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Near Conventional Forging of Titanium Aluminides

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In this paper detailed investigations on the applicability of a near conventional forging route for the fabrication of titanium aluminide (TiAl) components are presented. Microstructural characterizations have been carried out on two alloys in different stages of the manufacturing process. Simultaneously, the chemical compositions as well as the thermo-physical and thermo-mechanical properties of the alloys have been determined. All results are correlated to various forging trials on an industrial scale forging equipment. Based on the results an optimized pathway for the forging of alloy TNB-V5 (high Nb containing alloy developed by GKSS) is presented in detail. Combining the results from basic alloy investigations and industrial forging trials a very narrow hot working-array was determined for alloy TNB-V5, which can show severe restrictions due to slight changes in the chemical alloy composition. However, when using new alloy, TNM-B1, which contains additional β-stabilizing elements, instead of the TNB-V5, the forging window can be widened with respect to billet temperature, die temperature and die speed. Thus, a stable "near conventional" manufacturing process for aero engine components from TiAl alloys becomes feasible.

Keywords: *TiAl, β-phase, hot-working, phase transition, turbine blades*

I. Introduction

The continuous demand for weight reduction and higher engine efficiencies in aerospace and energy industries pushes the materials applied today towards their limits. Therefore, these industries have a strong need for developing novel light-weight materials which can withstand temperatures up to 800°C maintaining reasonable mechanical properties[1)]. Gamma titanium aluminides (γ-TiAl based alloys) are certainly among the most promising candidates to fulfill the required thermal and mechanical specifications. The latest generation of γ-TiAl based alloys exhibits a rather complex alloy composition including a high amount of Niobium[2)]. Due to their alloy composition they exhibit excellent mechanical strength and oxidation properties exceeding the specific strength of Nickel base alloys up to temperatures of 800°C.

However, up to now there is no well established processing route which guarantees an economic and continuous supply of aerospace and stationary turbine components made of γ-TiAl[3-5)]. Today, there are two main industrial approaches for the fabrication of turbine blades from γ-TiAl. On the one side, turbine blades are made by investment casting followed by hipping and additional machining. On the other side, the path of blade manufacturing via ingot pre-material and further wrought processing is assessed[2,6)]. Up to now neither of the two technologies ends up with near net shape components. Therefore, a further machining step, mainly by chemical milling needs to be performed to obtain blades with the desired final shape. Concerning the wrought processing route most of the approaches reported in literature include an isothermal forging step[6)]. Isothermal forging of γ-TiAl is performed at high constant temperatures, requiring special dies and environmental conditions which increase the manufacturing costs.

To Increase the economic feasibility of the wrought Processing route for γ-TiAl components a "near conventional" forging of γ-TiAl material using hot dies has been developed.

2. Process

The industrial forging process investigated in this work is called "near conventional", because hot-working is performed on a conventional hydraulic press (Fig. 1) without any special isothermal forging equipment. However, due to the high strain rate sensitivity of γ-TiAl and thus the necessity of very low deformation speeds it is vital to pre-heat the dies prior to forging to avoid excessive cooling during deformation. The die temperature during forging is approximately 400 — 800°C below the billet temperature. The hydraulic press has been specially automated to provide the possibility to run at low die speeds and to allow an exact control of die speed, position and temperature during the whole process.

The billets for the forging experiments are produced via the ingot metallurgy route including a double vacuum arc re-melting process to obtain high chemical homogeneity[7)]. The resulting ingot with 230 mm in diameter is steel canned and afterwards extruded to a diameter of 45-55 mm of TiAl core material. After extrusion a stress-relieve heat treatment is applied. The material is cut into cylinders between 40 and 80 mm in length and mechanically turned to a diameter of 40 or 50 mm. Prior to forging a heat

Figure 1. Conventional hydraulic press equipped with automation for low die speeds and exact die speed, position and temperature control (left). Forging of γ-TiAl on the hydraulic press (right).

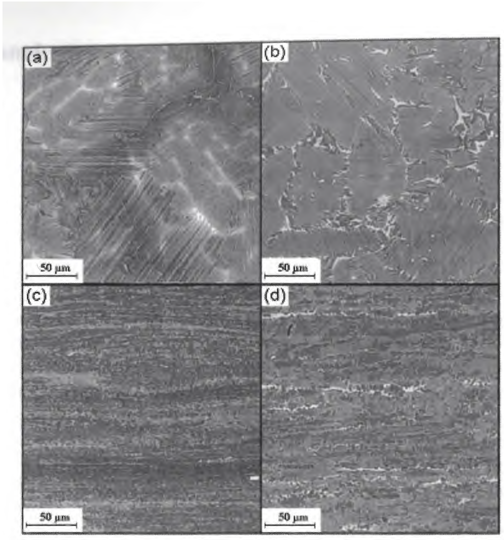

Figure 2. SEM microstructures in BSE-mode of (a) TNB-V5 as-cast, (b) TNM-B1 as-cast, (c) TNB-V5 hot-extruded, and (d) TNM-B1 hot-extruded.

shielding layer is applied to the billet to reduce heat loss during transfer from the furnace to the press as well as during the forging process. The TiAl billet is heated up to forging temperature in an electric furnace under argon atmosphere. After holding on temperature for a defined time the billet is manually transferred into the press within 5 seconds and the forging process is started. The total contact time between the billet and the dies can last up to 60 seconds depending on die speed and total stroke.

3. Alloys

Hot formability by "near conventional" forging was investigated for two γ-TiAl alloys. The first one is TNB-V5 with the chemical composition of Ti-45Al-5Nb-0.1B-0.1C (in at%) developed by GKSS. The second alloy is TNM-B1 which was developed after a detailed investigation of the hot-workability of TNB-V5 to achieve an increased hot-workability compared to TNB-V5, but keeping the desired mechanical properties[8]. The nominal composition of the TNM-B1 is Ti-43Al-4Nb-1Mo-0.1B (in at%).

In Fig. 2 the microstructure of the two alloys after casting and after extrusion is shown. After casting and air cooling TNB-V5 reveals a very coarse grained fully lamellar microstructure consisting of α_2/γ-grains showing strong segregation effects (Fig. 2a). For TNM-B1 in as-cast state, the grain size is smaller and the grains are surrounded by β(B2)-phase (light grey contrast). The β-phase has two beneficial effects: Firstly, the solidification of the alloy occurs via the β-phase which results in a more homogenous chemical element distribution. Secondly, the β-phase increases the hot-workability of the alloy at forging temperature.

After extrusion the average grain size of TNM-B1 is lager when compared to TNB-V5 which may be attributed to the same extrusion temperature used for both alloys. As will be shown below, the α-transus of TNM-B1 is about 45 °C below that of TNB-V5 which means that alloy TNM-B1 was extruded closer to α-transus, causing increased grain growth after recrystallization during extrusion.

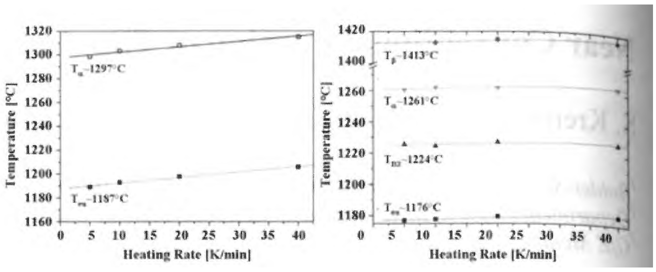

Figure 3. Phase transition temperatures T_{eu} and T_α determined from DSC as a function of the heating rate for TNB-V5 (left) and TNM-B1 (right).

3.1 Determination of Phase Transition Temperatures

For successful hot-working of γ-TiAl alloys it is essential to know the alloy phases which are present at a chosen processing temperature. Therefore, the phase transition temperatures during heating and cooling of the two alloys have been determined by differential scanning calorimetry (DSC) measurements at different heating rates using a Setaram Setsys Evolution. The equilibrium phase transition temperatures are obtained from a linear extrapolation of the measured transition temperatures towards a heating rate of 0 K/min. The phase transitions for different heating rates and the corresponding extrapolated equilibrium temperatures are presented in Fig. 3a and 3b for TNB-V5 and TNM-B1, respectively.

From the DSC measurements two main characteristic differences can be observed between the two alloys. Firstly, the α-transus temperature of TNM-B1 is about 45°C lower than the one obtained for TNB-V5 which can be attributed to the lower Al content in TNM-B1. The second difference, but also the most important one concerning the hot formability of TNM-B1, is the observation of the phase reaction B2/β4↔γ[9], pointing out the presence of a cubic body centered phase at temperatures between the eutectoid and the α-transus temperature, as well as above the α-transus temperature. The presence of a β- or B2-phase in the TNM-B1 was already demonstrated by thermodynamical calculations using the CALPHAD method[10]. As stated above, the excellent deformability of the unordered bcc β-phase has a very beneficial influence with regard to the hot-workability of the alloy.

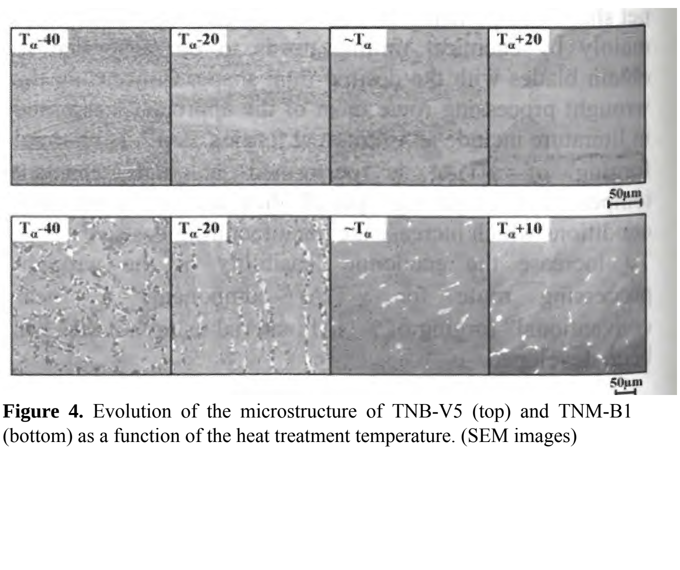

Figure 4. Evolution of the microstructure of TNB-V5 (top) and TNM-B1 (bottom) as a function of the heat treatment temperature. (SEM images)

3.2 Heat treatment experiments

To support the results of the DSC measurements heat treatment experiments were performed on both alloys around the α-transus temperature (T_α). In Fig. 4 the microstructural evolution during heat treatment of the two alloys is summarized for different temperatures.

At temperatures far below T_α a duplex microstructure is observed after air cooling consisting of γ-grains (dark-grey contrast) and α_2/γ-grains (light-grey) in case of TNB-V5 and of a mixture of γ-grains (dark-grey), α_2/γ-grains (light-grey) and β-grains (white) in case of TNM-B1. From these investigations on the microstructural evolution as a function of temperature together with the theoretical calculations from CALPHAD, it could be demonstrated that for the TNM-B1 the amount of β-phase decreases from the eutectoid temperature towards the α-transus temperature, exhibits a minimum at T_α, but increases again for temperatures above T_α [8].

From the microstructural characteristics of TNB-V5 it is obvious that high chemical inhomogeneities exist at the micro-scale, resulting in residual γ-grains aligned along the extrusion direction even for heat treatment temperatures of 20°C above the macroscopic T_α. This microstructural inhomogeneity is not observed in case of the TNM-B1 alloy, where already at 10°C above T_α all γ-grains are dissolved and transformed into α-grains, providing at room temperature a fully lamellar structure with some residual βphase. These reduced microstructural inhomogeneities in case of TNM-B1 can be attributed to the solidification path which occurs via the β-phase for TNM-B1 alloy, but via a peritectic reaction for TNB-V5.

Above the α-transus temperature alloy TNB-V5 enters a single α-phase field resulting in remarkable grain growth. In case of alloy TNM-B1, above T_α a (α + β) phase region is present. From the distribution of the β-phase around the a_2/γ-grains for a temperature of 10°C above T_α (Fig. 4) it is obvious that the β-grains suppress an excessive growth of the α-grains. This situation provides a thermally more stable microstructure in case of TNM-B1, when hot working of the alloy is performed close or even above T_α.

4. Forging Experiments

The hot-workability of the two γ-TiAl alloys was evaluated by performing various forging experiments on the industrial hydraulic press shown in Fig. 1 using a parameter matrix of different billet temperatures, die temperatures and die speeds. All the experiments were supported by simulation using the finite element simulation software Deform 2D and 3D. From the simulation it was possible to obtain information on the temperature, strain, strain rate, and stress distribution during the experiment. This information is essential, because due to the non-isothermal nature of the forging process, all the parameters in the experimental matrix depend on each other (e.g. cooling of billet and die during the process). Thus, when decreasing the die speed to reduce the strain rate during the deformation, the temperature drop in the billet and the dies is more severe and can cause a failure of the forging

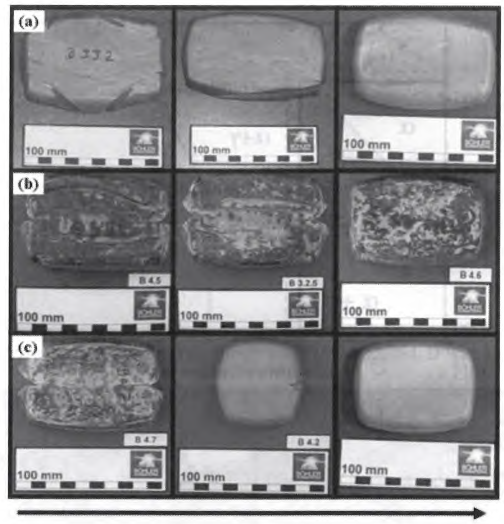

Figure 5. Overview of selected results from the forging experiments on the TNB-V5 showing the optimization of the forging process in the direction of the arrow by (a) increasing billet temperature, (b) increasing die temperature, and (c) decreasing die speed.

experiment (appearance of cracks or even breaking of the billet).

To evaluate the feasibility of a blade forging process where upsetting (pressing along extrusion direction) and side-pressing (pressing perpendicular to extrusion direction) of the material is necessary, the experimental setup included upsetting experiments using cylinders with a diameter of 40 mm and a length of 40 mm as well as side pressing experiments with cylinders 40 mm in diameter and 40-80 mm in length.

Fig. 5 presents a compilation of selected experimental results which shows the optimization of the forging process by means of side pressing trials conducted on alloy TNB-V5. In order to avoid heavy grain growth during the forging experiments the forging temperature was kept below T_α. The die surface temperature was varied between 400 and 800°C below billet temperature and the die speed was below 5 mm/s. However, only by using very high billet and die temperatures as well as very low die speeds it was possible to obtain the results shown on the right hand side of Fig. 5. This means that as far as the alloy TNB-V5 is concerned the process window for a "near conventional" forging process is very narrow.

Despite the fact of a very narrow forging window for TNB-V5 it has to be considered that the Aluminum content can vary by about ±0.5 at% within a single ingot (thus within one extrusion) and by about ±0.7 at% from ingot to ingot according to the specifications of the materials supplier. This specification is valid for macroscopic chemical variations, whereas the microscopic variations can be even much higher[7]. The effect of such variations in the chemical composition is summarized in Fig. 6. The solid line in the phase diagram shows the nominal Al content of TNB-V5, whereas the dashed lines represent a variation in Al content in the range of ±0.5 at%. This deviation from the nominal composition can be converted into a shift of T_α of about 31°C (from the highest to the lowest Al content). The effect of such a shift in T_α on the

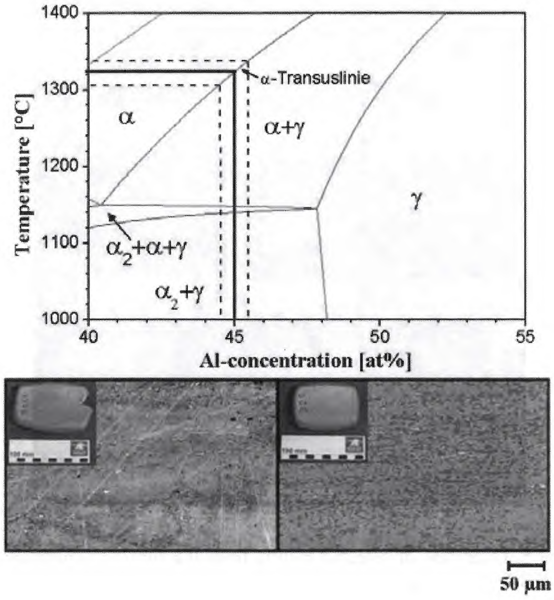

Figure 6. ***Top:*** Calculated phase diagram of the alloy TNB-V5 showing the effect of a variation of the Al composition of ±0.5% (dashed lines) around the nominal Al composition (solid line) on the T_α. ***Bottom:*** Microstructural differences due to a variation of the chemical composition of the forged billets (inset) using identical forging parameters. (SEM images)

forging behavior is shown in the SEM images of Fig. 6 which were taken after forging of two TNB-V5 billets using exactly the same parameter set. It is obvious that for the left billet shown in Fig.6 the forging temperature was far above the T_α, whereas the other billets exhibits a duplex microstructure which indicates a forging below T_α. A chemical analysis of the two billets revealed a 1.2 at% lower Al content for the left billet, explaining the observed microstructural differences and, thus, the failure of the material during forging.

From the whole set of forging experiments conducted on alloy TNB-V5 it was found that the combination of a very narrow forging window along with the strong influence of a alloy composition on T_α and the fact that forging is performed very close to this T_α, heavily constrains the establishment of a robust industrial "near conventional" process.

Considering all the observations made on TNB-V5, the TNM-B1 alloy was developed and investigated with regard to the hot-workability using the described "near conventional" process. In Fig. 7 representative results from a set of various TNM-B1 forging experiments are shown together with a forged blade pre-form. Analysing the forging experiments, it was found that the forging window of alloy TNM-B1 is significantly extended compared to TNB-V5. The observed widening can mainly be attributed to the presence of the bcc β-phase at forging temperature. Further, it is possible to forge this alloy above T_α, where extensive grain growth is suppressed due to the presence of a (α+β) phase region. Therefore, variations in the chemical composition of the alloy and their effect on the shift of T_α explained above are not critical in case of alloy TNM-B1.

Thus, the hot-working behaviour of TNM-B1 has provided the foundation to establish a robust industrial forging process for TiAl components using conventional

Figure 7. TNM-B1 billets after side-pressing to an overall strain of 0.7 (a), upsetting to an overall strain of 0.7 (b), and side-pressing to an overall strain of 1.3 (c). (d) Grinded TNM-B1 blade pre-form manufactured by upsetting both sides of a cylindrical bar and consequent side pressing to the final shape.

forging equipment with minor and inexpensive modifications.

5. Conclusions

The hot-workability of two alloys, TNB-V5 and TNM-B1, was investigated, combining basic characterizations including SEM studies, DSC measurements, chemical analysis, and theoretical simulations with the results of "near conventional" forging experiments conducted on an industrial scale. It was found that the forging window of TNB-V5 is very narrow and can be heavily shifted due to variations in alloy composition. Therefore, a robust industrial process using conventional forging equipment could not be established for this alloy.

The use of the alloy TNM-B1 significantly extended the forging window with regard to billet temperature, die temperature, and die speed, providing the foundation for a robust industrial forging process. The technical feasibility of manufacturing a forged turbine blade component in a three step forging process was demonstrated.

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