Additive Manufacturing of Intermetallics: Microstructures and Mechanical Properties of Iron Aluminides Processed by SLM and LMD

M. Palm
• Iron aluminides and investigated Fe–Al–X alloys

• Microstructures
  – Misorientations within individual grains
  – Influence of preheating temperature, scan strategy, annealing

• Mechanical properties: yield strength, creep, ductility

• Chemically graded Fe–Al and Fe–Al/steel samples
Motivation

- Fe-base materials with low density 5.7 - 6.7 g/cm$^3$ vs. 7.85 (steel)
- Excellent corrosion resistance (metal dusting, steam, molten salts…)
- High wear resistance
- Progress in increasing strength at high temperatures
- Limited ductility at ambient temperatures (0.5–1.5% elongation at RT)
- Cheap material (materials costs, no strategic elements & production)

⇒ Possible replacement for Cr steels and Ni & Co base alloys
AM of Fe–Al

AM processing is of specific interest for the production of intermetallic parts because their high hardness and brittleness makes near net-shape production desirable.

Why additive manufacturing (AM) of Fe–Al?

- Alternative processing route to casting
- Near net-shape processing ⇒ minimum of machining
- High cooling rates ⇒ fine microstructure ⇒ higher ductility
- Highly exothermic reaction between Fe and Al
  ⇒ build up from elemental powders ✓
- Possibility to build chemically graded steel/Fe–Al parts?
Employed AM processes

1. Laser Metal Deposition **LMD**
2. Selective Laser Melting **SLM**
3. (Electron Beam Melting **EBM**)

Build up on various preheated substrates: Fe, Fe–Al, 1.4301, 1.1730, P92…

Reaction zone between substrate and iron aluminide \(\leq 1\) mm
Investigated Fe–Al–X alloys

Fe–28Al
“Demo”

Fe–30Al–10Ti
Increase of L2₁ ↔ B2

Fe–30Al–5Ti–0.7B
Precipitation of borides

Fe–22Al–5Ti
Coherent A₂+ L₂₁

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Fe–28Al

“Demo”

Defect free samples by preheating at 200 °C (LMD) and 600 °C (SLM)

Compositions
- SLM: 27.2 ± 0.3 at.% Al
- LMD: 27.8 ± 0.3 at.% Al
- Powder: 28.3 at.% Al

Elongated grains up to mm (as cast several 100 μm)
Growth preferentially in [001] direction
(direction of the highest heat flow in α-Fe,Al)
Fe–28Al: Effect of preheating temperature on grain size

Grain size decreases with increasing preheating temperature
⇒ grain size increases with increasing temperature gradient
Fe–28Al: Misorientations within individual grains

Misorientations within individual grains up to 20°

Higher misorientations in SLM compared to LMD (due to higher cooling rate)
Fe–28Al: Effect of scanning strategy on misorientations

SLM; processed at RT

Scan strategy has a marked influence on misorientations

Unidirectional 2

1

layer 1, 3, ...

layer 2, 4, ...

Unidirectional 4

2

layer 1, 5, ...

layer 2, 6, ...

layer 3, 7, ...

layer 4, 8, ...

Meander

3

layer 1, 3, ...

layer 2, 4, ...

Max.: 15.2°

Av.: 3.2°

Max.: 16.9°

Av.: 4.0°

Max.: 11.9°

Av.: 2.8°
Fe–28Al: Effect of preheating temperature on misorientations

Misorientations decrease with increasing preheating temperature

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Fe–28Al: Recrystallisation

No recrystallization after various heat treatments
TEM analysis of the dislocation structure

Very high dislocation densities
Subgrains are formed through formation of dislocation walls
Dislocations are arranged in a network of parallel screw dislocations
Dislocation densities are about one magnitude higher in SLM samples than in LMD samples
Subgrains are located on common zone axes and are inclined by 1–2° against each other.
Fe–28Al: XRD analysis of internal stresses

**XRD: local stress measurements**

<table>
<thead>
<tr>
<th>Strategy</th>
<th>Stress S1 [MPa]</th>
<th>Stress S2 [MPa]</th>
</tr>
</thead>
<tbody>
<tr>
<td>LMD 1, top</td>
<td>130 ± 60</td>
<td>36 ± 67</td>
</tr>
<tr>
<td>LMD 2, top</td>
<td>82 ± 29</td>
<td>29 ± 70</td>
</tr>
<tr>
<td>LMD 3, top</td>
<td>109 ± 37</td>
<td>-10 ± 103</td>
</tr>
<tr>
<td>LMD 1, side</td>
<td>-101 ± 90</td>
<td>84 ± 95</td>
</tr>
<tr>
<td>LMD 2, side</td>
<td>-148 ± 66</td>
<td>-65 ± 37</td>
</tr>
<tr>
<td>LMD 3, side</td>
<td>24 ± 104</td>
<td>16 ± 27</td>
</tr>
<tr>
<td>SLM side (20 °C)</td>
<td>432 ± 172</td>
<td>159 ± 45</td>
</tr>
<tr>
<td>SLM side (200 °C)</td>
<td>378 ± 92</td>
<td>187 ± 27</td>
</tr>
</tbody>
</table>

Local stress measurements show mixture of tenseile and compressive stresses. High tensile stresses vertical to individual tracks, low stresses along tracks. Compressive stresses vertical to individual layers, low stresses along layers. Stresses are higher in SLM than in LMD samples.
No marked difference in compressive yield strength between AM and as-cast samples or whether SLM and LMD samples are tested horizontal (parallel) or vertical to BD.

LMD samples show improved ductility compared to SLM and as-cast samples. The strong crystallographic texture and high dislocation density have no influence on ductility.

D. Risanti et al.: Intermetallics 13 (2005) 1337
Fe–30Al–10Ti: Microstructure

Fe–30Al–10Ti

Increase of L2₁ ↔ B2

Defect free samples by preheating at 700 °C (LMD) and 800 °C (SLM)

Equiaxed grains

Average grain size < 5 μm (as cast > 100 μm)

No preferred orientation of the grains

No misorientations within individual grains
Fe–30Al–10Ti: Effect of preheating temperature on grain size

No effect of preheating temperature on grain size
Fe–30Al–10Ti: Yield stress and ductility

No marked difference in compressive yield strength and ductility between AM and as-cast samples or whether SLM and LMD samples are tested horizontal (parallel) or vertical to BD (yield stress).

M. Palm, G. Sauthoff; Intermetallics 12 (2004) 1345
Fe–30Al–5Ti–0.7B: Microstructure

Precipitation of borides

Defect free samples by preheating at 400 °C (LMD) and 600 °C (SLM)

Equiaxed Fe–Al grains (< 5 μm) + TiB₂ (50 nm) in Fe–Al grains and at grain boundaries (as cast: Fe–Al (10 μm) + TiB₂ (100 nm) at GB)

No preferred orientation of the grains

No misorientations within individual grains

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Yield stress of Fe–30Al–5Ti–0.7B at 700 °C

No marked difference in yield strength between AM and as-cast samples or whether SLM and LMD samples are tested horizontal (parallel) or vertical to BD.
Secondary creep of SLM and LMD samples of Fe–30Al–5Ti–0.7B at 600 °C and 700 °C established in compression by stepwise increasing the load.

No marked difference in compressive creep strength between AM and as-cast samples or whether SLM and LMD samples are tested horizontal (parallel) or vertical to BD.
Marked difference in ductility between SLM and LMD or as-cast samples, whether SLM samples are tested horizontal (parallel) or vertical to BD and if LMD samples are annealed.
Local stress measurements show mixture of **moderate tensile** and **compressive** stresses. **Tensile** or **compressive** (SLM) stresses vertical to individual tracks, low stresses along tracks. **Compressive** stresses vertical to individual layers, low stresses along layers.
No direct relation between internal stresses and ductility of the samples.
Fe–22Al–5Ti: Microstructure

Defect free samples by preheating at >800 °C (LMD) and 800 °C (SLM)

Equiaxed grains (10 µm) with coherent A2 + L2₁ microstructure (30 nm)
(comparable to as-cast microstructure)

No preferred orientation of the grains

No misorientations within individual grains
Compressive yield stress of Fe–22Al–5Ti at 700 °C

Marked difference whether SLM samples are tested horizontal or vertical to BD. Reduction of yield strength after annealing at 700 °C due to coarsening of coherent microstructure.
Secondary creep of SLM samples of Fe-22Al-5Ti at 600 °C and 700 °C established in compression by stepwise increasing the load.

No marked difference in compressive creep strength between SLM and as-cast samples or whether SLM samples are tested horizontal (parallel) or vertical to BD.
Fe–22Al–5Ti: Ductility

Ductility of Fe–22Al–5Ti determined in 4-point bending

SLM samples show improved ductility compared to as-cast samples.
Chemically graded Fe–Al samples produced by LMD
YAG-Laser: 2000 W

Fe-28Al: powder particle size 45 – 90 µm

Chemical composition varied by the rotation speed $m_p (\text{Umin}^{-1})$ of the feeder
Calibration of the composition

Production of binary Fe–Al samples by varying the speed ($m_p$) of the Al feeder

<table>
<thead>
<tr>
<th>$m_p$ Al (Umin$^{-1}$)</th>
<th>Al (at%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.25</td>
<td>31.0 ± 0.8</td>
</tr>
<tr>
<td>0.50</td>
<td>33.7 ± 0.6</td>
</tr>
<tr>
<td>0.75</td>
<td>36.4 ± 1.1</td>
</tr>
<tr>
<td>1.00</td>
<td>38.5 ± 0.6</td>
</tr>
</tbody>
</table>

Linear dependence between rotation speed of the feeder and obtained composition.

⇒ Production of LMD samples with defined composition gradients possible
Production of chemically graded Fe–Al samples by LMD is possible.
Chemically graded Fe–Al/steel 316L sample

Production of chemically graded Fe–Al/steel samples by LMD is possible.
Conclusions

• Intermetallic Fe–Al alloys can be processed by SLM and LMD

• Fe–Al alloy concepts developed for casting can be transferred to AM

• Yield strength and creep strength are comparable to that of as cast alloys

• Partly improvement of ductility (through internal stresses (?) not through reduction of grain size)

• Chemically graded Fe–Al samples with defined concentration profiles can be manufactured by LMD
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