

Final Draft
of the original manuscript:

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In: Intermetallics (2008) Elsevier

DOI: 10.1016/j.intermet.2008.02.004

Microstructure and Mechanical Properties of Ti 45Al 5Nb +(0 - 0.5C) Sheets

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Abstract

The possible application of gamma titanium-aluminides in aerospace industry requires a detailed understanding of the microstructure-property relationship of sheets made from this material. This paper reports the mechanical properties of sheets up to 1000°C, based on alloy concepts with high Nb concentrations and small additions of C. Sheets were manufactured by rolling powder metallurgical compacts with compositions Ti 45Al 5Nb and Ti 45Al 5Nb 0.5C. The microstructures of both sheets are „near gamma“ and consist of γ -TiAl and α_2 -Ti₃Al phases. The texture of both phases is very weak. The strengths levels are very high and that of the C-

containing sheet exceeds that of the C-free material at RT by ~ 200MPa. While the mechanical properties of Ti 45Al 5Nb are independent from the direction in the sheet, this is for the C-containing sheet only true in the upper temperature range.

Keywords: A. Titanium aluminides based on TiAl; B. Mechanical properties at high temperatures, texture; C. Powder metallurgy; D. Microstructure;

1. Introduction

Gamma titanium aluminides are novel light-weight structural materials for applications in car engines as valves, turbocharger turbine wheels or conrods and in aero engines as blades, vanes or discs where they have to withstand temperatures up to ~ 800°C. A good overview with respect to the current status of research and development can be obtained from [1].

In addition to the mentioned applications, sheets are an important mill product for the manufacturing of further aerospace-components such as divergent/inlet flaps, nozzle tiles or thermal protection shields [2,3].

For the manufacturing of TiAl-based sheets special hot rolling techniques for ingot-metallurgical (IM) and powder-based metallurgical (PM) processing routes have been developed [4] and applied to a variety of different alloy systems. Hitherto the most experience with respect to performance and properties have been gained for sheets based on Ti 47Al 4(Nb, Mn, Cr, Si, B) (at%) (called γ -TAB) [5] and Ti 46.5Al 4(Cr, Nb, Ta, B) (at%) (called γ -Met) [6].

While these alloys contain about 1% Nb and Al concentrations in excess of 46%, Huang [7] and Chen et al. [8] have shown, that higher Nb contents and reduced Al concentrations can result in considerably increased strength levels of TiAl alloys. At the end of the „1990s“ this concept was taken up again and alloy developments finally resulted in so-called TNB alloys showing strength levels of up to 1GPa at room temperature. Fundamental research on these alloys was carried out by Appel and coworkers [9,10] with focus on ingot metallurgy and thermomechanical processing.

Considering the mentioned alloy concepts, in a first approach sheet material with the composition Ti 46Al 9Nb (at%) has been manufactured via a PM route [11]. These sheets turned out to be characterized by increased strength levels and improved creep behaviour [12] if compared to γ -Met and γ -TAB sheets. Sheets of the latter alloys showed a pronounced mechanical anisotropy especially in the temperature range 600 to 900°C. For example in γ -TAB sheets directional differences in strength ranged up to more than 200MPa [5]. In contrast, in Ti 46Al 9Nb sheets the mechanical anisotropy parallel and perpendicular to the rolling direction is very low over the whole temperature range from RT to 1000°C.

For the present investigations the chemical composition has further been tuned. The Al concentration has been reduced by 1at% and that of Nb by 4at%. Former investigations [13,14,15] have shown that small additions of C can result in a strengthening of TiAl alloys. Such an influence of C is also reported for novel alloys with lower Al and higher Nb concentrations [10]. Therefore, for a second set of sheets a small amount of carbon was added. Consequently, two sets of sheets with

compositions Ti 45Al 5Nb and Ti 45Al 5Nb 0.5C (all in at%) were manufactured by hot rolling of PM-slabs.

In this paper, the processing steps are briefly described. The powder-based hot isostatically pressed compacts, the „as-rolled“ and finally the „primary annealed“ sheets are characterized with respect to microstructures, impurity levels, textures, hardness, and tensile properties.

2. Experimental

For the powder production, the elemental constituents were alloyed and homogenized by means of a plasma torch in a water cooled copper crucible. The resulting ingots were atomized using the PIGA-technique. Upon application of this technique, a homogeneous melt bath is generated by a plasma torch. From this melt bath the melt runs through a water cooled induction heated copper funnel, which forms a thin melt stream and guides it into the center of a gas nozzle. Inside the gas nozzle, the melt is atomized by argon gas at a high pressure. The solidified droplets are gathered as alloy powder in a powder can, which is mounted gas tight to the atomization facility. The subsequent powder handling was done under argon in glove boxes. More details about the PIGA-technique can be found in [16].

The alloy powder of both compositions, Ti 45Al 5Nb and Ti 45Al 5Nb 0.5C, are characterized by the same powder particle size distribution: $d_{50} \sim 90\mu\text{m}$, 85wt% < 180 μm and 19wt% < 45 μm . For both compositions powder of the size fractions 0 - 180 μm were filled in rectangular Ti-cans with dimensions around 350 × 140 × 25 mm³. The cans were degassed during 16h at 400°C and finally sealed gas tight. Hot isostatic pressing (hip) was conducted for 2h at 1270°C and 200MPa. The hip'ed

compacts were multi-pass rolled within the ($\alpha + \gamma$) phase field at Plansee AG, Austria. The dimensions of the sheets ranged around $600 \times 440 \times 1.2 \text{ mm}^3$. The „as rolled“ sheets were subjected to a „primary annealing“ for 2h at 1050°C . The objective of this heat treatment was to reduce internal stresses and to stabilize the microstructure. More details about the sheet rolling can be read in [4,17].

The microstructures were characterized by X-ray diffraction (XRD), light-optical microscopy (LOM) using polarized light, scanning electron microscopy (SEM) in back-scattered electron (BSE) mode and transmission electron microscopy (TEM). XRD was conducted on a conventional diffractometer from Bruker AXS using Cu-K_α radiation. A Zeiss DSM 962 was used for the SEM investigations. TEM measurements were performed at a Philips CM12 operated at 120 kV. Samples for LOM were mechanically ground and subsequently polished. For a better distinction between γ - and α_2 -phase samples for LOM were also color etched [18]. For SEM the samples were electropolished in a solution of 62% methanol, 32% buthanol and 6% perchloric acid at -35°C and 19V. Thin TEM foils were prepared by mechanical grinding and polishing, followed by dimpling to about $25 \mu\text{m}$ in the central part of the 3 mm disc. Finally, the foils were ion-milled with 3.6 keV Ar-ions using a Gatan PIPS until electron transparency occurred.

The volume fractions of the different phases within the samples were estimated by XRD-measurements (Rietveld analyses) as well as by quantitative metallographic evaluation conducted on color etched LOM- and SEM-images.

Hardness data were obtained using a Vickers hardness tester from Zwick and a load of 10kg (HV10).

Texture measurements have been performed by XRD in reflection geometry with filtered Cu K_{α} radiation using a texture goniometer with eulerian cradle and parallel beam optic from Bruker AXS. Pole figures of the γ -phase and of the α_2 -phase were measured up to a pole distance angle (χ) of 85° and subsequently corrected for background and intensity loss. The orientation distribution function (ODF) of the γ -phase was calculated by using the series expansion method [19] up to an order of 22. The ODF of the α/α_2 -phase was calculated employing the WIMV algorithm of the BEARTEX Berkeley Texture Package [20]. Subsequently, complete and inverse pole figures were recalculated from the ODF. The texture of the tetragonal γ -TiAl phase can be described by components similar to those found in as-rolled fcc-metals and the texture of the hexagonal α -Ti(Al) phase is comparable to that in Ti-based alloys. For further details the reader is referred to [21,22].

Samples for tensile tests were prepared by spark erosion parallel (RD) and perpendicular (TD) to the rolling direction. The dimensions of gauge length and width were 25 and 4mm, respectively. The samples were tested at an initial strain rate of $1 \cdot 10^{-4} \text{ sec}^{-1}$. Between RT and 800°C the tests were conducted at standard pressure in air, above 800°C at a pressure of $5 \cdot 10^{-4} \text{ Pa}$. At RT up to four samples were tested, while above RT the tensile data are based on two to three tested samples. The yield strength data were taken at 0.2% plastic elongation.

The concentrations in oxygen, nitrogen and carbon were analyzed in alloy powders, hip'ed compacts and primary annealed sheets using a conventional LECO melt extraction system. In summary, the impurity levels in nitrogen and oxygen ranged for

all sample types around 20 to 100 $\mu\text{g/g}$ and 370 to 410 $\mu\text{g/g}$, respectively. These low levels indicate that the nitrogen and oxygen pick-up during powder handling or one of the subsequent processing steps could largely be avoided. In the Ti 45Al 5Nb sheets the C concentration was analysed to around 20 $\mu\text{g/g}$, which corresponds to 0.006at%. This contamination is seen to come along with the elemental constituents used for the alloying. In Ti 45Al 5Nb 0.5C sheets the carbon concentration was determined to 1480 $\mu\text{g/g}$ which relates exactly to 0.50at%.

3. Results and discussion

3.1 Hot isostatically pressed compacts

X-ray diffraction patterns obtained from the hip'ed powder compacts of both alloys indicated the presence of only two phases, γ -TiAl and α_2 -Ti₃Al. The volume fractions of the γ - and α_2 - phases as obtained by the different techniques are in good accordance and amount to around 75 and 25%, respectively. As shown by SEM-images, the general microstructures of the two hip-compacted materials are identical, Figs. 1a,d. Under these imaging conditions the γ -grains are dark grey to black while the α_2 -grains appear white, i.e. a globular „near gamma“ microstructure with numerous annealing twins is found. Such a high density of annealing twins has been observed earlier also in hip-compacted Ti 46Al 9Nb-powder [11] and is seen as an indirect proof for the low stacking fault energy in TiAl alloys with high Nb-concentrations [10,23].

The hardness of Ti 45Al 5Nb 0.5C was analysed to $\sim 363\text{HV}_{10}$, which is about 15% higher compared to the C-free compact ($\sim 316\text{HV}_{10}$), Fig. 2.

The texture measurements of both hiped samples show uniform pole figures without any preferred orientation. This is visualized for the $\{001\}_{\gamma}$ - reflection in Figs. 3a,b.

3.2 Sheets “as-rolled”

After hot rolling in the upper range of the $(\alpha+\gamma)$ phase field, both alloys show a phase composition of about 70 vol% α_2 -Ti₃Al and 30 vol% γ -TiAl, Figs. 1b,e. These volume fractions are a consequence of the relative fast cooling after the last rolling step, which is fast enough to preserve approximately the high temperature state. The microstructures of the sheets consist for both alloys of large previous α -grains which transformed to lamellar α_2/γ colonies during cooling, γ -grains and a small portion of α_2 - grains. The γ -grains and the lamellar colonies are elongated in rolling direction. There is a tendency visible for the lamellae being aligned in rolling direction or being tilted under a small angle to RD. The elongation of α_2/γ colonies and γ -grains in RD is less pronounced while the alignment of the lamellae is similar in Ti 45Al 5Nb 0.5C if compared to the C-free alloy. The lamellar spacing is too small to be resolved by SEM, but has been analysed by TEM. Fig. 4 shows bright field TEM images of the two as-rolled sheets taken at lamellar colonies. The lamellar spacings range around 17 and 30nm for the C-free and C-containing sheet, respectively. Furthermore, in both alloys extremely fine grained dynamically recrystallized areas are visible in between the larger grains.

The hardness of the as-rolled sheets is significantly increased with respect to the hip-compacted state (Fig. 2). This is seen to be a consequence of the high dislocation

density after deformation. This effect almost superimposes the influence of the different C-contents as is visible after hip.

After hot rolling γ - as well as α -textures were measurable. The γ -texture of the C-free material is dominated by a strong Goss2 component (Fig. 3c), whereas the γ -texture of the C-containing alloy shows a higher amount of brass component (Fig. 3d). The α -textures of both alloys consist of a basal fiber that is tilted about $\pm 20^\circ$ in rolling direction and a transversal component (Figs. 3g,h). However, it must be pointed out, that the textures are weak and the differences between the two alloys are in fact very small.

3.3 Sheets “primary annealed”

3.3.1 Microstructures and textures

The phase fractions in both alloys range around 80 vol% γ -TiAl and 20 vol% α_2 -Ti₃Al after primary annealing. Figs. 1c,f show the bimodal character of the two microstructures. They are characterized by large primary γ -grains, which are surrounded by a mixture of very fine α_2 - and γ -grains (1 – 3 μm). These fine grained areas have been formed from previous large α -grains by transformation to lamellar α_2/γ colonies and subsequent spheroidization. The primary γ -grains show an increased size compared to the as-rolled state. In the carbon-free sheet the γ -grains have an average diameter of $\sim 10\mu\text{m}$ and are elongated in rolling direction, which, however, is clearly less pronounced compared to the as-rolled state of this alloy. Lamellar colonies are not visible. In the C-containing sheet, the primary γ -grains have an average diameter of 6 μm . Different to the C-free sheet, some lamellar colonies of different diameters up to $\sim 15\mu\text{m}$ are visible, a TEM micrograph of such an area is

displayed in Fig. 5. The lamellae still seem to be predominantly aligned in RD. In contrast to the C-free sheet, directions parallel and perpendicular to the rolling direction can not be distinguished by the primary γ -grains. Typical for both microstructures are the α_2 -plates interspersed at the grain boundaries of the larger γ -grains. Most of these plates, when they are close to each other, arrange nearly parallel.

Recovery and recrystallization during the annealing also cause a decrease of the Vickers hardness (Fig. 2) to 407 and 351HV10 for Ti 45Al 5Nb 0.5C and Ti 45Al 5Nb, respectively. Both values, however, are still slightly above the respective hardnesses of the as-hiped states and reflect the slightly finer microstructure in the primary annealed sheets. The hardening due to the C-content is in the same order of magnitude as in hiped compacts.

In both alloys the intensity as well as the sharpness of the γ -texture decrease during annealing (Figs. 3e,f). Since the γ -phase fraction increased considerably from 30 to 80 vol.%, this demonstrates that the newly formed γ -grains show nearly no preferred orientation. Only a small amount of modified cube component appears additionally, due to recrystallization [24,25]. The transversal component of the α -phase is almost completely dissolved after annealing (Figs. 3i,j). This means that especially α -phase with this orientation transformed to γ -phase. In contrast to this, the basal component becomes only weaker and more diffuse. The texture evolution corresponds well to the evolution of the microstructure from the as-rolled to the primary annealed state.

Previously investigated γ -TAB and γ -Met sheets exhibit strong textures solely consisting of modified cube texture component showing maxima of ~ 12 and ~ 5 multiple random distribution, respectively [12,26]. The reason for significantly weaker textures in sheets like Ti-46Al-9Nb [12] is the large amount of Nb, which diminishes the diffusion and thus reduces the recrystallization kinetics. In addition, both alloys of this study have lower Al contents, which results in a higher α -phase fraction during the rolling process. A large part of this α -phase then transforms to new randomly orientated γ -phase in the subsequent heat treatment and thus reduces the texture furthermore.

3.3.2 Accommodation of carbon

Carbon usually is alloyed in order to increase strength and creep strength in TiAl alloys [e.g. 27,28]. Both, the solid solution of C and the formation of carbon containing precipitates is seen to promote this effect [14]. The formation conditions and thermal stabilities of such precipitates as the cubic Ti_3AlC Perovskite-phase and the hexagonal Ti_2AlC -phase ("H"-phase) were investigated by Chen et al. [13] using powder metallurgical Ti 48Al 2Nb with additions of C (around 0.2at.%). In order to decide whether the carbon is kept in solid solution or precipitates were formed in the present Ti 45Al 5Nb 0.5C sheet, a number of different techniques were applied.

Chemical analyses conducted on the two types of sheet material indicated the C-concentration to match with the nominal concentrations, i.e. near 0 and 0.5at%. XRD-data of the primary annealed sheets were thoroughly analysed with respect to lattice parameters. For the γ -phase, lattice constants of 0.4019nm (a-axis) and 0.4064nm (c-axis) were determined for Ti 45Al 5Nb. For the carbon-containing material the

respective lattice constants were analysed to be 0.4020nm and 0.4065nm. For the α_2 -phase the hexagonal lattice parameters for the C-free and C-containing sheet, respectively, were determined to: a-axis 0.5771nm and 0.5778nm (with C), c-axis 0.4643 and 0.4656nm (with C). In conclusion, the addition of carbon causes a very small increase in the lattice constants for the γ -phase and a distinct increase in the lattice constants of the hexagonal α_2 -phase. These results indicate that a higher percentage of the solved carbon is rather accommodated in the α_2 - than in the γ -phase. Conclusions with respect to a formation of precipitates can not be drawn.

Carbon containing precipitates in TiAl alloys often are detected and characterized by TEM [e.g. 9,14,15,29]. Respective investigations using primary annealed sheet material of both compositions, however, gave no indication for the presence of such precipitates.

In previous investigations on Ti 48.5Al 0.4C, Perovskite precipitates were not only detected by TEM but also by small angle neutron scattering (SANS) [30]. Therefore, samples from both sheets were also investigated by SANS. Thorough evaluation of the scattering data and close inspection of TEM images from both alloys, however, indicated the microstructure and the related scattering effects to be too complex, and the presence of carbon containing precipitates with dimensions around a few nanometer could not be proofed.

In another approach for detecting precipitates, samples from Ti 45Al 5Nb 0.5C-sheet material were also analysed by 3D atom probe (AP). Data analysis gave no indication for the presence of C-precipitates which could be responsible for a

precipitation hardening effect. A detailed report of the results obtained by AP is under preparation and will be published shortly.

In conclusion, none of the different characterization techniques could provide evidence for the existence of C-precipitates in Ti 45Al 5Nb 0.5C. Since chemical analyses unequivocally indicated the sheet to contain 0.5at% carbon, it has to be concluded that the thermal treatments of the sheet which were applied in the course of manufacturing and primary annealing did not cause the formation of carbon containing precipitates.

3.3.3 Tensile tests

Fig. 6 shows the variation of the yield strength (YS) with temperature for both sheets with the test direction parallel to the rolling direction. For comparison the respective data for a Ti 46Al 9Nb-sheet [11] are also shown. At RT the YS of the Ti 45Al 5Nb-sheet ranges around 800MPa and decreases slightly with increasing temperature to around 570MPa at 700°C. To higher temperatures the loss in strength is increased and amounts to 54MPa at 1000°C. Over the whole range from RT to 700°C the strength of this sheet is about 50MPa higher with respect to the formerly investigated Ti 46Al 9Nb-sheet. At 800°C the YS is identical for both materials and to higher temperatures the strength of the 9at% Nb containing sheet ranges about 50MPa above that with the lower Nb concentration. The reason for this different mechanical behavior is seen to be caused by differences in phase distributions and grain sizes. Although the type of microstructure as presented in section 3.3.1, Fig. 1c, is identical for both alloys, the primary γ -grains are slightly smaller in Ti 45Al 5Nb, while the α_2 -phase fraction is much larger (~20vol%) if compared to Ti 46Al 9Nb (~5vol%). In

addition, the fine grained area between the larger primary γ -grains is much more extended in Ti 45Al 5Nb than in Ti 46Al 9Nb.

The carbon containing sheet is characterized by a very high YS-level of 1070MPa at RT. At 700°C the YS of 726MPa is still 150MPa above that of Ti 45Al 5Nb. Only at 1000°C the strength levels of the C-containing and C-free sheet are about the same and slightly below the strength of Ti 46Al 9Nb, Fig. 6. Apart from the carbon addition other microstructural aspects might be responsible for the gain in YS. The visible differences in the microstructures of Ti 45Al 5Nb and Ti 45Al 5Nb 0.5C, however, are quite small: the types of microstructure are identical (Figs. 1c,f) as are the α_2 - and γ -phase fractions. The concentration of a few small isolated lamellar colonies in Ti 45Al 5Nb 0.5C is too small to be responsible for this effect. The only concise difference is the size of the initial primary γ - grains which is about 10 and 6 μ m in 0- and 0.5at%C bearing sheets, respectively. If the increase in strength is assumed to be a grain size effect and Hall-Petch constants are used which are commonly accepted for TiAl [e.g. 31,32,33], the carbon containing sheet should have a RT YS of around 890MPa, which is only one third of the observed increase in strength. The reason for the high strength, therefore, is primarily seen as a carbon-effect. Since until now none of the applied analysis techniques provided information for the presence of C-precipitates, we attribute the strengthening to a solid solution effect by carbon.

In Fig. 7a the variation of the ultimate tensile strength (UTS) and YS with temperature are shown for Ti 45Al 5Nb for both directions, parallel and perpendicular to the rolling direction. The figure shows that the mechanical anisotropy parallel and perpendicular to the rolling direction is very small over the whole temperature range. This holds especially for the range 600 to 900°C, where sheets based on a former TiAl-

generation showed large differences in strength depending on the test direction. The negligible anisotropy is in accordance with the results of the texture analyses. An elongation of primary γ -grains in RD is still visible after primary annealing, however, this topological effect seems to have no marked influence on the TD strength level. A similar almost vanishing anisotropy was also observed in Ti 46Al 9Nb [11]. This is a pronounced difference to sheets based on TiAl-alloys of the second generation as γ -Met or γ -TAB where stronger textures caused pronounced mechanical anisotropies [5,26,34].

Fig. 7b shows for Ti 45Al 5Nb 0.5C the course of UTS and YS with temperature for both, RD and TD. It should be pointed out, that all tensile test specimens prepared in TD failed at temperatures below 500°C at plastic elongations even smaller than 0.1%. YS-data therefore are not available at RT and 350°C.

Different to the C-free sheet, there is a considerable anisotropy in RD and TD for both, UTS and YS in the temperature range from RT to 500°C. At 700°C and above this temperature the anisotropy, if any, is very small. At its maximum, at 500°C, the anisotropy is as high as 200MPa. At that temperature the YS-level of the TD direction exceeds that of the C-free sheet by only ~24MPa. The reason for this pronounced anisotropy is not clear. The texture is quite low and a topological effect of the primary γ -grains is not visible. Alone the presence of lamellar colonies with the lamellae oriented primarily in RD are a significant difference to the C-free sheet.

Fig. 8 shows the variation of the plastic fracture elongations with temperature for both alloys and both directions, RD and TD. For a better resolution at small plastic fracture

elongations a logarithmic scale was chosen. The plastic elongations below 500°C are very low. Alone Ti 45Al 5Nb shows in RD plastic elongations around 1% between RT and 500°C. From 500 to 1000°C a continuous increase in plastic elongation can be observed for both sheets and both directions, RD and TD. Ti 45Al 5Nb seems to exhibit slightly higher elongations if compared to Ti 45Al 5Nb 0.5C, and for each alloy there seems to be a tendency such that the plastic elongation until fracture is larger in rolling direction than in transverse direction.

“The development and evaluation of TiAl sheet structures for hypersonic applications” by S. L. Draper et al. [3] is based on “GM PX” sheet material. This is also a newly developed TNB- alloy with a composition around Ti 45Al 5Nb 0.2B 0.2C (at%). The presented tensile properties are similar to those analyzed for the present Ti 45Al 5Nb 0.5C alloy. However, since the GM PX material contains 0.2at% boron and the microstructure of the sheets is of “duplex” type, a direct comparison to the present results is difficult. Nevertheless, this gives further evidence for the outstanding mechanical properties of these alloys.

4. Summary and outlook

Ti 45Al 5Nb and Ti 45Al 5Nb 0.5C sheets have successfully been manufactured by hot rolling of PM-prematerials. The microstructure of Ti 45Al 5Nb sheets is of “near gamma” type, consisting of larger γ -grains (average diameter 10 μ m) separated by a very fine mixture of γ - and α_2 -grains (1 – 3 μ m). The sheets are characterized by high strength in the range RT to 700°C. The texture is very small as is the anisotropy in strength parallel (RD) and perpendicular (TD) to the rolling direction. The plastic fracture elongation of these sheets ranges in RD slightly above 1% between RT and

500°C with a considerable increase to higher temperatures. In TD the plastic fracture elongation is slightly smaller between RT and 500°C.

The strength of this sheet is found to about 50MPa higher between RT and 700°C with respect to the previously investigated Ti 46Al 9Nb sheet.

The C-containing sheet exhibits the same type of microstructure. The γ -grains are slightly smaller (6 μ m) compared to the C-free sheet. Some isolated lamellar colonies, up to 15 μ m in diameter were retained from the as-rolled state. The strength in rolling direction is very high: It ranges between RT and 700°C by about 250MPa above that of Ti 45Al 5Nb. Different to the C-free sheet, however, there is a considerable anisotropy and the strength in TD is up to 250MPa below that in RD. The reason for the high mechanical anisotropy is not clear at moment. The plastic fracture elongation in rolling direction is slightly smaller if compared to the C-free sheet; in TD the plastic elongation to fracture is below 0.1% at temperatures smaller than 500°C. The texture in this sheet is also very small. The increase in strength is seen to originate from the C-addition and - since carbon-containing precipitates were not found - from a solid solution strengthening mechanism.

Currently the creep properties of both alloys are under investigation and will be published in a forthcoming paper. Future developments will concern mainly the C-containing sheet. Modifications of the primary annealing treatment shall result in a microstructure without lamellar colonies and the precipitation of carbon containing phases. The response of the tensile properties to these microstructural changes will be of particular interest.

Acknowledgement:

The TEM studies were performed at the TEM of the Erich-Schmid-Institute of Materials Science (Leoben, Austria). Technical support from Jörg Thomas (Leoben) is kindly acknowledged.

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Figure Captions

Fig. 1: SEM images in BSE-mode of the microstructures of Ti 45Al 5Nb (as-hiped (a), as-rolled (b), after primary annealing of the sheet (c)) and Ti 45Al 5Nb 0.5C (as-hiped (d), as-rolled (e), after primary annealing of the sheet (f)). Rolling direction horizontal; the scale is identical for a – f.

Fig. 2: Vickers hardness values of Ti 45Al 5Nb and Ti 45Al 5Nb 0.5C after different processing steps.

Fig. 3: Recalculated (001)-polfigures of γ -TiAl (a - f) with positions of rolling texture components (c - f) and (0001)-polfigures of α -Ti(Al) (g - j) with transversal (T) and basal (B) texture components.

Fig. 4: TEM micrographs of the as-rolled (a) Ti 45Al 5Nb and (b) Ti 45Al 5Nb 0.5C sheets.

Fig. 5: Bright field TEM image and corresponding diffraction pattern of a ($\alpha_2 + \gamma$) lamellae colony found in the primary annealed Ti 45Al 5Nb 0.5C sheet.

Fig. 6: Variation of the 0.2% yield strength with temperature in sheets for three different alloys. All samples were tested in rolling direction.

Fig. 7: Variation of the ultimate tensile strength (UTS) and 0.2% yield strength (YS) in rolling direction (RD) and perpendicular to the rolling direction (TD) with temperature for (a) Ti 45Al 5Nb and (b) Ti 45Al 5Nb 0.5C sheets.

Fig. 8: Variation of the plastic elongation with temperature in Ti 45Al 5Nb and Ti 45Al 5Nb 0.5C sheets tested parallel (RD) and perpendicular (TD) to the rolling direction.