Fundamental research issues for understanding the emerging class of Gamma Titanium Aluminide Alloy Technologies

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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
Contents

1. Foreword/Summary

2. Pre-conference literature, including workshop programme

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):

i) The University Conference Park

   single/double..................................................

   30 April 1 May 2 May

   □ □ □

ii) A local hotel

   30 April 1 May 2 May

   □ □ □

   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 Microstructural development in gamma TiAl alloys containing dispersoids of Ti B2 - 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

\[ \dot{\varepsilon} = f \cdot \rho_m \cdot v_d = f \cdot \rho_m \cdot v_0 \cdot \exp \left( -\frac{\Delta G}{kT} \right) \]

- present study

- mobilities and multiplication of dislocations
- structural stability
- metallurgical factors affecting high-temperature strength
### Alloy compositions and microstructure

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

concentration of interstitial elements in the starting material

\[
\begin{align*}
N_2 & : & 100 - 200 \text{ wt. ppm} \\
O_2 & : & 500 - 700 \text{ wt. ppm} \\
C & : & 100 - 200 \text{ wt. ppm}
\end{align*}
\]

- 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

\[ \sigma_{\text{HP}} = \sigma_0 + K_y d^{-1/2} \quad , \quad K_y = 0.9\ldots1.5 \text{ MPa m}^{1/2} \]

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
  \[ v_D = v_o \exp \left[ -\frac{\Delta G (\tau^*)}{kT} \right] \]

- strain rate
  \[ \dot{a} = \rho_D b v_D = \dot{a}_o \exp \left[ -\Delta G (\tau^*) / kT \right] \]
Overcoming of glide obstacles with the aid of thermal activation model

\[ \nu_D = \nu_o \exp \left[ -\frac{\Delta G (\tau^*)}{kT} \right] \]

strain rate
\[ \dot{\varepsilon} = \rho_D b \nu_D = \dot{\varepsilon}_o \exp \left[ -\frac{\Delta G (\tau^*)}{kT} \right] \]

\( \tau^* \) 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 diagram]

- activation parameters:

  \[ V = \text{lb} \Delta R \]
  \[ \Delta F^* = \Delta G + V\tau^* \]
  \[ \tau^* = (1/V) (\Delta F^* + kT \ln \dot{\alpha}/\dot{\alpha}_0) \]

- 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:

\[ (\Delta \sigma / \Delta \ln \dot{\epsilon})_T \]
strain rate cycling tests

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

\[ (\Delta \sigma / \Delta T)_\dot{\epsilon} \]
temperature cycling tests

determination of these parameters as function of \( \sigma \), \( T \) and \( \epsilon \)

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

\[ \sigma \text{ (MPa)} \]

\( \dot{\varepsilon} = 1.25 \% \)

\( \dot{\varepsilon} = 4.16 \cdot 10^{-4} \text{ s}^{-1} \)

\[ \frac{\dot{\varepsilon}_2}{\dot{\varepsilon}_1} = 3 \ldots 20 \]

\[ (10^{-18} v) \text{ (mm}^{-3} \text{)} \]

\[ T(\text{K}) \]

I  II  III  IV

Interpretation region I
Athermal stress parts

\[ \sigma_\mu = \sigma_{\text{dis}} + \sigma_{\text{HP}} \]

\( \sigma_{\text{dis}} \): long-range interaction of dislocations

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

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

\( \sigma_{\text{HP}} \): interaction of dislocations with grain boundaries and lamellar interfaces

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

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

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

near gamma microstructure

\( T = 300 \text{ K}, \ \varepsilon = 1.25\%, \ \dot{\varepsilon} = 4.16 \times 10^{-4} \text{ s}^{-1} \)

\( \sigma = 550 \text{ MPa} \)
\( \sigma_\mu = 430 \text{ MPa} \)

\( \tau^* = 40 \text{ MPa} \)
\( V = 95 \text{ b}^3, \quad b = 1/2 <110> \]
\( V \tau^* = 0.5 \text{ eV} \)
\( \Delta G = 0.8 \text{ eV} \)
\( \Delta F^* = \Delta G + V \tau^* = 1.3 \text{ eV} \)

- low dislocation mobility

\[\downarrow\]

TEM-observations
Pinning of $1/2 <110>$ screw dislocations by localized obstacles and jogs. (CM 3938)
Conclusions

- The estimated activation parameters
  \[ \tau^* = 40 \text{ MPa}, \ V = 95 \text{ b}^3, \ G = 0.8 \text{ eV}, \ F^* = 1.3 \text{ eV} \]
  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

σ (MPa)

ε = 1.25 %
ε = 4.16 \cdot 10^{-4} \text{ s}^{-1}

\frac{\dot{\varepsilon}_2}{\dot{\varepsilon}_1} = 3 \ldots 20

\left(10^{-18}/\nu\right) \text{ (mm}^{-3}\text{)}

T(K)

I  II  III  IV
Temperature dependence of the flow stress

Region II: $T = 450 \ldots 700\; K$

Yield drop effects and serrations in a narrow temperature interval,

dependence on strain rate $\rightarrow$ load elongation traces

formation of impurity atmospheres, further investigations
Static strain ageing

load - elongation trace

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

$\sigma = 100 \text{ MPa}$
$\varepsilon = 1\%$

$56 \text{ s}$
$450 \text{ s}$
$225 \text{ s}$
$112 \text{ s}$
$56 \text{ s}$

$\Delta \sigma$
Static strain ageing

experimental investigations

\[ T = 300, 423, 523, 623 \text{ K} \]

\[ \varepsilon = 4.16 \cdot 10^{-4} \text{ s}^{-1} \]

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

\[ \Delta \sigma = g (\sigma_a) \]

\[ \Delta \sigma = h (t_a) \]

\[ \Delta \sigma = u (\varepsilon) \]

\[ \downarrow \]

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 $t_a$

[Graph 1]

$T=523K$

$T=623K$

$\varepsilon=1.25\%$

$\dot{\varepsilon}=4.16\cdot10^{-4}\text{ s}^{-1}$

[Graph 2]

$T=523K$

$T=623K$

$T=423K$

$\varepsilon=1.25\%$

$\dot{\varepsilon}=4.16\cdot10^{-4}\text{ s}^{-1}$
Static strain ageing

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

dependence on ageing stress $\sigma_a$

![Graph showing dependence on ageing stress $\sigma_a$.]

dependence on strain $\varepsilon$

![Graph showing dependence on strain $\varepsilon$.]
Static strain ageing

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

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

<table>
<thead>
<tr>
<th>alloy</th>
<th>$\Delta G \text{ (eV)}$</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ti-47Al-2Cr-0.2Si</td>
<td>0.47</td>
</tr>
<tr>
<td>Ti-49Al-0.4C</td>
<td>0.55</td>
</tr>
</tbody>
</table>
Flow stresses and activation volumes

\[ \tau^* = (1/V) \left( \Delta F^* + kT \ln \frac{\dot{a}}{\dot{a}_0} \right) \]

Ti-47 at.% Al-2 at.% Cr-0.2 at.% Si
○ near gamma, □ nearly-lamellar

\[ \varepsilon = 1.25 \% \]
\[ \dot{\varepsilon} = 4.16 \cdot 10^{-4} \text{ s}^{-1} \]

\[ \dot{\varepsilon}_2/\dot{\varepsilon}_1 = 3 \ldots 20 \]

activation parameters
Activation parameters

Region III: 700...1000 K

Strong increase of $1/V$ with $T$

$$\tau^* = \frac{1}{V} (\Delta F^* + kT \ln \frac{a}{a_0})$$

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?

Implications on high-temperature strength
**Activation parameters**

Region III: 700...1000 K

- strong increase of $1/V$ with $T$

\[ \tau^* = (1/V) (\Delta F^* + kT \ln \hat{a}/\hat{a}_0) \]

- nearly lamellar microstructure:
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**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?**

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

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

present situation: \( R = 10^4b, \ L = 10^2 - 10^3b \)

<table>
<thead>
<tr>
<th>( T ) (K)</th>
<th>( v ) (mm/s)</th>
<th>( D_s ) (m²/s)</th>
</tr>
</thead>
<tbody>
<tr>
<td>820</td>
<td>2.4 x 10⁻⁹</td>
<td>5 x 10⁻²¹</td>
</tr>
<tr>
<td>900</td>
<td>2 x 10⁻⁷</td>
<td>2 x 10⁻¹⁹</td>
</tr>
</tbody>
</table>

comparison: \( T = 1173 \) K

\( D_s = 10⁻¹⁷ - 10⁻¹⁵ \) m²/s
(Ouchi et al., Kroll et al.)
Evaluation

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

$$\ln \left( \frac{c}{c_0} \right) = \frac{\mu b \Omega}{L 2\pi(1-\nu) kT} \ln(\alpha / 1.8 b)$$

-present situation: $\alpha = 4$

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

$\frac{c}{c_0} = 3 - 1.7$

-small supersaturation in comparison with those met in rapid quenches ($\frac{c}{c_0} \sim 10^4$)

-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

\[ \gamma \text{ phase: domain structure of six ordered variants} \]

\[ \text{tetragonality of the } \gamma \text{ phase } \frac{c}{a} = 1.02 \]
HREM Observation of lamellar interfaces

Interfaces types:

$\alpha_2/\gamma$

$\gamma/\gamma_T$ true twin

$\gamma_1/\gamma_2$ matrix/matrix

$\gamma_1/\gamma_2$ pseudo-twin

$<101>$ projection of the pseudo-twin:

![Diagram of pseudo-twin with labels (002)$_\gamma$, (111)$_\gamma$, (111)$_\gamma$, and examples]
Interface between γ variants with true twin orientation; Ti-48at.%Al-2at.%Cr. (CM 4204)
Interface between $\gamma$ variants with matrix/matrix orientation; Ti48-at.%Al-2at.%Cr. (CM 4232)
α₂/γ interface in a Ti-48at.%Al-2at.%Cr alloy. Creep deformation: \( \sigma = 140 \text{ MPa}, T = 700 \text{ °C}, t = 6000 \text{ h}, \varepsilon = 0.69\% \). (CM 4278)
Stress state of lamellar interfaces

- lattice mismatch largely accommodated by misfit dislocations
- residual homogeneous straining of adjacent lamellae

long-range internal stresses $\tau$ at the interfaces
Origin of the residual stresses

- lattice mismatch at semicoherent interfaces
  \( \Delta \varepsilon = 1 \ldots 2 \% \)

- largely accommodated by misfit dislocations

- residual homogeneous straining \( \Delta \varepsilon \) of adjacent lamellae

- high elastic stiffness of \( \gamma \)-TiAl,
  \( \mu = 4.3 \cdot 10^4 \) MPa

long-range residual stresses at the interfaces: \( \tau = \Delta \varepsilon \cdot \mu \)
Origin of the residual stresses

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

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

schematic drawing and example
Stress state of lamellar interfaces

Coherency stresses at semicoherent (111)γ 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

\[ T = \left[ \frac{\mu b^2}{4\pi(1-v)} \right] \left[ \ln \left( \frac{R}{r_0} \right) + C(\Theta) \right] \]

\[ R = l/5, \quad r_0 = b/8 \quad \text{(Hirth, Lothe)} \]

\[ q = \left[ \frac{\mu b}{4\pi(1-v)} \tau \right] \left[ \ln 1 + \ln \left( \frac{8}{5b} \right) \right] \]

\[ \tau = \text{const.} \]

\[ q = E' \ln 1 + D \]

Evaluation:
Stress state of lamellar interfaces

Distribution of the internal stresses $\tau$ acting on dislocation loops emitted from interfacial boundaries.

Comparison with deformation experiments:

\[ \sigma_a = 430 \text{ MPa} \rightarrow \tau_a = \sigma_a/3 = 140 \text{ MPa} \]
\[ \tau = 20 \ldots 220 \text{ MPa} \]

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, 24.
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 $\gamma/\gamma_T$ between lamellae with true twin relation. $T = 300 \text{ K}$, $\varepsilon_f = 0.2\%$ (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


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

load-deflection trace

\[ K_{lc} = 15.2 \pm 2.3 \text{ MPa m}^{1/2} \]

\[ K_{lc} = 22.1 \pm 1.9 \text{ MPa m}^{1/2} \]

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, $\tau = 70 \ldots 290$ MPa
Crack propagation in a two-phase (α₂+γ) 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 (α₂+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.

Semicohrent \(\alpha_2/\gamma\) and \(\gamma/\gamma\) 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:
  \[ T = 700 \, ^\circ C, \sigma = 150 \, MPa, \, t = 10.000 \, h \rightarrow \varepsilon \leq 1\%, \]
  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

\[ \rightarrow \]

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

- in situ heating studies

- defect structure of samples crept at
  \[ T = 700 \, ^\circ C, \sigma = 150 \, MPa \, \text{for} \, 6000 \, h \rightarrow \varepsilon = 0.69\%
  \]
  nominal creep rate \( \dot{\varepsilon} = 3 \times 10^{-10}s^{-1} \)
In situ heating experiment
- relationship to strength properties

\[ \sigma \] (MPa)

\[ \varepsilon = 1.25 \% \]

in situ - exp.

\[ 10^{-18} / \nu \] (mm\(^3\))

\[ \varepsilon_2 = 4.16 \cdot 10^{-4} \text{ s}^{-1} \]

\[ \varepsilon_2 / \varepsilon_1 = 3 \ldots 20 \]

T(K)

in situ study, \( T = 900 - 970 \text{ K} \)
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 $\gamma$ grain within a lamellar colony of a Ti-48at.%Al-2at.%Cr alloy. Creep deformation:

$\sigma = 140$ MPa, $T = 700$ °C, $t = 6000$ h, $\varepsilon = 0.69\%$.

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

- [Graph showing σ_{1.25\%} (MPa) vs. C (wt ppm) with solutions at 296 K and 973 K.]

- [Graph showing 10^{-18}N (mm^{-3}) vs. C (wt ppm) with solutions.]

- 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$_3$AlC precipitates. Ti-49 at.% Al-0.4 at.% C. Deformation at 300 K to $\varepsilon = 3$ %. (CM 5845)
Interactions of deformation twins with Ti$_3$AlC 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 Ti₃AlC 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?
Simple Analysis (Fracture Mechanics)

\[ K = \sigma (\pi a)^{1/2} \]

Material Failure 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 \( K_{\text{failure}} \)
(Understand microstructural size scale)

Failure: Brittle Ductile
Fatigue loading: "Metal Disease"

"Dead load"

"Live load"

\[ \sigma \]

Failure only under live load

| \[ a \] | \[ a_o \] |
---|---|
Cycles (Time)

\[ a_o = \text{initial defect population} \]
Simple Crack Growth Laws:

\[
\frac{da}{dN} = A \Delta K^m
\]

Rate of crack growth \hspace{1cm} \text{Driving force}

A, m are material's "constants"

Conventional materials \( m = 2 - 4 \)

Intermetallics, ceramics \( m \) upto 50
Orientation relationship between the crack and the lamellar microstructure.
(i) Crack arrester orientation, (ii) Crack divider orientation and (iii) crack delamination orientation.
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.
\[ \frac{da}{dN} = A \Delta K^m \]
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
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.

Effects of lamellar colony size and lamellar plate thickness.
Load

"resistance curve"

\[ \text{K}_{\text{pop in}} \] (min. \( \approx 12 \text{MPa}\text{m}^{1/2} \))
(min. \( \approx 25 \text{MPa}\text{m}^{1/2} \))
(max. \( \approx 17 \text{MPa}\text{m}^{1/2} \))

(colony size 400-1800 \( \mu \text{m} \))
(Y5-22 MPa m^{1/2})

(max. \( \approx 27 \text{MPa}\text{m}^{1/2} \))

Use of "resistance curve"

Engineering sense

Use \( K_{\text{pop}} \) values even in "ductile" systems.

Need to ensure that microcracks do not join up

Need to utilize fatigue crack growth to

generate resistance curve

or above system variation in \( K_{\text{max}} \) and \( K_{\text{pop}} \) is

extreme for engineering alloys? (Weibull modulus \( \approx 7.8 \)).
Process zone sampling effects at failure (Stresses > $\sigma_f$ (tension) over volume sampled).

Fracture toughness test  No of "grains"  $\sigma_{II}$ (MPa)

<table>
<thead>
<tr>
<th>$K_c$</th>
<th></th>
<th>19</th>
<th>1250 - 300</th>
</tr>
</thead>
<tbody>
<tr>
<td>13 MPa m$^{1/2}$</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>25 MPa m$^{1/2}$</td>
<td></td>
<td>300</td>
<td>1250 - 300</td>
</tr>
</tbody>
</table>

Tensile test  No. of "grains"  $\sigma_{II}$ (MPa)

| $\sigma_f$ = 300 MPa |  | 24,000 | 300 |

---

NB. For coarse "randomised" (400µm+) fully lamellar, % E1 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 \( m \)) \( \Rightarrow \) must design on total life (S-N curves).

\[
\frac{da}{dN} (\text{mm/cycle})
\]

- 700°C/air
- 800°C/vacuum
- ambient temp./air
- 700°C/vacuum

\[
\Delta K \ (\text{MPa} \sqrt{m})
\]

\[
(\Delta K)
\]

Figure 6.27

Fatigue crack growth curves \( da/dN \) versus \( \Delta K \) for cast and HIPed XD™ gamma.

\[
\sigma_{0.2} = 340 \text{ MPa} \ (327 - 346 \text{ MPa})
\]

\[
\sigma_f = 400 \text{ MPa} \ (383 - 415 \text{ MPa})
\]

\[
\epsilon_f = 1\% \ (0.88 - 1.10\%)
\]

\[
E = 160 \text{ GPa}
\]

\[
K_c = 15-18 \text{ MPa m}^{1/2}
\]
NB: Steep $da/dN$ v $\Delta K$ curves $\Rightarrow$ 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 down => 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. (≈ 0.95σf)

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

Surface area tension is x10 that in bending.
Volume tension is x26 that in bending.
$\delta_{\text{surface}} = \frac{6M}{W^2\beta}$

$M = \text{PS/ft}$

Bend

Tension = $\pi D \times \text{gauge length}$

Match for $\delta_{\text{surface}} \geq \delta_{\text{run-out in tension?}}$.
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 $\Delta K$ and $K_c$ (brittle fracture).

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

Optimisation of fully lamellar, "fine" colony size required (randomised). Even then steep $da/dN$ vs $\Delta K$, modest $K_c$ (15MPam$^{-3/2}$).

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

\[ \sigma \text{ (MPa)} \]

\[ \varepsilon = 1.25\% \]

in situ - exp.

\[ \varepsilon = 1.25\% \]

\[ \dot{\varepsilon}_2 = 4.16 \times 10^{-4} \text{ s}^{-1} \]

\[ \dot{\varepsilon}_2 / \dot{\varepsilon}_1 = 3 \ldots 20 \]

\[ (10^{-18}/v) \text{ (mm\textsuperscript{-3})} \]

\[ T(K) \]

in situ study, \( T = 900 - 970 \text{ K} \)