-48 at.% Al-2 at.% Cr

- in situ heating studies

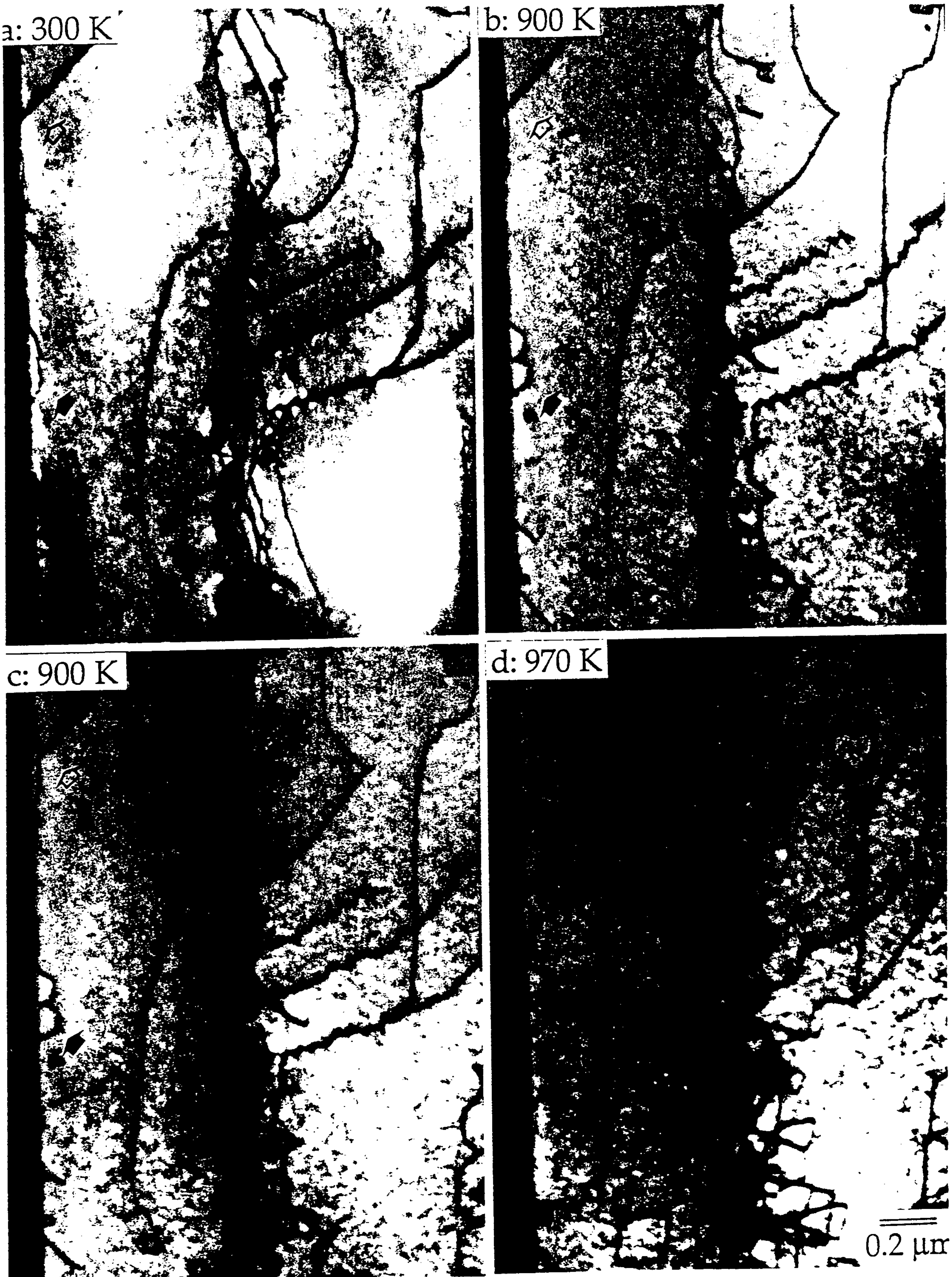
- defect structure of samples crept at
$T = 700\ °C$, $\sigma = 150$ MPa for 6000 h to $\varepsilon = 0.69\%$
nominal creep rate $\dot{\varepsilon} = 3 \times 10^{-10} s^{-1}$

In situ heating experiment

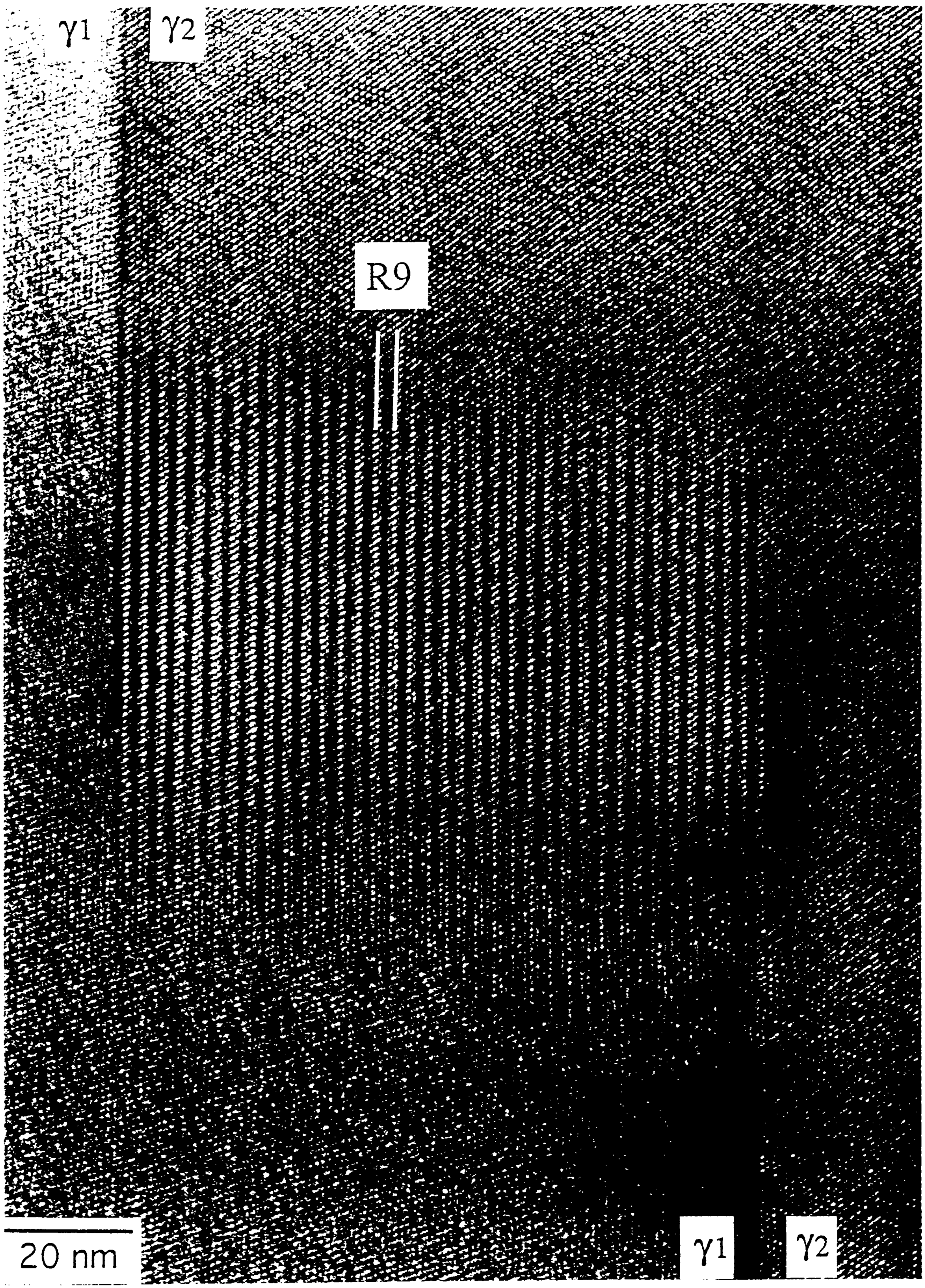
- relationship to strength properties



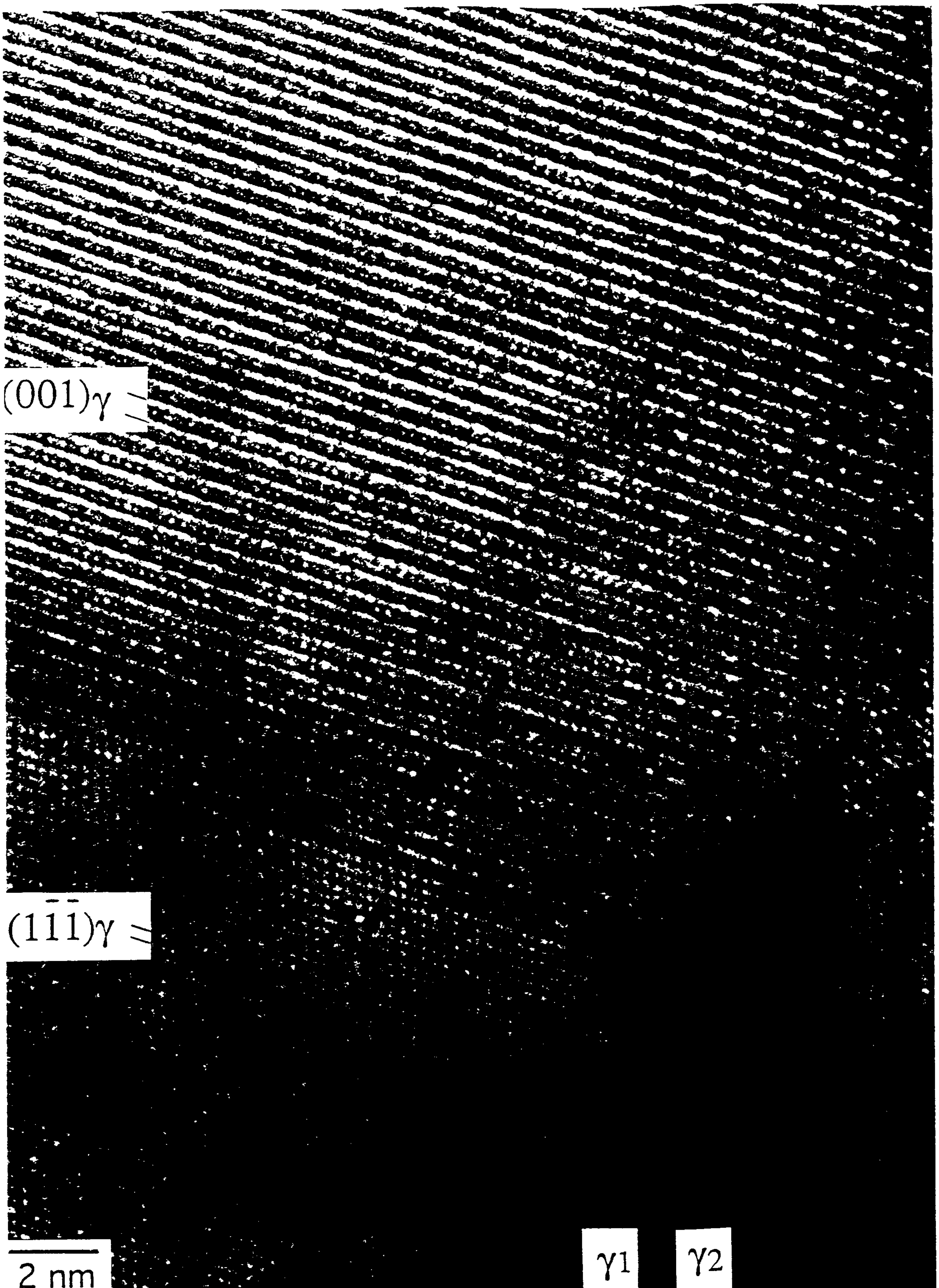
in situ study, T = 900 - 970 K



Dislocation loops emitted from an interfacial boundary during in situ heating inside the electron microscope
(400 T 691, 694, 695, 698)



Formation of the R9 structure at a ledge in a semicoherent interface of a Ti-48at.%Al-2at.%Cr alloy. Creep deformation: $\sigma = 140$ MPa, T = 700 °C, t = 6000 h. $\varepsilon = 0.69\%$. (CM 4178)



Recrystallized γ grain within a lamellar colony of a Ti-48at.%Al-2at.%Cr alloy. Creep deformation: σ = 140 MPa, T = 700 °C, t = 6000 h, ε = 0.69%.
(CM 4214)

Recrystallized γ grain within a lamellar colony of a Ti-48at.%Al-2at.%Cr alloy. Creep deformation: σ = 140 MPa, T = 700 °C, t = 6000 h, ε = 0.69%.
(CM 4216)

How to improve high-temperature strength?

many metallurgical factors have to be considered: grain size, alloying additions, phase distribution and stability etc.

► regarding present observations:

- dislocation glide and climb should be impeded
- structural changes have to be prevented during service

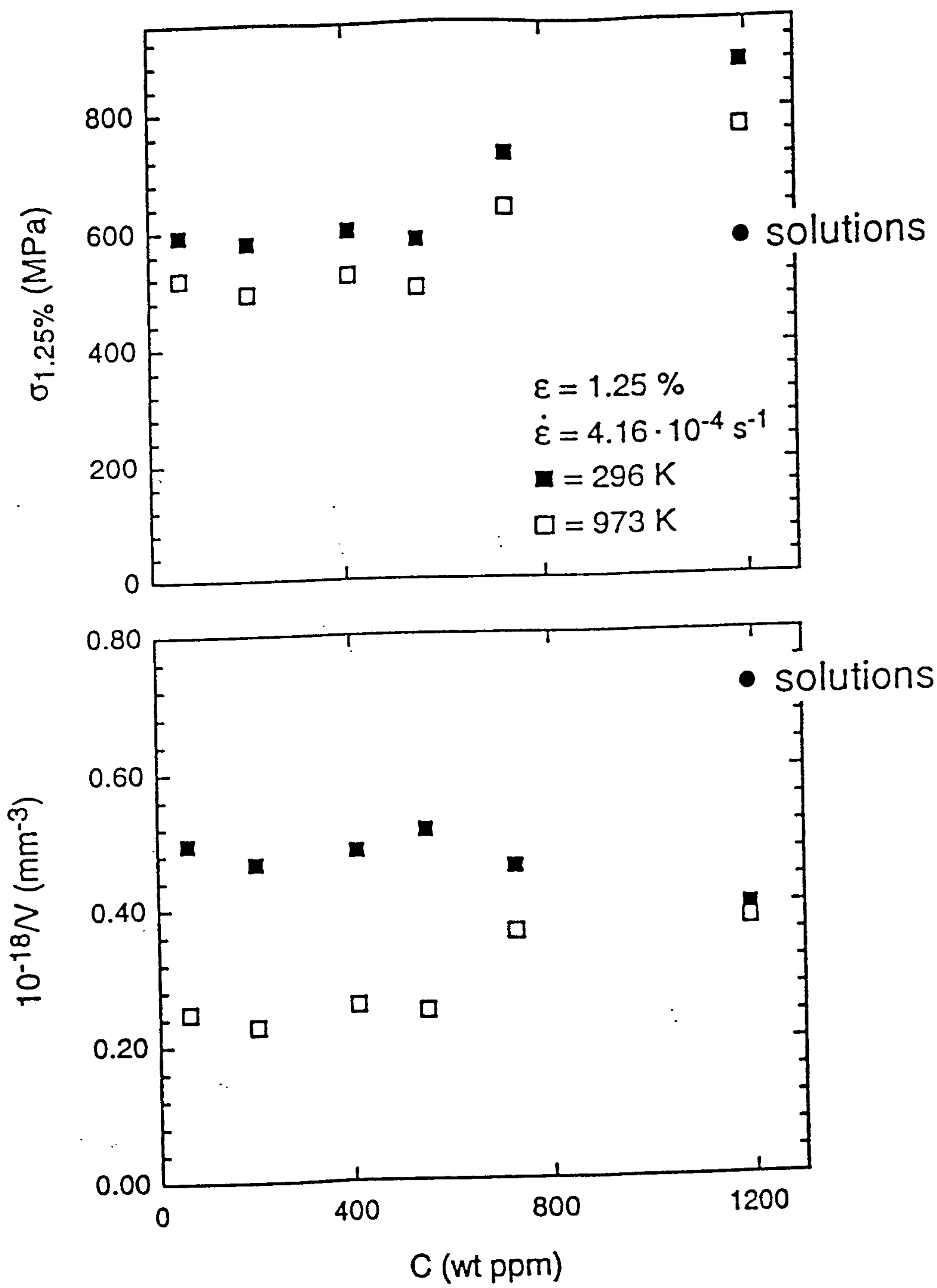
► potential mechanisms:
dislocation locking due to solutions, precipitates, ageing

⬇

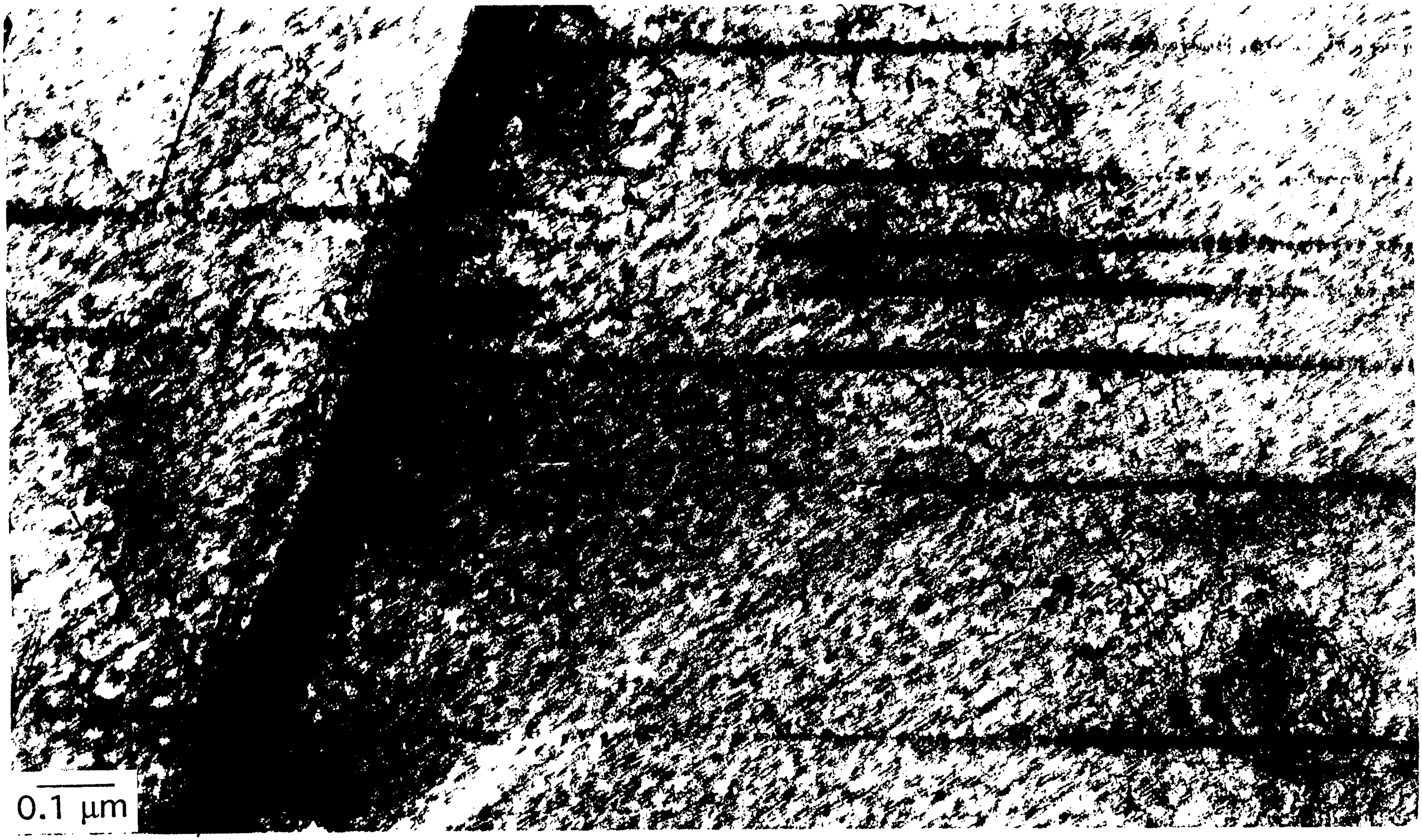
flow stresses and activation volumes of Ti-49 at.% Al + (60 - 1200) wt.ppm C, thermal treatments for solution and precipitation of C and N

Precipitation hardening in γ(TiAl) containing carbon and nitrogen

Ti-49 at.% Al + (60 - 1200 wt.ppm C

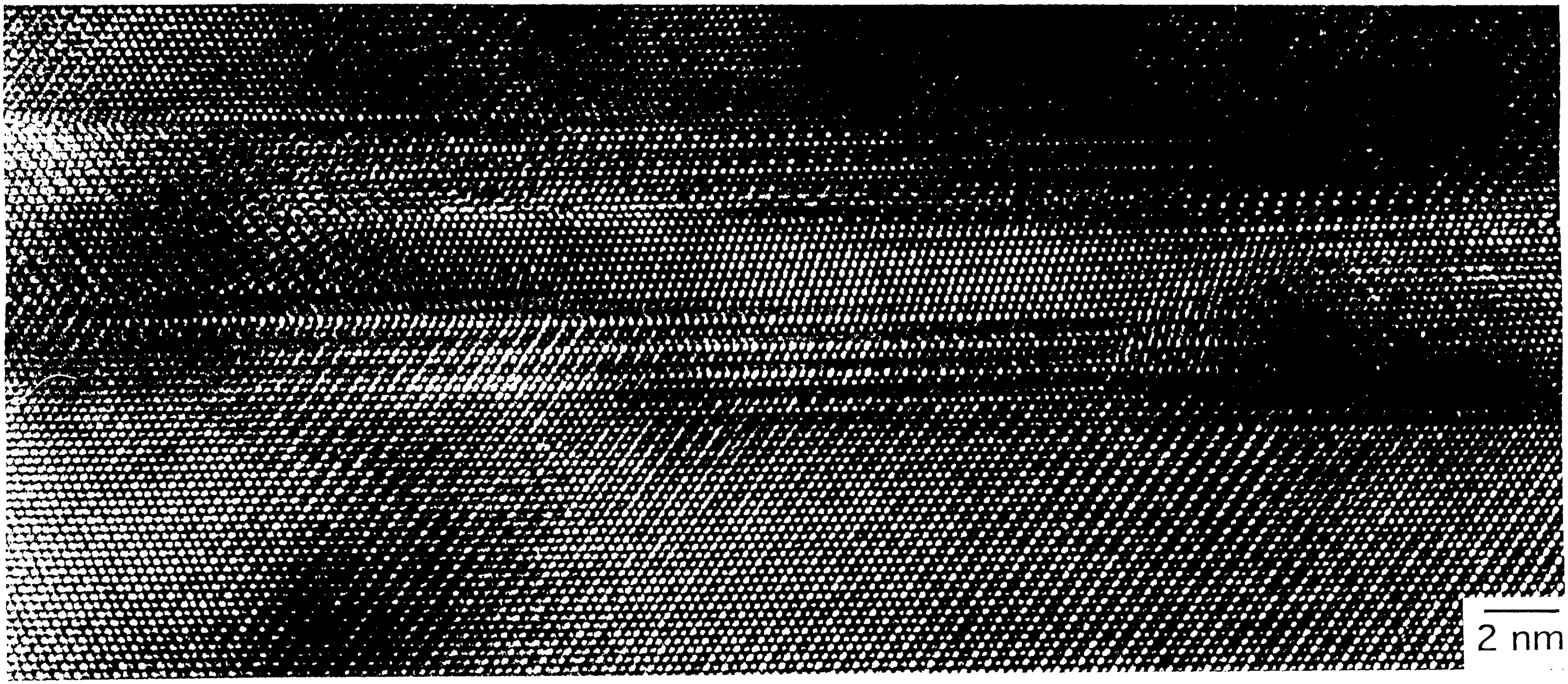


▶ hardening due to large precipitates, athermal contribution to flow stress



Deformation structure of a Ti-49 at.% Al-0.4 at.% C alloy.
Compression at 300 K to strain $\varepsilon = 3$ %. (CM 5858)

Interactions of deformation twins with Ti_3AlC precipitates. Ti-49 at.% Al-0.4 at.% C. Deformation at 300 K to $\varepsilon = 3$ %. (CM 5845)



Interactions of deformation twins with Ti_3AlC precipitates. Ti-49 at.% Al-0.4 at.% C. Deformation at 300 K to $\varepsilon = 3$ %. (CM 5807)

Conclusions

degradation of strength properties of two-phase titanium aluminides at elevated temperatures due to

- non-conservative dislocation mechanisms
- dislocation multiplication by climb sources
- significant changes of the microstructure, particularly during long-term creep loading

potential mechanism to improve high-temperature strength:
hardening due to Ti_3AlC precipitates

problems:
- thermal stability of precipitates during service
- balanced properties of low-temperature ductility and high temperature strength

Design against fracture and fatigue in TiAl-based aluminides

Paul Bowen
Professor of Mechanical Metallurgy

School of Metallurgy and Materials/IRC
The University of Birmingham

Issues

1. Lower bound fracture toughness values in fully-lamellar microstructures

2. Crack growth resistance curves and the use of defect tolerance design

3. Total life: traditional concepts of S-N curves

4. Problems:

    i) sampling volume

    ii) stress concentrations

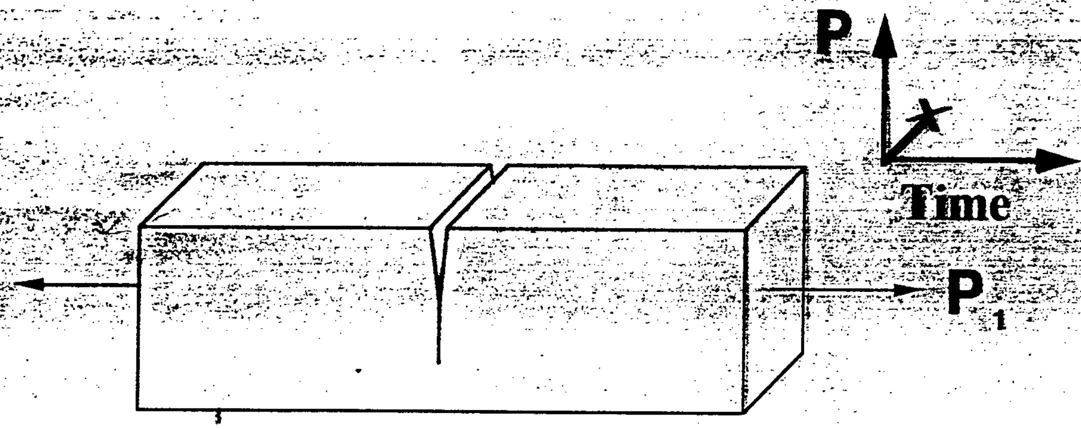
5. Microstructural features:

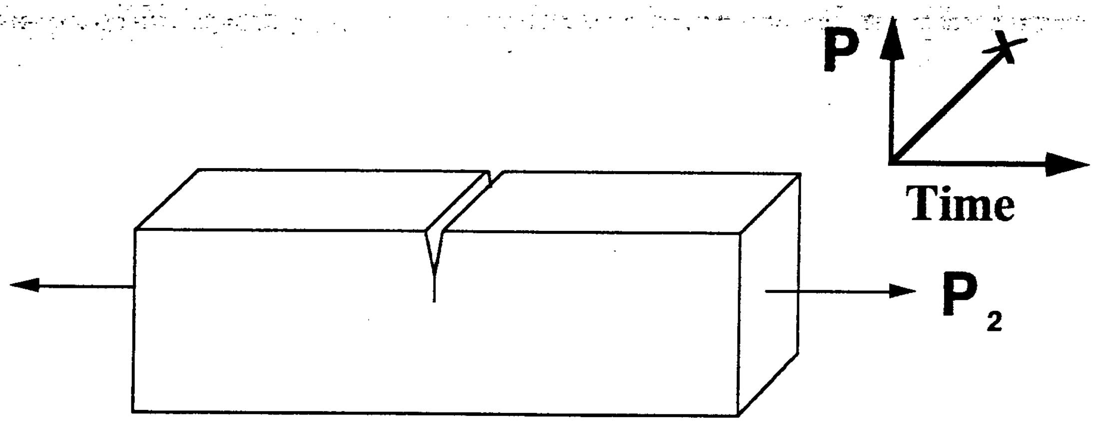
    i) lamellar plate thickness

    ii) lamellar colony size

Design against failure

Accept materials contain sharp defects





At failure $P_1 < P_2$ (easy to understand)

But ask how much can $P_2 > P_1$ and still be safe?

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Simple Analysis (Fracture Mechanics)

$$K = \sigma \quad (\pi a)^{1/2}$$

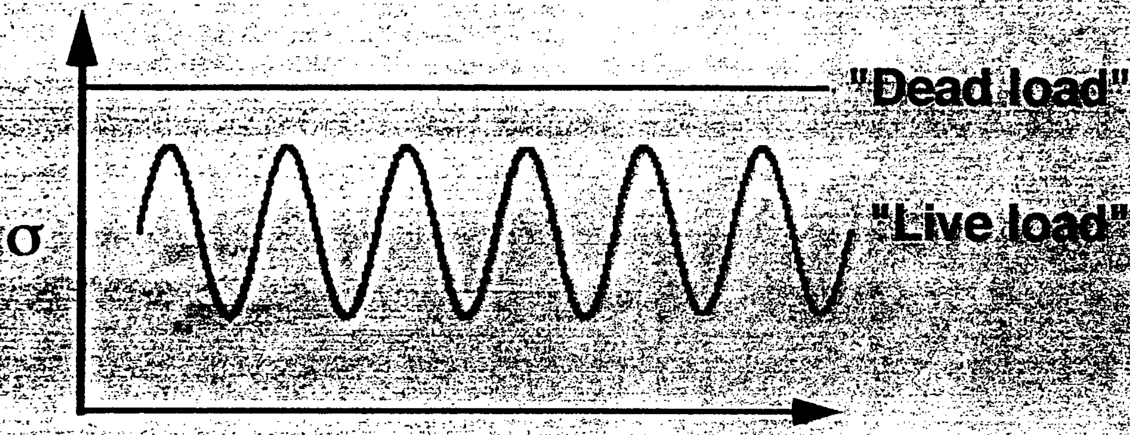
Material Failure Limit	**Applied Stress**	**Defect size, life defined by rate of growth**

Engineering : predict value of K for range of crack sizes, shapes and stress fields

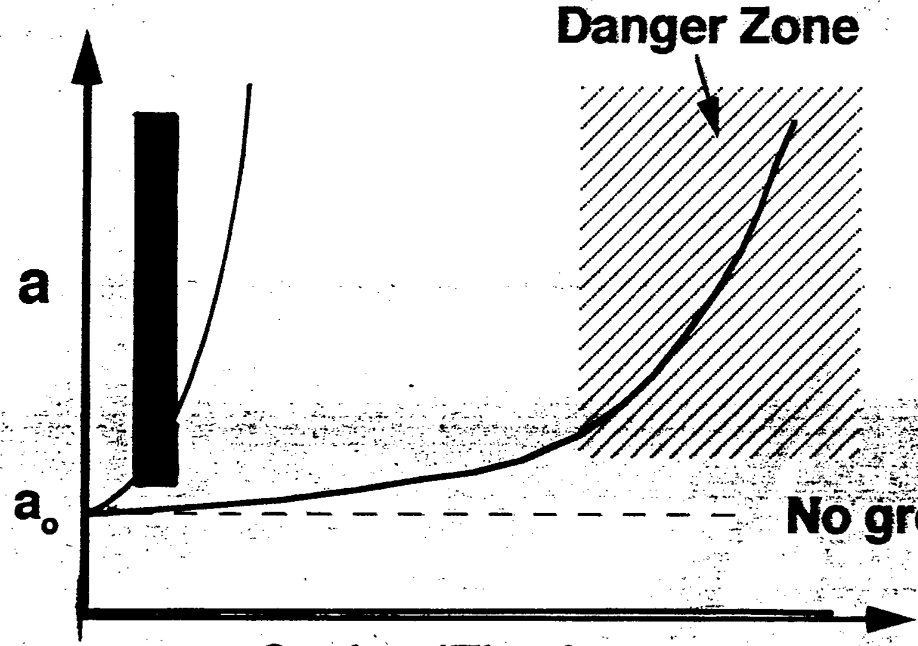
Metallurgy/Materials Science : control $K_{failure}$
(Understand microstructural size scale)

Failure : Brittle
Ductile

Fatigue loading: "Metal Disease"



→ Failure only under live load



a_o = initial defect population

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Simple Crack Growth Laws:

$$da/_{dN} = A \Delta K^m$$

Rate of crack growth

Driving force

A, m are material's "constants"

Conventional materials m = 2 - 4

Intermetallics, ceramics m upto 50

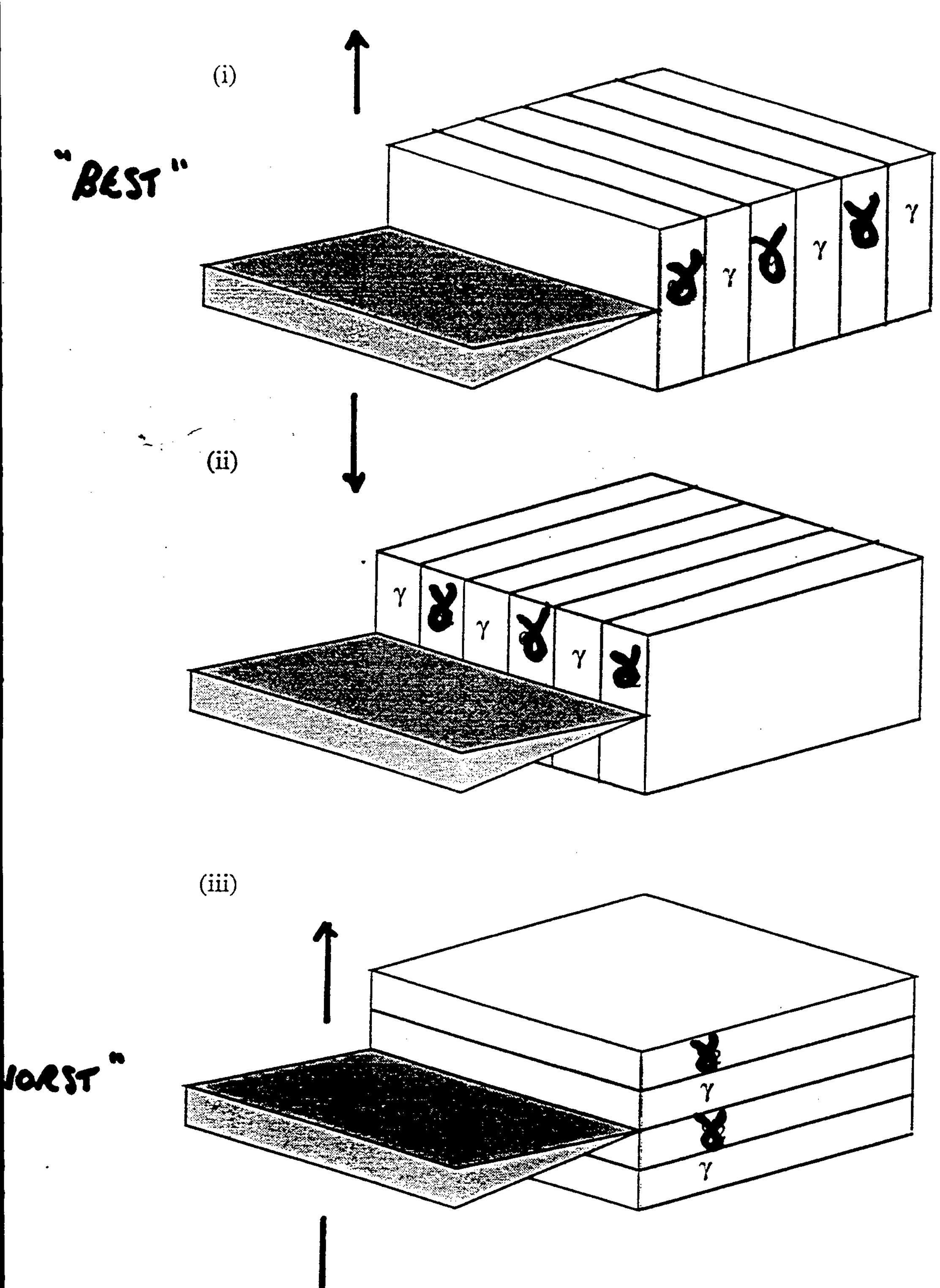
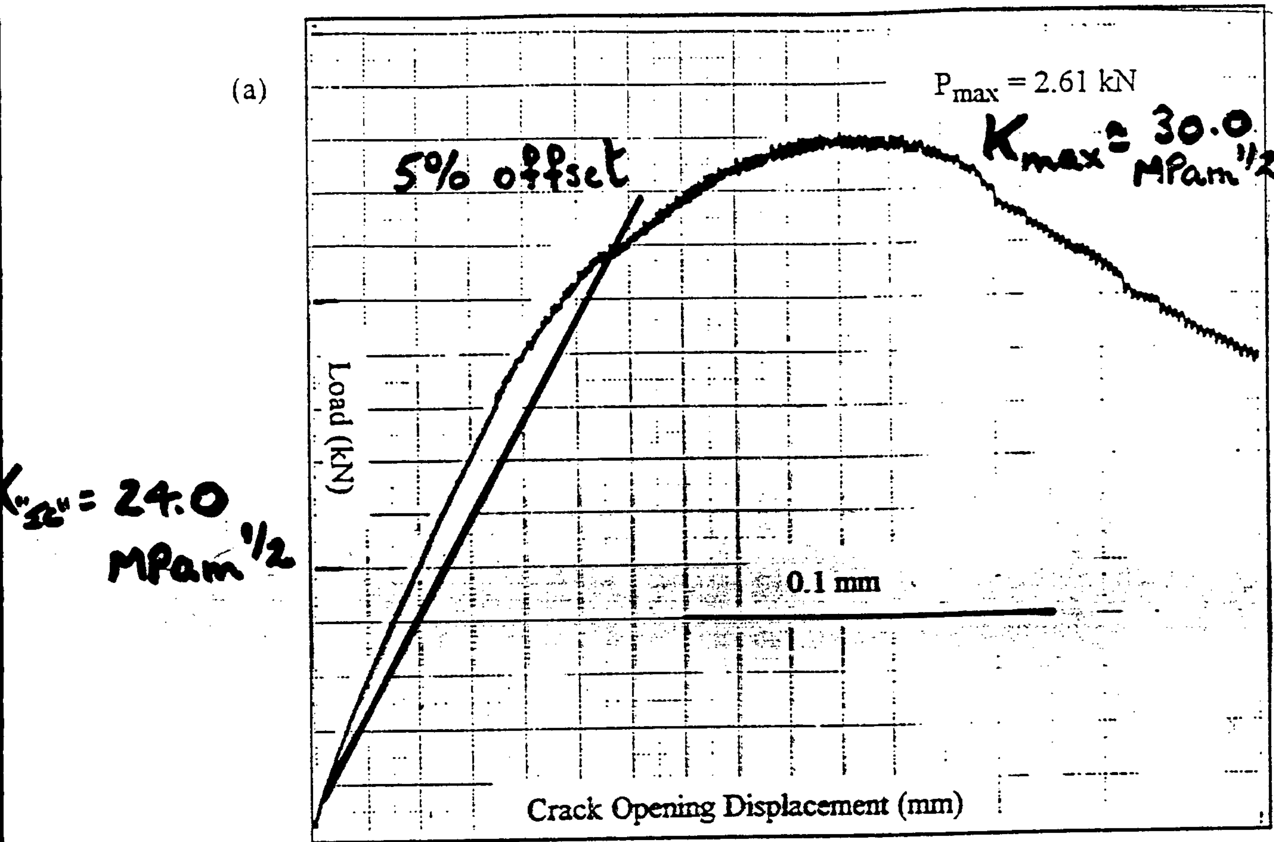


Figure 5.1(d)

Orientation relationship between the crack and the lamellar microstructure.
(i) Crack arrester orientation, (ii) Crack divider orientation
and (iii) crack delamination orientation.

(a)



(b)

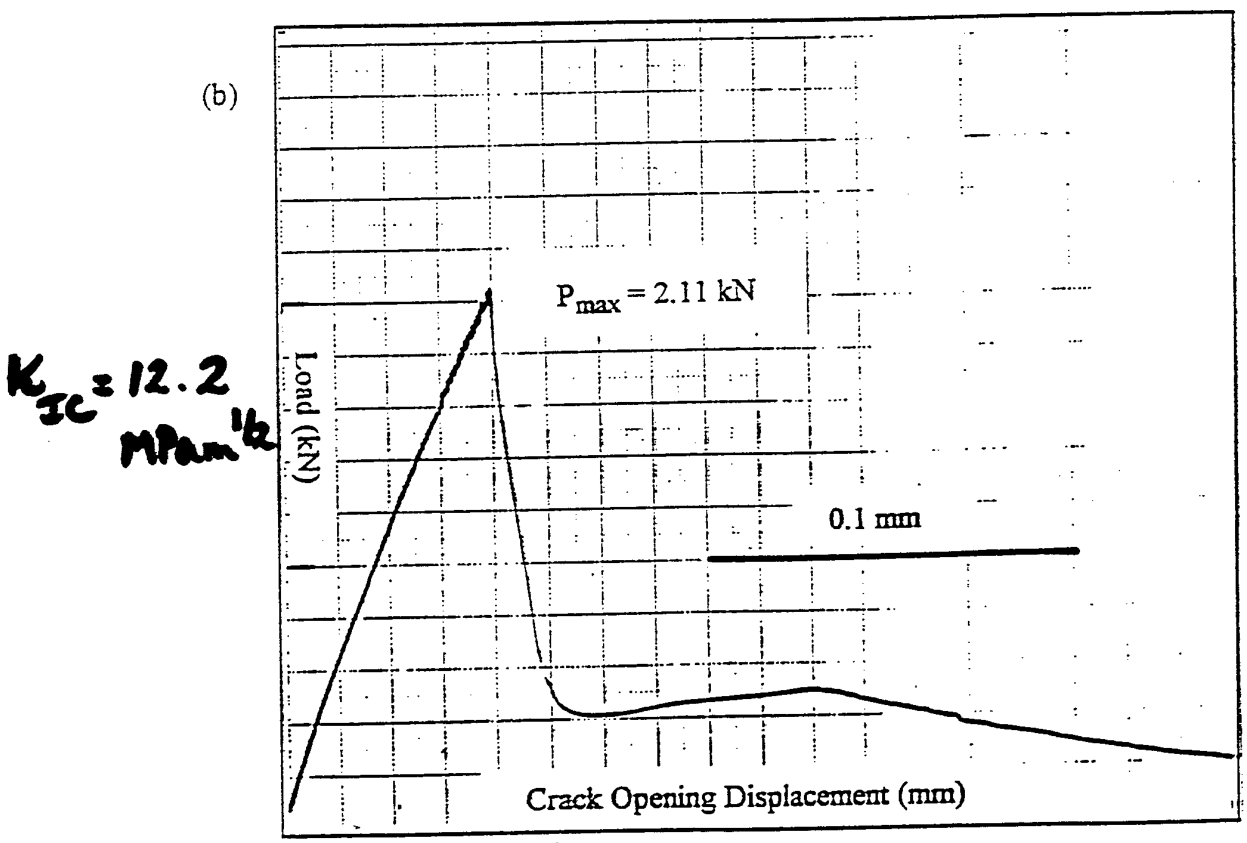
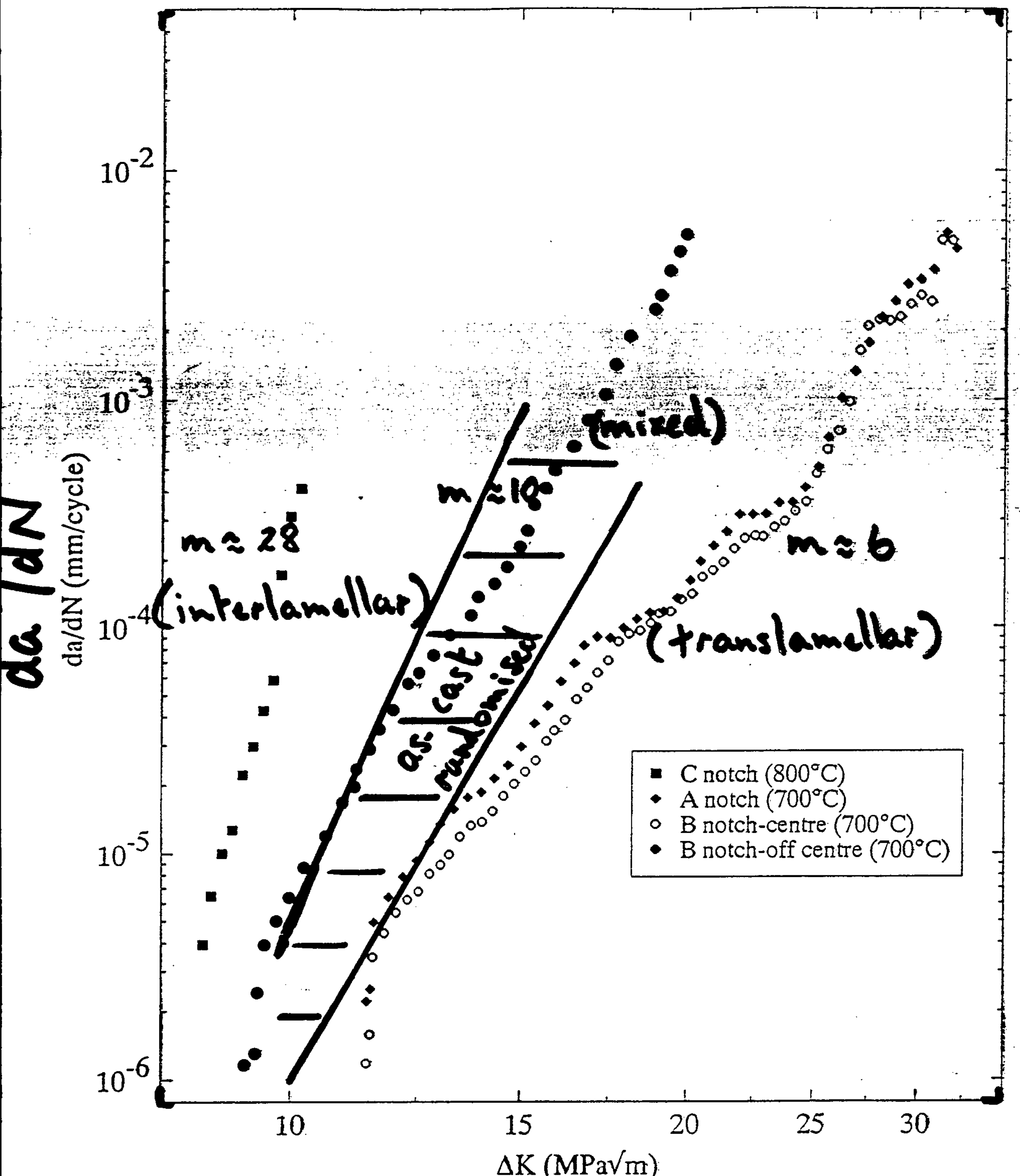


Figure 6.6

Clip gauge opening displacement versus load traces for fracture toughness tests performed in air at ambient temperature.
(a) translamellar failure and (b) interlamellar decohesion.

Ti-48Al - vacuum



ΔK

$$da/dN = A\,\Delta K^m$$

Effect of environment, 700°C, 10Hz

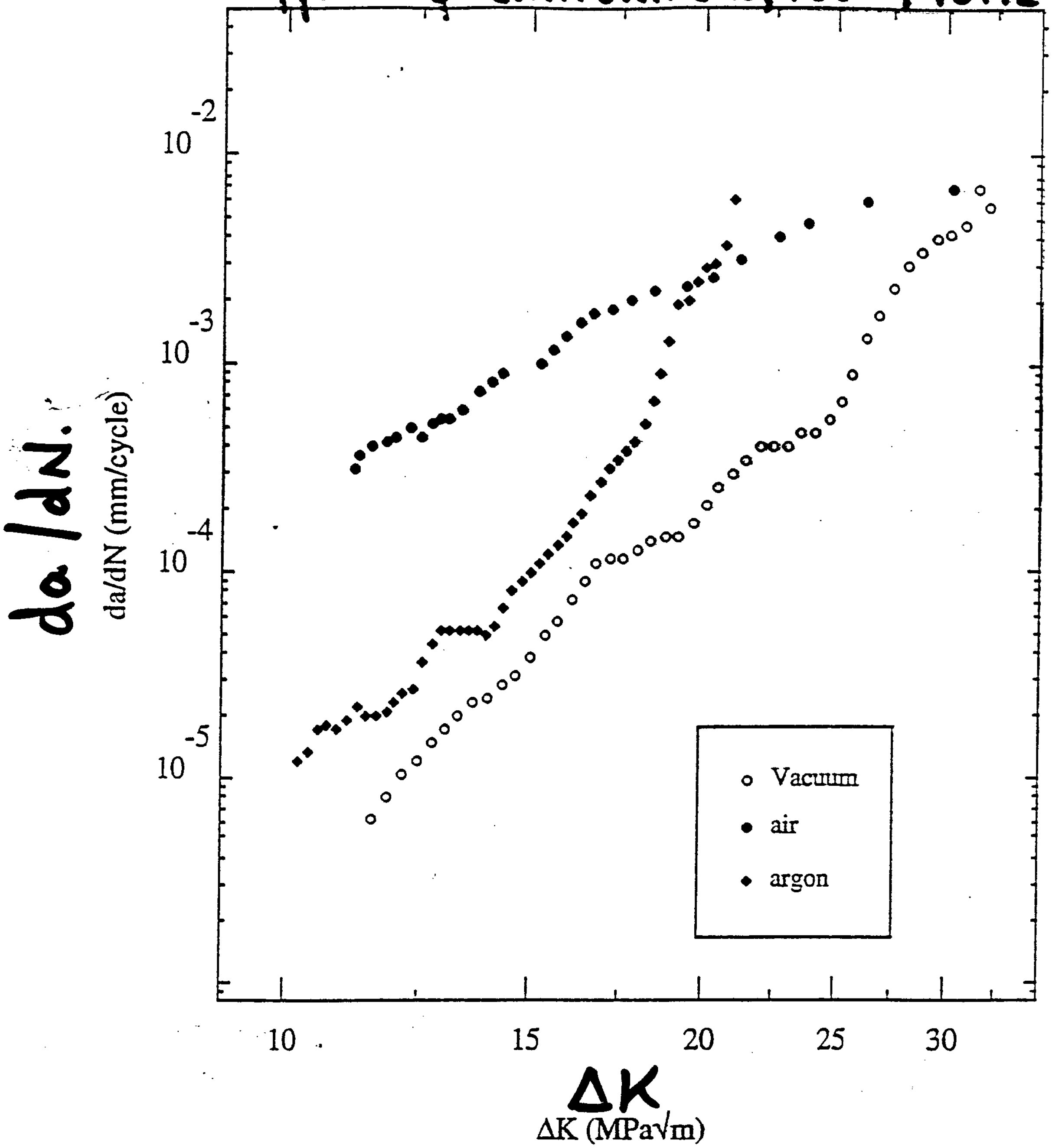
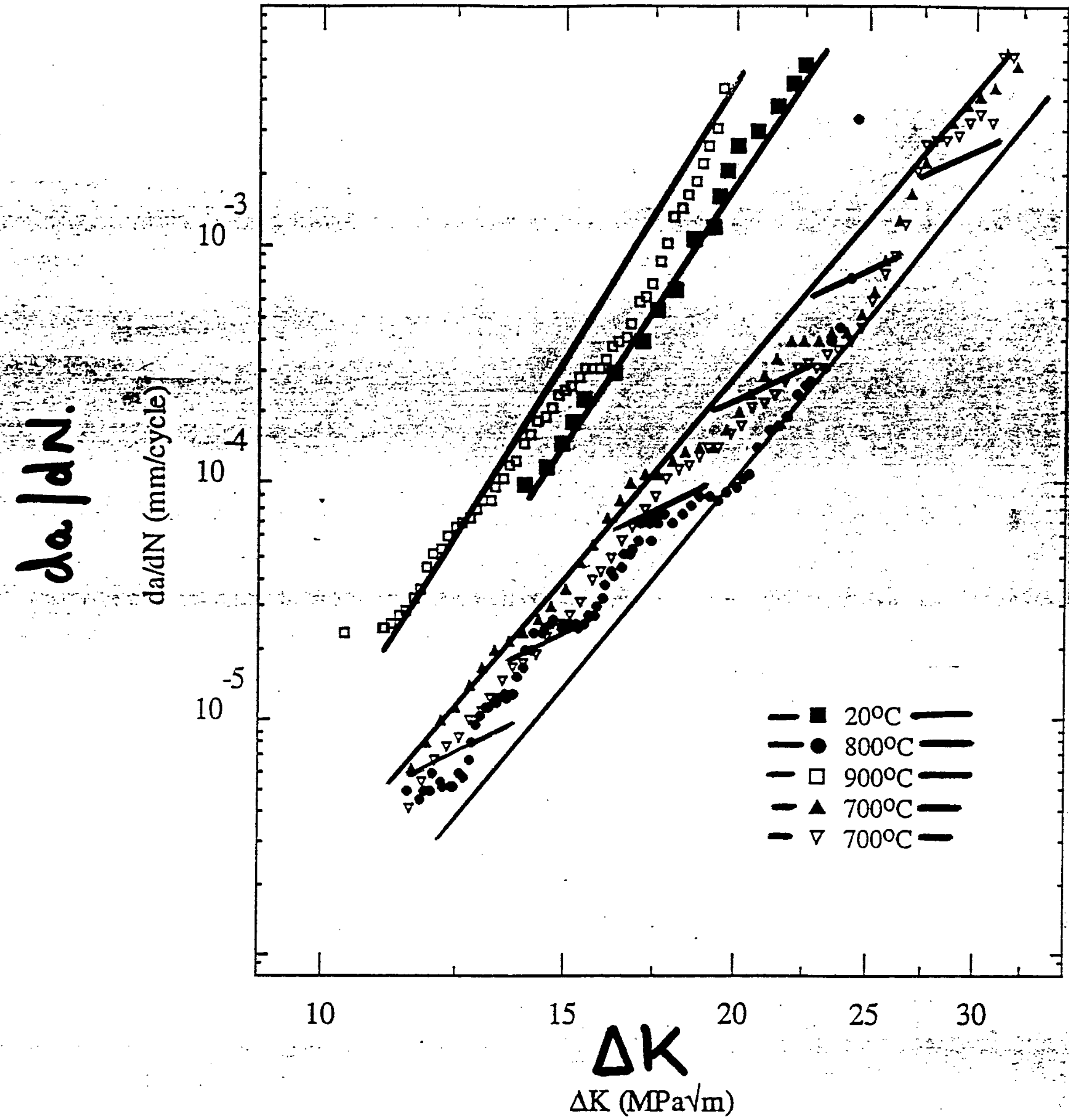


Figure 6.8

Fatigue crack growth resistance curves - da/dN versus ΔK for 'as cast' Ti-48Al tested at a temperature of 700°C as a function of environment.

FULLY TRANS LAMELLAR FAILURE

fects of test temperature (in vacuum).



(* indicates transverse testpiece with type (ii) notch)

Figure 6.9

Fatigue crack growth resistance curves - da/dN versus ΔK for 'as cast' Ti-48Al tested in vacuum at temperatures of 700, 800 and 900°C.

(Unusual trend compared with conventional alloys).

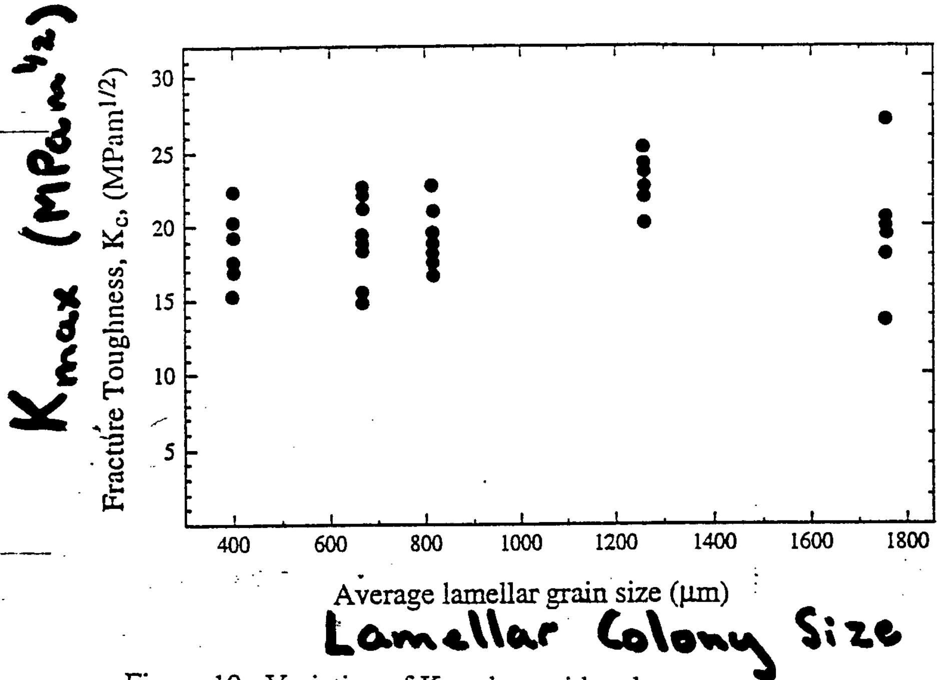


Figure 10. Variation of K_c values with colony size (fully lamellar microstructure-randomised colonies).

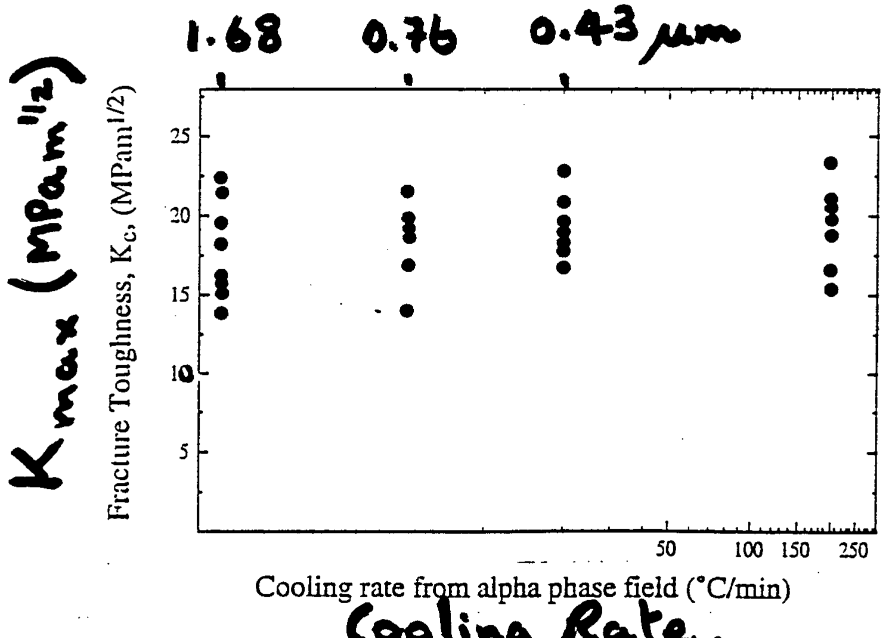
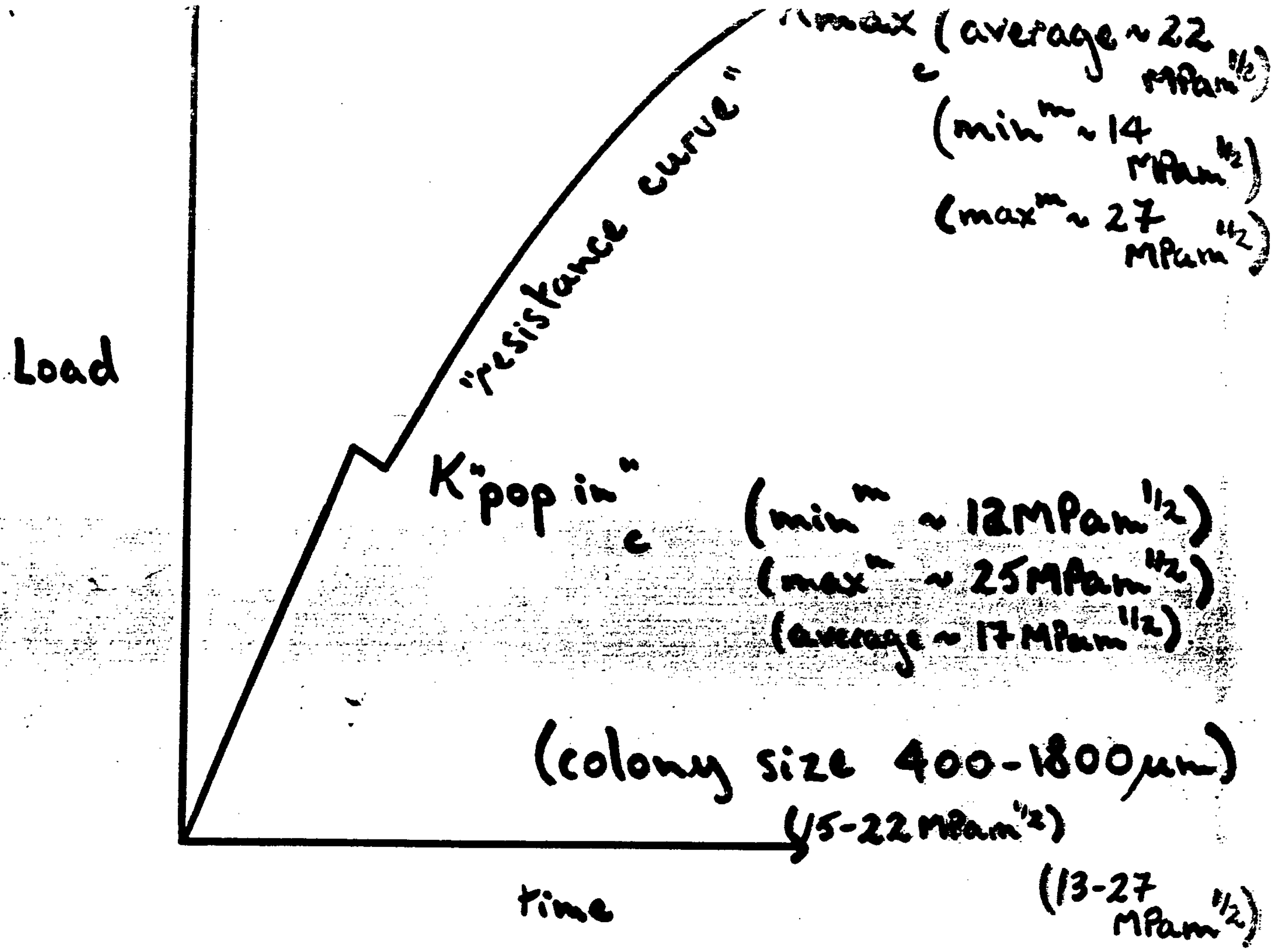


Figure 11. Variation of K_c values with plate thickness (fully lamellar microstructure-randomised colonies), see text.

ffects of lamellar colony size and lamellar plate thickness.

6q



K_{pop} min^m ~ lower bound for interlamellar aligned testpiece (12.2 MPam$^{1/2}$)

Use of "resistance curve"

Engineering sense

Use K_{pop} values even in "ductile" systems.

Need to ensure that microcracks do not join up

Need to utilise fatigue crack growth to

generate resistance curve.

or above system variation in K_{max} and K_{pop} is

xtreme for engineering alloy? (Weibull modulus ~ 7.8).

Process zone sampling effects at failure

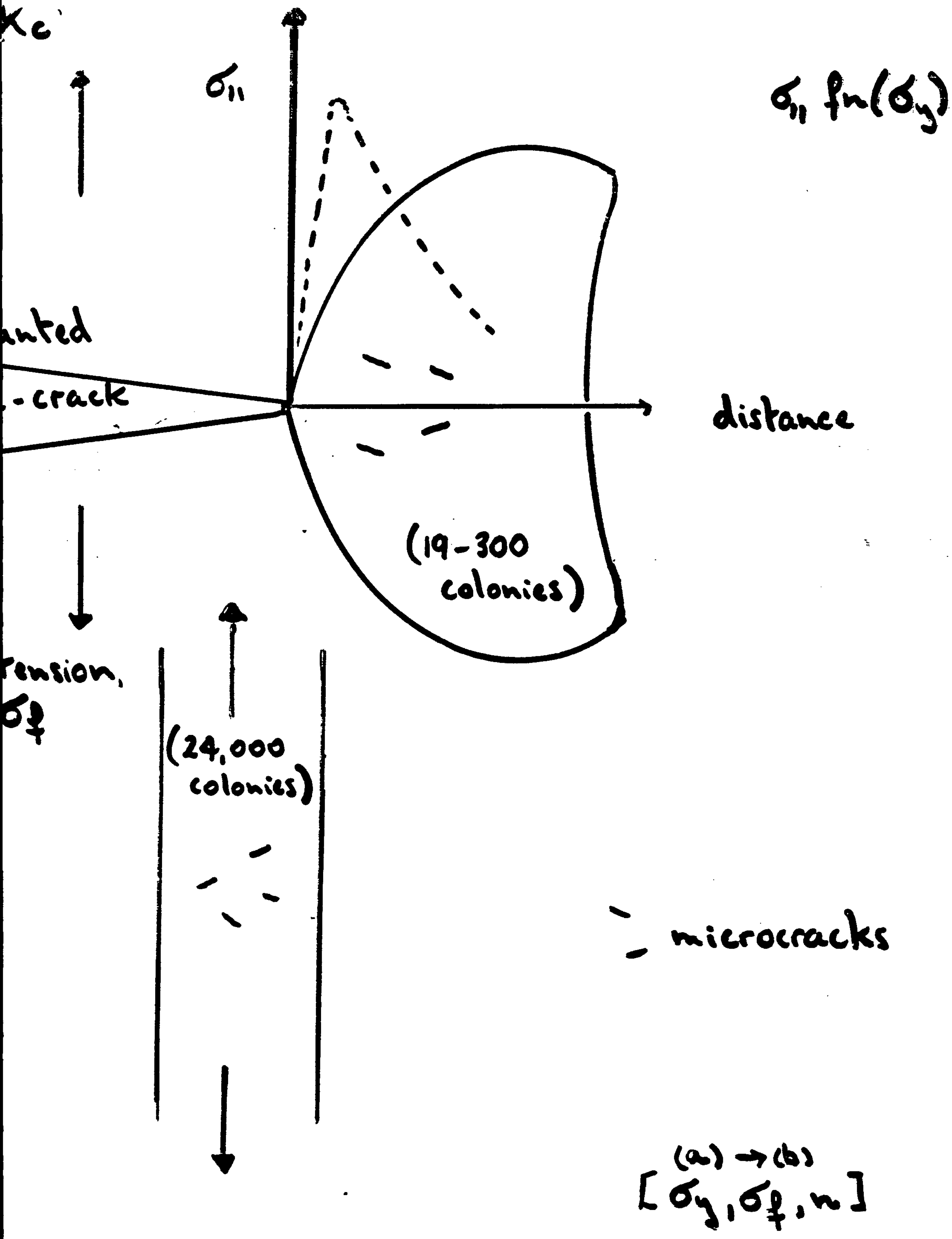
[stresses > σ_f (tension) over volume sampled).
(lamellar colony ~ 400μm)

Fracture toughness test	No of "grains"	σ_{11} (MPa)
$K_c = 13 MPa m^{1/2}$	19	1250-300
$K_c = 25 MPa m^{1/2}$	300	1250-300

Tensile test	No. of "grains"	σ_{11} (MPa)
$\sigma_f \sim 300 MPa$	24,000	300

NB.
For coarse "randomised" (400μm+) fully lamellar, % El to failure and lower bound K_c value, are both lower than values obtainable from duplex microstructures.

Creep resistance of fully-lamellar microstructures still vastly superior to duplex microstructures ?.



Consider just prior to failure.

18. Steep da/dN curve (high "m") ⇒ must design on total life (S-N curves).

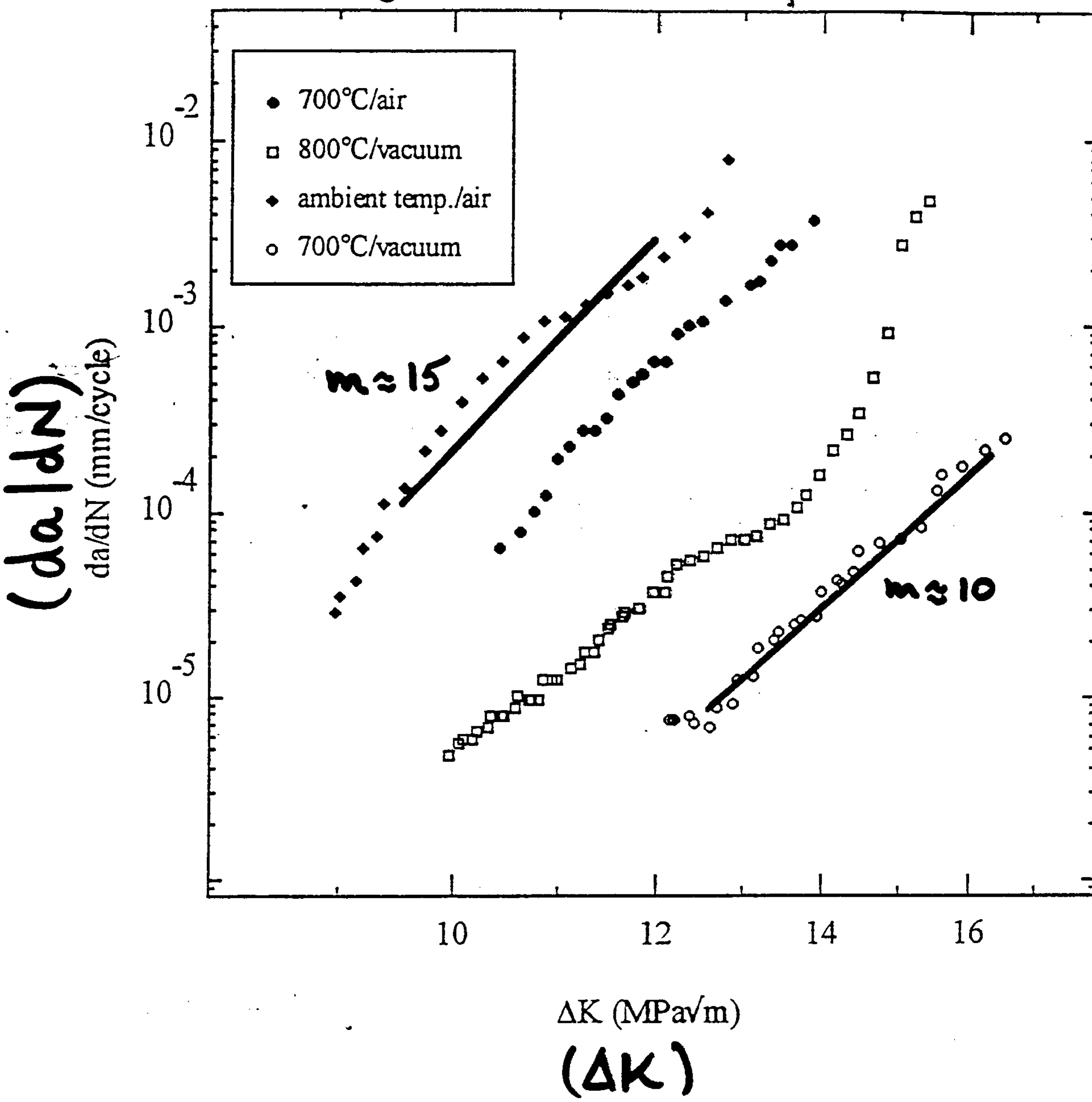


Figure 6.27

Fatigue crack growth curves da/dN versus ΔK for cast and HIPed XD™ gamma.

$\sigma_{0.2}$ = 340 MPa (327-346 MPa)

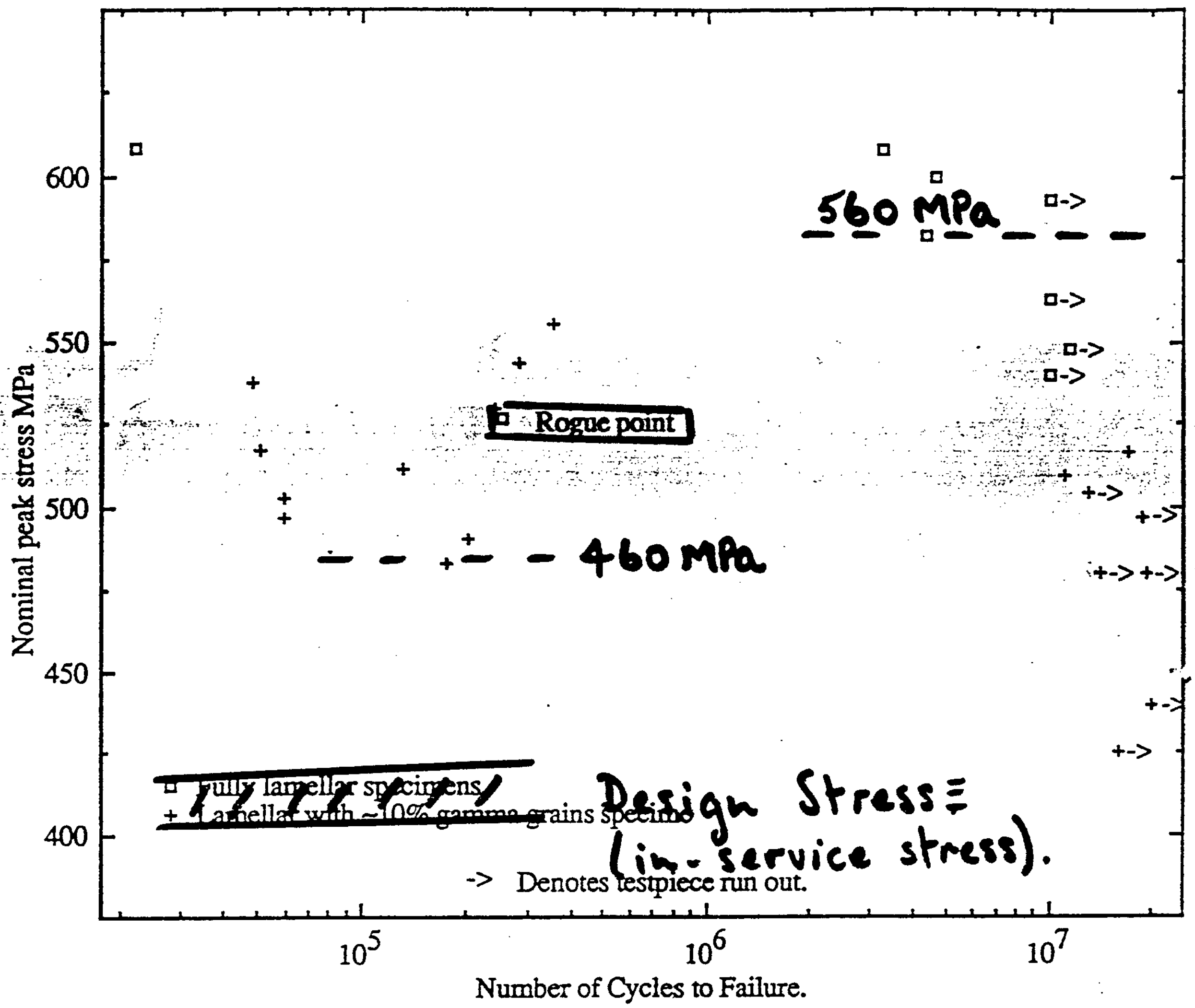
σ_f = 400 MPa (383-415 MPa)

ε_f = 1% (0.88-1.10%)

E = 160 GPa

K_c = 15-18 $MPa\,m^{1/2}$

N.B. Steep da/dN v ΔK curves ⇒ flat S-N curves.



S-N plots for ground and polished fully lamellar and near lamellar specimens.

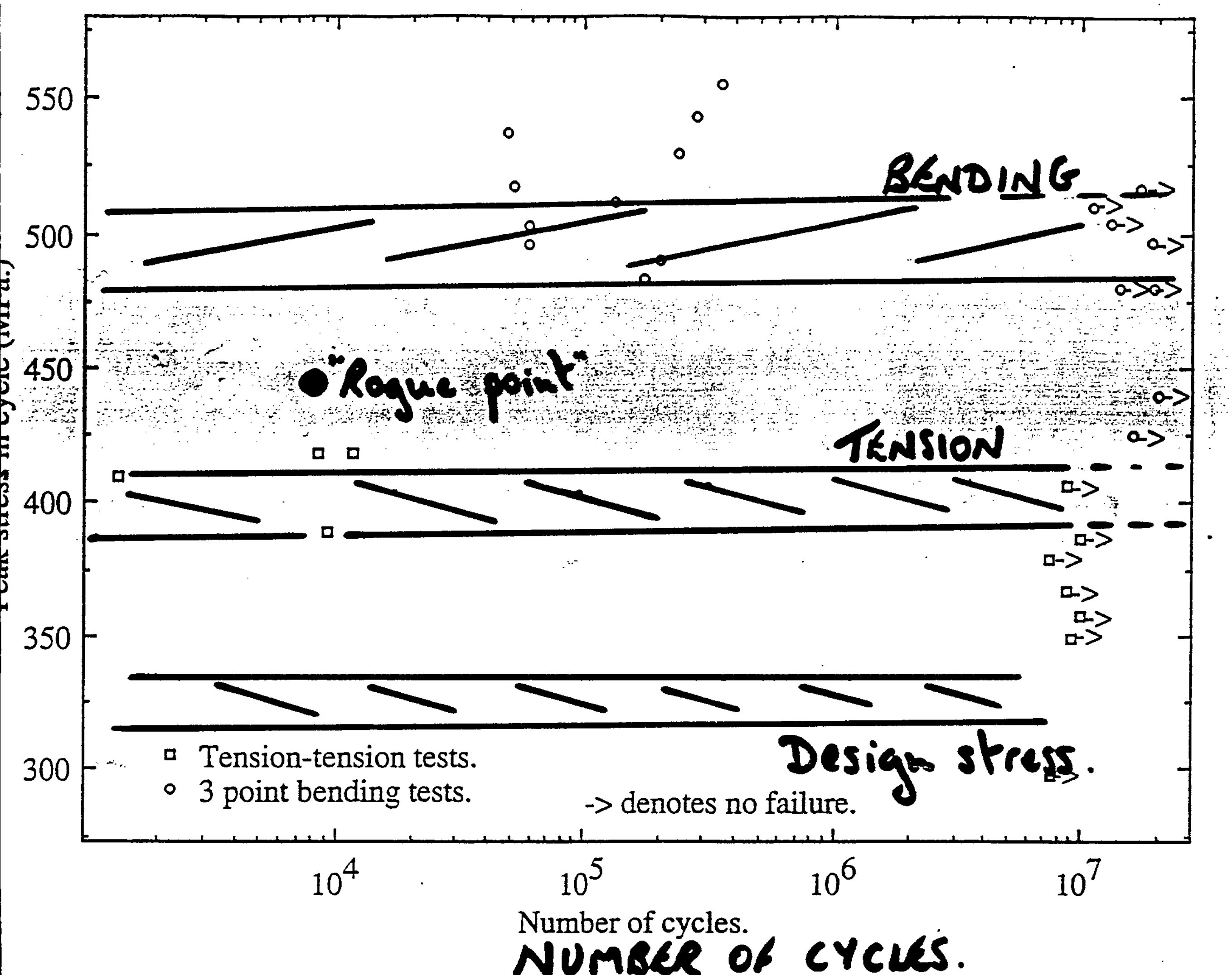
RUN OUT STRESS FOR FULLY LAMELLAR SPECIMENS = 560 MPa.*

RUN OUT STRESS FOR LAMELLAR+10% GAMMA GRAINS = 460 MPa.

* care because of "rogue point".

Steep $da/dN \Rightarrow$ flat S-N "curves"

Tension-Tension and 3 Point Bending



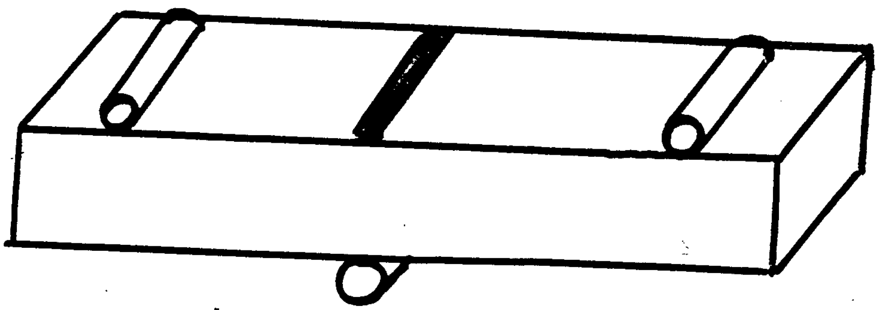
Run out stress for 3 point bending = 460MPa.
Run out stress for tension-tension = 380MPa.
($\sim 0.95\sigma_f$).

0.2% Proof Stress = 340MPa.
Tensile Strength = 410MPa.
(σ_f)

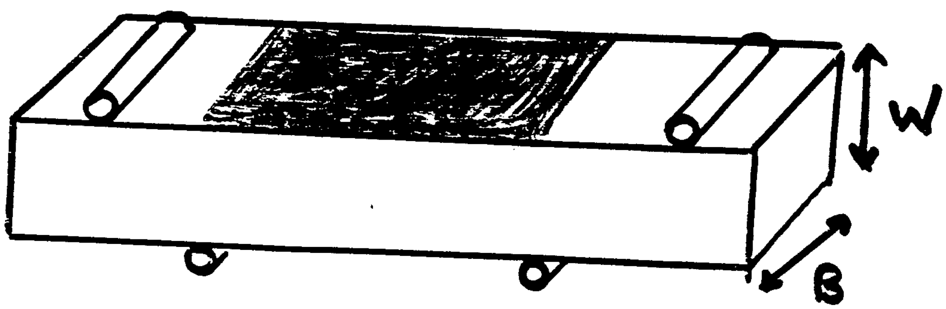
{ Surface area tension is x10 that in bending
Volume tension is x26 that in bending }

$\sigma_{surface} = 6M / W^2B$ $\quad$ $M = PS/2$

t bend

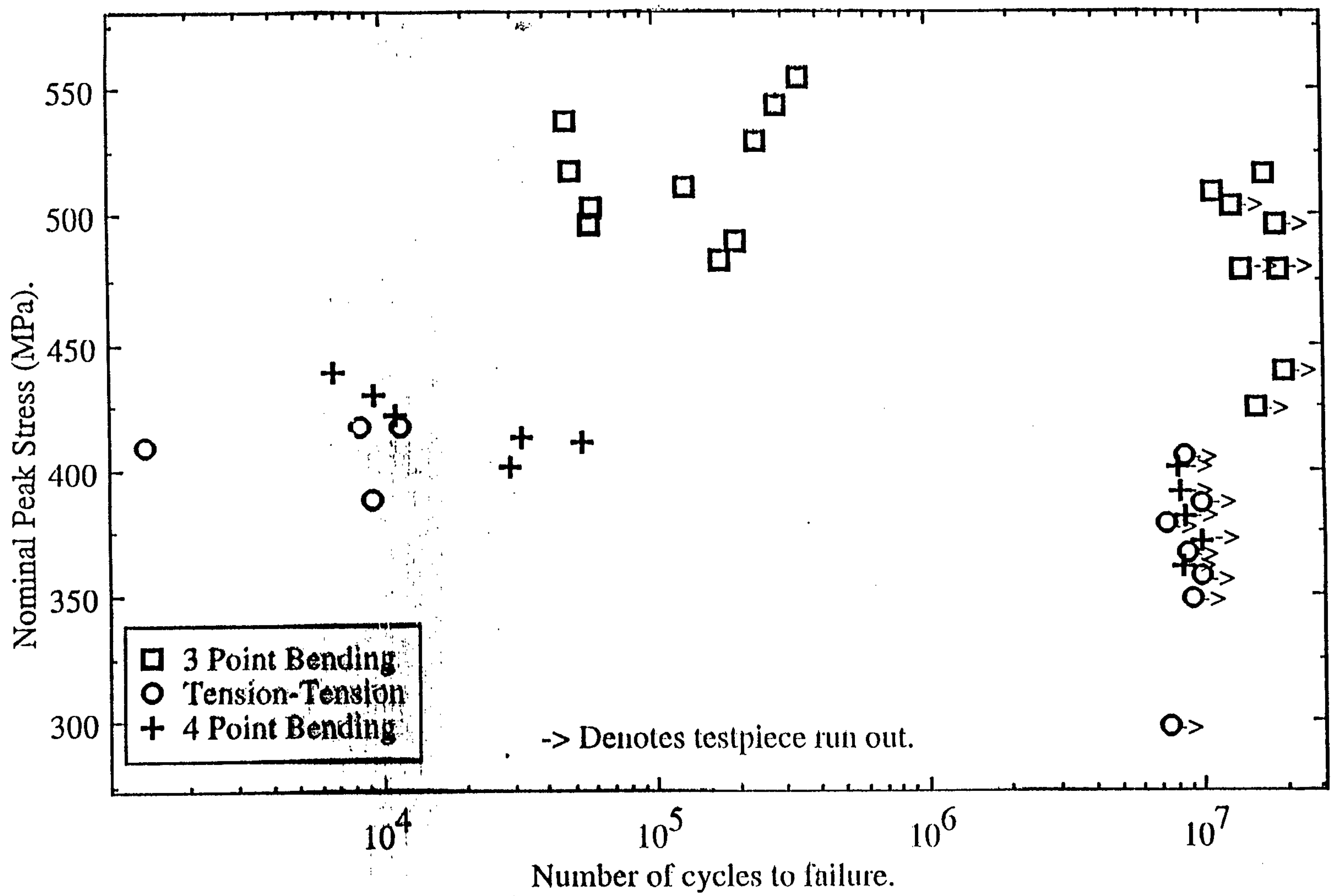


bend



ension $= \pi D \times$ gauge length

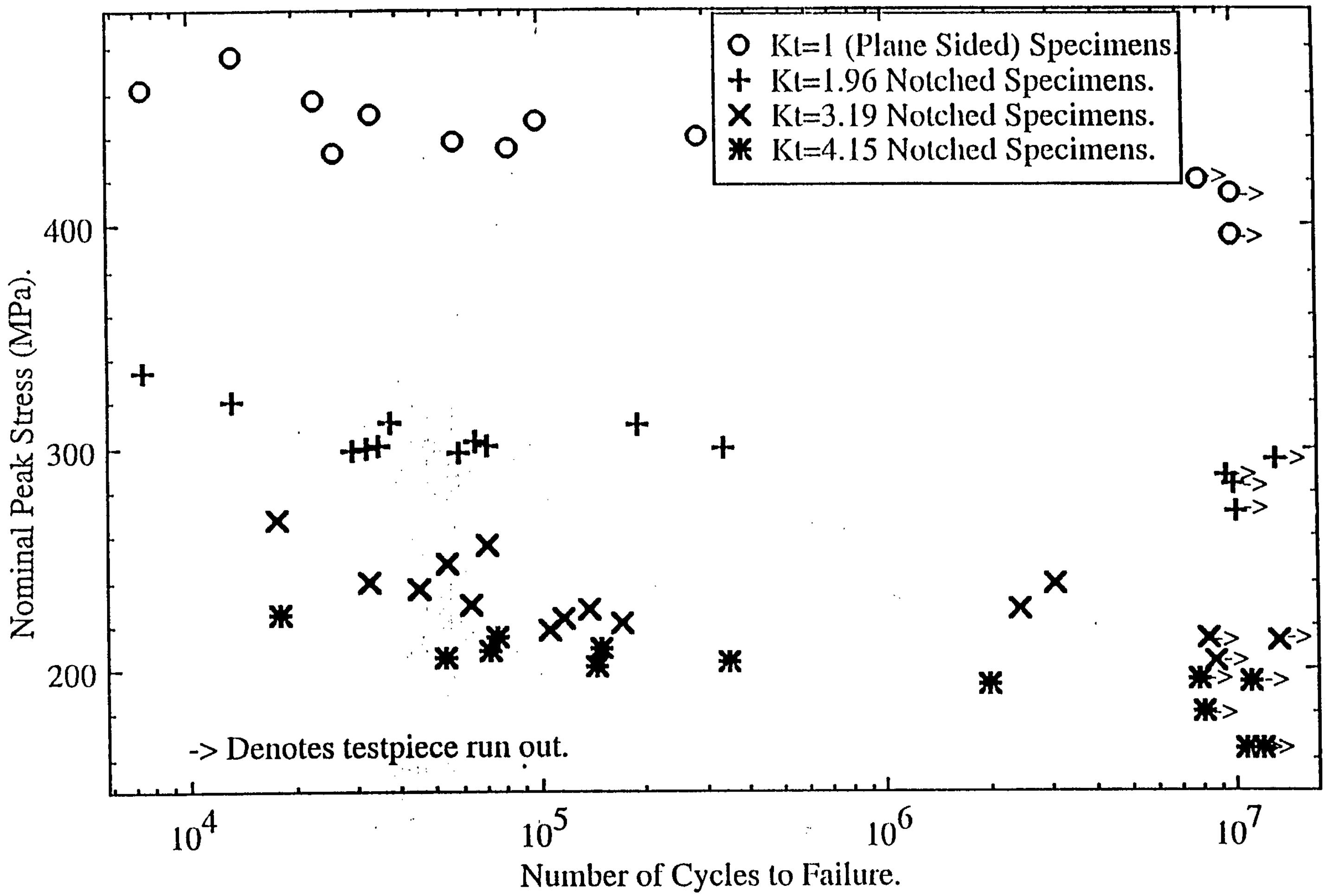
Match for $\sigma_{surface} \geqslant \sigma_{run\text{-}out}$ in tension?.



S-N plots of as-received ingot specimens tested in three and four point bending and tension-tension loading.

Nominal peak stress vs. number of cycles to failure.

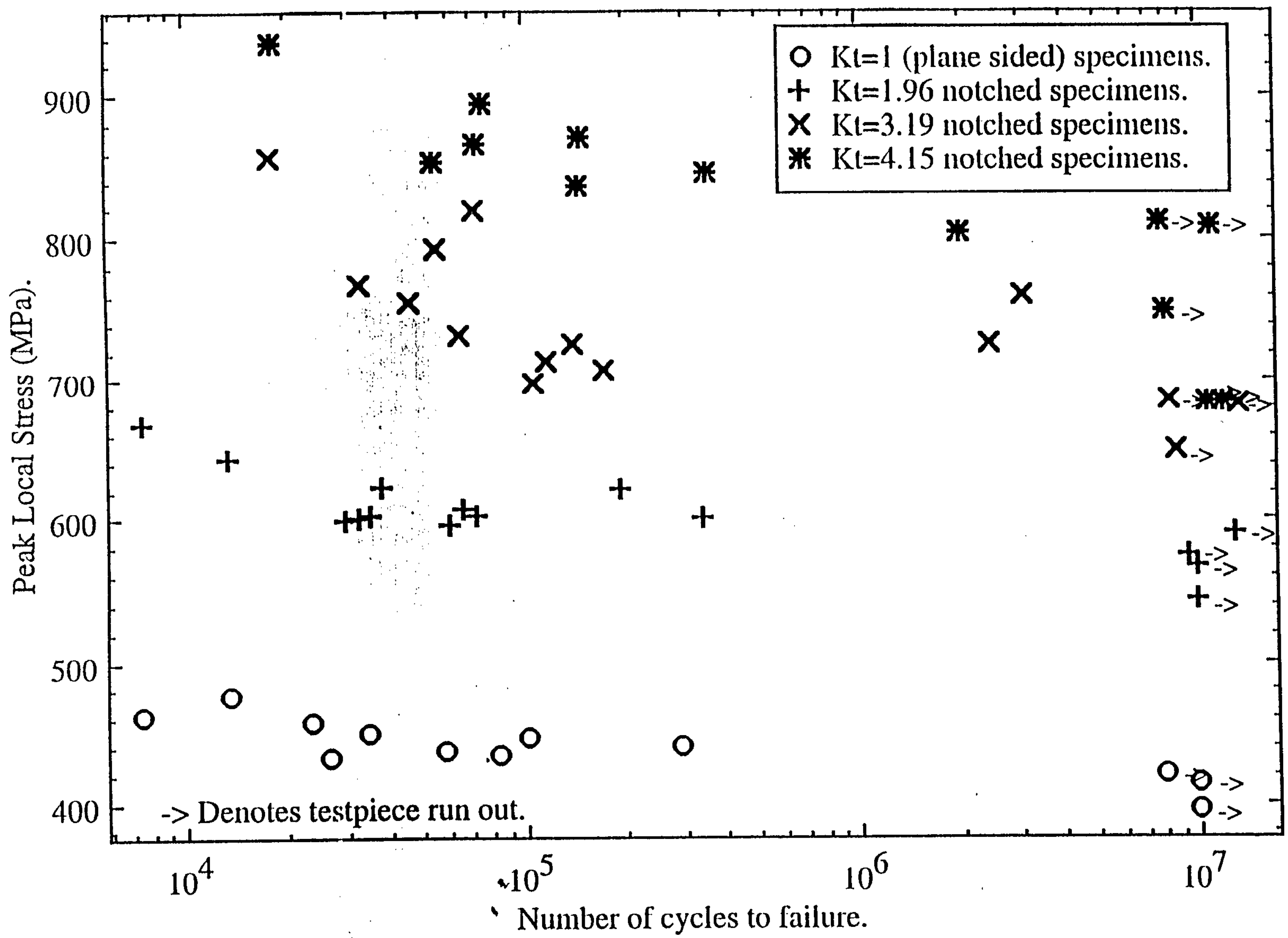
Figure 6.24



S-N plots of notched and plane-sided specimens machined with three passes of the EDM wire.

Nominal peak stress vs. number of cycles to failure.

Figure 6.18



S-N plots of notched and plane sided specimens machined with three passes of the EDM wire.

Peak local stress vs. number of cycles to failure.

Figure 6.19

Conclusions

Strong microstructural effects on da/dN vs ΔK and K_c (brittle fracture).

Fully lamellar microstructures show promise. Care required because K_c and da/dN vs ΔK can be highly anisotropic in as-cast microstructures.

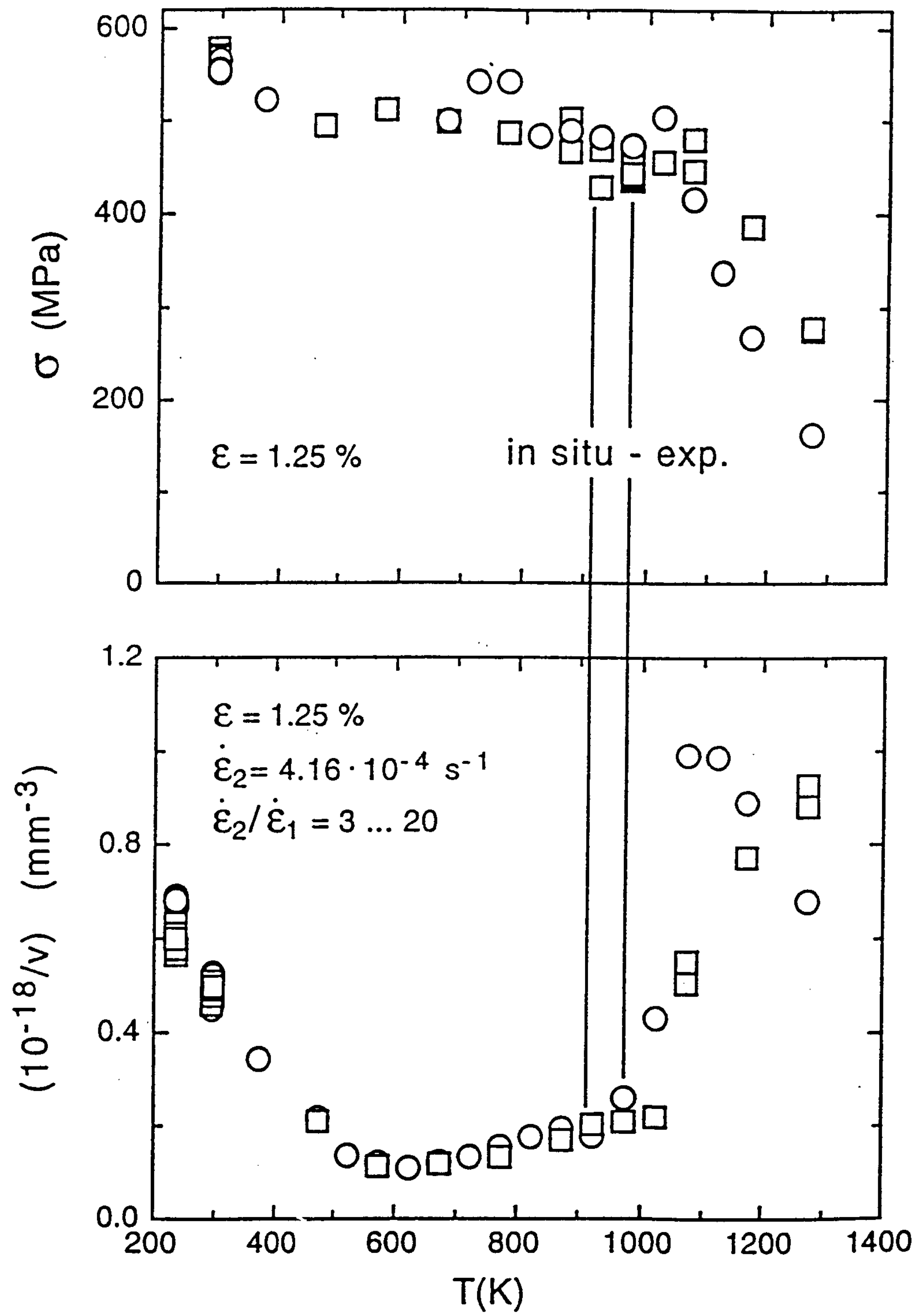
Optimisation of fully lamellar, "fine" colony size required (randomised). Even then steep da/dN vs ΔK, modest K_c ($15 MPa m^{1/2}$).

Use S-N approach but keep in-service stresses as low as possible (? notches).

Sampling arguments (stress and volume) need careful consideration.

In situ heating experiment

- relationship to strength properties



in situ study, T = 900 - 970 K