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REVIEW PAPER

Isothermal and Near Isothermal Processing of Titanium Alloys

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ABSTRACT

Isothermal and near isothermal forging are specialised metal-forming techniques used for producing critical aeroengine components from advanced materials such as titanium alloys. The process is used to produce net/near-net shape components, leading to optimum utilisation of materials. As titanium alloys are highly sensitive to temperature and strain rate, these processes help to deform these under slow and controlled strain rates. Further, these processes can be combined with other conventional and non-conventional metal-forming processes to refine the microstructure. For example, multi-axial isothermal forging coupled with pack rolling can be used to produce thin sheets out of titanium alloys with sub-micron grain size. The refined structure exhibits superplastic characteristics at low temperatures and high strain rates. This lower temperature superplastic characteristic can be exploited to establish technologies for producing various components. The paper highlights the capabilities of isothermal forging process and its variants.

Keywords: Isothermal processing, near-isothermal processing, forging process, multi-axial isothermal forging, Titanium alloys, aeroengine components, metal-forming techniques, superplastic materials, structural materials

1. INTRODUCTION

Titanium and titanium alloys are used extensively both as structural materials and engine component materials in aerospace applications because of their high specific strength which results in significant weight reduction. These also have high thermal stability, excellent corrosion resistance and are very much amenable for composite production which can be used as structural materials. Titanium alloys are generally vacuum-arc melted[1,2] and cast into ingots, which are subjected to primary and secondary thermomechanical processing to obtain semi-finished or finished products. The intent behind such thermomechanical processing is to tailor the microstructure to obtain the desired combination of static and dynamic mechanical properties[3-5]. The high temperature dependence on flow stress and high strain rate sensitivity of these materials makes thermomechanical processing more cumbersome. Further, the ever increasing demand for higher performance of titanium alloys in aerospace sector requires new processing sequences to be developed to tailor the microstructure. This has led to the development of unique forging methods like isothermal and near-isothermal forging or hot-die forging[6-8.] These forging processes help to exercise a high degree of control over the amount of deformation that can be imparted which influences the microstructure evolution and thereby control the properties that can be achieved. This paper highlights how isothermal and near-isothermal forging processes and their variants produce various finished and semi-finished airworthy components from titanium alloys.

2. ISOTHERMAL / NEAR-ISOTHERMAL FORMING PROCESSES

Metal-forming processes, which help to exercise a high degree of control over the deformation behaviour, microstructure evolution, and hence control the properties that can be achieved, are called as advanced forming processes[3,4]. Isothermal forging, near-isothermal or hot-die forging, multiaxial isothermal forging plus pack rolling, superplastic forming and diffusion bonding are some of such processes being used to produce various components from titanium alloys.

In Isothermal forging, the die and the workpiece are maintained at the same temperature throughout the forging cycle[3,4]. As the die and workpiece are maintained at the same temperature, die chilling is eliminated, thereby resulting in uniform deformation of the material. The inherent advantage of the process can be used to produce net to near-net shape components with fewer processing steps. Further, forgings with small corners and fillet radii, less draft angles and smaller forge envelopes can be produced, leading to optimum utilisation of the materials[3,4]. High degree of control can be exercised over the processing parameters and forging can be carried out at very slow strain rates, thereby reducing the load required to process strain-rate sensitive materials like titanium alloys. All these lead to a high degree of consistency in the structure and property that can be achieved from one forging to another. Further, as low deformation rates are used, the forging pressure required is lower than in conventional forging. This aids to extend the life of expensive dies such as IN100, Mar M 200 and

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molybdenum dies[3,4].The process is expensive relative to the conventional forging. Therefore the process can only be justified when long-term benefits outweigh the high startup and operational costs. The process requires expensive dies and hence the design and fabrication of dies should be carried out only after extensive validation of the design using numerical techniques such as finite element analysis.

Near-isothermal or hot-die forging is a variation of isothermal forging where the dies are maintained at about 100 °C to 200 °C less than the workpiece temperature throughout the forging[3-5]. The process facilitates the use of low-cost dies. Unlike the isothermal forging, some amount of die chilling is observed with hot-die forging but it does not produce adverse results as in conventional hot forging where the dies are maintained at very low temperatures (200 °C - 400 °C) compared to the workpiece. Material input weight can be reduced to as much as 60 per cent with proper perform and die designs. The technique is most suited for *Ti*, *Ni* and other exotic materials[3-5].

Multi-axial isothermal forging[6, 7] is another process used to refine the microstructure of strain-rate sensitive titanium alloys. The grain-refined material is then pack-rolled to produce thin sheets and foils[8-9], can be used to fabricate various structural and aeroengine components. The grain-refined microstructure show enhanced superplasticity at low temperature and high strain rate which can be used to form complex shapes from titanium alloys or can be combined with diffusion bonding to produce corrugated bonded sheets or honeycomb structures. In superplastic forming and diffusion bonding processes, two surfaces are placed under intimate contact which under pressure and temperature (similar to superplastic deformation) unite to form a monolithic homogenous piece. All these processes are being used to produce various products in at DMRL.

3. NEAR-ISOTHERMAL FORGING OF AEROENGINE COMPRESSOR DISCS

3.1 Material

Near-alpha titanium alloy LT26A with a nominal composition of *Ti*-6*Al*-5*Zr*-0.5*Mo*-0.25*Si* was procured from M/s MIDHANI, Hyderabad. The as-received microstructure of the material is shown in Fig.1. The alloy is used extensively as a disc material in the high pressure compressor region of the aeroengine. The alloy exhibits excellent thermal stability and creep properties up to 520 °C. The β transus of the material was estimated to be 1025 °C. As the alloy has been developed to provide better creep resistance, it is solution treated in the β region and quenched in oil to obtain a basket weave transformed beta (Tβ) structure which exhibits excellent creep and fracture toughness[12-14].

3.2 High Temperature Flow Behaviour

To evaluate the high temperature flow behaviour of the material, small cylindrical samples were prepared from LT26A. The sample dimensions were controlled to maintain a constant height-to-dia ratio of 1.5. Deltaglaze 347M from M/s Achelson Colloids, used as a lubricant and oxidation-resistant agent was applied to the sample. The samples were then subjected to hot compression testing. After testing, the samples were water quenched to freeze the high temperature microstructure. The quenched samples were then subjected to microstructural evaluation to identify the deformation mechanisms. Hot compression testing was carried out on a computer-controlled servo hydraulic testing machine custom built by DARTEC, UK. The machine has capacity of 100 kN and is capable to provide constant true strain rates in the range of 3×10^{-4} /s to 10^{2}/s. The temperature and strain rate selected are based on the alloy composition and the desired end-use, as shown in Table.1.The load stroke curves obtained from the data acquisition system of the machine were converted into true stress-true strain curves and were corrected for elastic strain to obtain the true plastic strain. At low strain rates, the stress-strain curve are nearly flat/steady which indicates that the material behaves as an ideal plastic material. However at high strain rates, the material exhibits flow-softening characteristics. Using the true stress-true strain data, the processing map has been developed following the procedure outlined by Prasad[15-17], *et al.* and one such map is shown in Fig.2. The processing map helps to identify, not only the optimum processing parameters but also the underlying deformation mechanism. It also helps to identify the unsafe region where the material should not be processed.

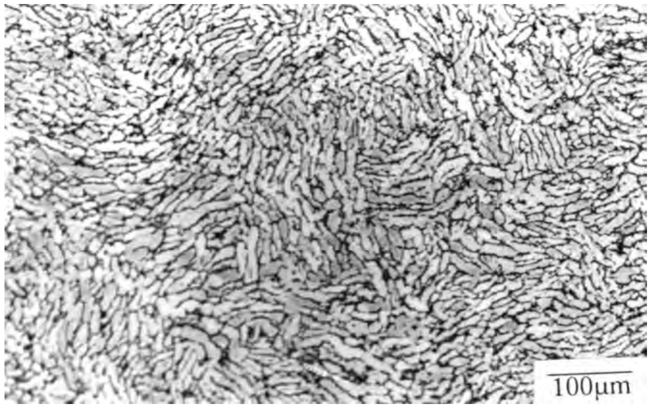

Figure 1. As received microstructure of LT26A.

From the processing map it can be seen that the material exhibits inhomogeneous deformation at high strain rates (Regions a & b in Fig. 2) irrespective of the temperature. The behaviour is typical of lath microstructure which is more susceptible to shear/deformation banding over a

Table 1. Temperature and strain rate domain for compression testing of LT26A

Parameter	LT26A
β Transus (°C)	1025
Temperature (°C)	880, 920, 960, 1000, 1025
Strain rate /s	3×10^{-4}, 10^{-3}, 10^{-2}, 10^{-1}, 10°
Reduction (per cent)	50 per cent = True Strain (ε): 0.694

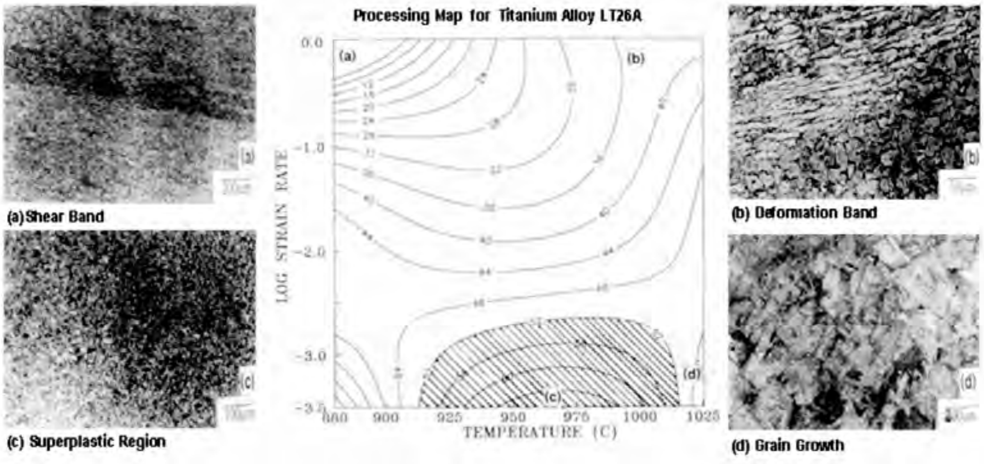

Figure 2. Processing map of titanium alloy LT26A.

larger temperature and strain rate range than the equiaxed primary α plus transformed β microstructure[12-14]. The material exhibits superplasticity between 910 to1010 °C at low strain rates of 3 x 10^{-4}/s to 10^{-3}/s (Region c). The microstructure of the sample deformed in this domain exhibits a fine equiaxed α-β structure. This is the domain where the material should be processed. At high temperatures (close to the β transus) and slow strain rates the material exhibits grain growth (Fig. 2 Region d). Therefore for LT26A, low strain rate corresponds to high workability. If the alloy is forged conventionally using conventional hot-forging process, there is a huge loss in the billet temperature due to longer forging time because of the slow strain rates employed. This temperature loss would result in die chilling and inhomogeneous deformation. Further, as the temperature of the billet decreases, the flow stress increases, which requires higher forging loads to process the material. To avoid these complications, isothermal or near-isothermal forging is preferred.

3.3 Near-isothermal Forging

Based on the sonic profile of the high pressure compressor disc, the forge profile was designed using conventional design principle and the design was validated using finite element analysis studies[18]. The sonic profile of the disc (hatched region) and the forge profile (solid line encompassing the hatched region) of the compressor disc are shown in Fig. 3.Near-α titanium alloy LT26A can be forged into compressor discs in a single step using isothermal or near-isothermal forging technique but the true stress-true plastic strain curves obtained at the safe processing domain for the as received material with lath structure exhibit flow-softening characteristics during initial stages of deformation and eventually reach a steady state. It is believed that under such conditions, the microstructure of the material evolves during processing and finally attains a stable structure after a certain amount of strain[19,20]. If the material is forged into a disc which has varying geometry along its cross section, various regions in the disc will undergo different amounts of deformation, and hence, the resulting microstructure in the disc will not be uniform. Therefore forging is generally avoided under these circumstances when the material undergoes microstructural evolution. To avoid forging in flow-softening region, it was decided to forge the material in two stages. The initial lath structure of the material was converted to an equiaxed structure by upset forging the feedstock. The resulting equiaxed microstructure obtained exhibits ideal plastic behaviour at almost all strain rates. The deformation behaviour of LT26A with initial lath structure and equiaxed microstructure is compared in Fig. 4. The upset forged material with equiaxed microstructure was then used for forging the compressor disc. Thus a two-stage process was devised to forge compressor from LT26A under near-isothermal conditions using IN100 flat and shaped dies maintained at 920 °C. The parameters used for near-isothermal forging and the as forged and sonic machined discs are shown in Table 2. The as forged discs were then subjected to post-processing heat treatment and the microstructure obtained at various locations of the disc was inspected. It was observed that the disc had uniform basket weave transformed β structure throughout the cross-section with an average prior β grain size of 1mm.

The advantage of isothermal/near-isothermal forging

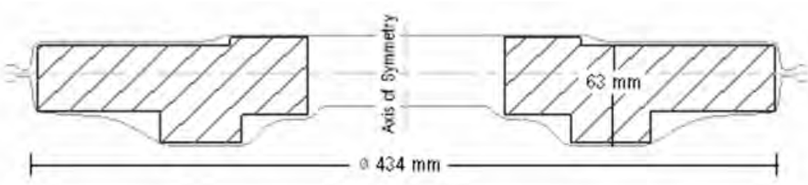

Figure 3. Sonic and forge profile of the compressor disc.

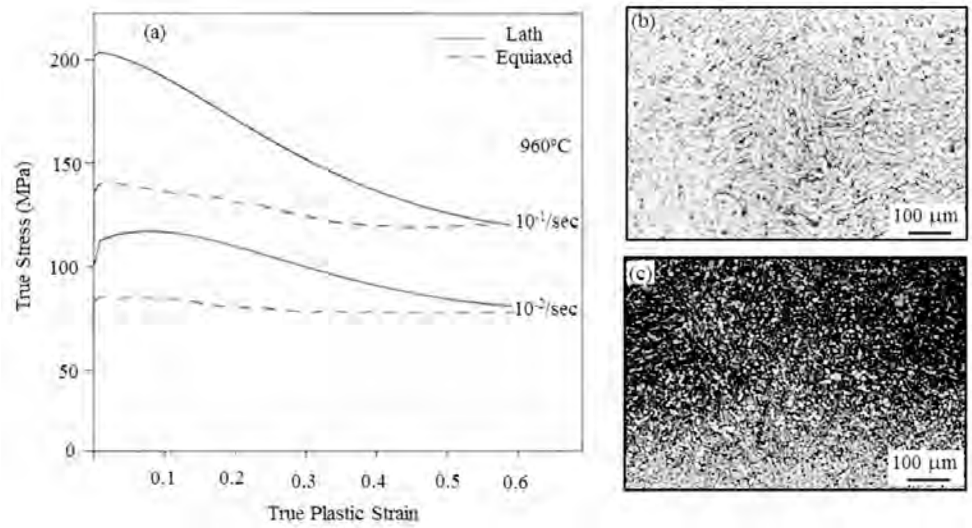

Figure 4. (a) High temperature deformation behaviour of LT26A with (b) lath and (c) equiaxed starting microstructures.

technique can be inferred from this work. The disc was produced to its near-net shape with only two stages of processing. Only one shaped die was used for forging. Microstructure evolution was controlled to achieve uniform microstructure throughout the cross-section of the disc with varying geometries. The compressor discs were required to have a combination of mechanical properties. The discs produced (Fig. 5) met the required property level, as shown in Table 3.

4. MULTIAXIAL ISOTHERMAL FORGING AND PACK ROLLING

Multi-axial isothermal forging is a novel severe plastic deformation technique. The process consists of deforming the material for a specific amount of deformation in all the three directions[6,7]. Judicial selection of process parameters helps in achieving uniform deformation and appropriate control over the microstructure, often isotropic behaviour is achieved. It is also possible to introduce large amount of deformation into the material in multiple processing steps. The method can be used to produce sub-microcrystalline (SMC) materials (grain size < 1µm). The materials with SMC microstructure exhibit increased strength, increased fatigue resistance, superplasticity at low temperature and high strain rates.

Pack rolling is yet another technique wherein heat loss associated with conventional rolling is minimised by encapsulating the material. Absence of roll chill effects results in uniform deformation of the material. Large deformations are possible and usually the surface quality of the rolled material is good. Pack material also helps in minimising the propensity for rolling defects. Both these techniques can be combined to produce sheets and foils with SMC microstructure.

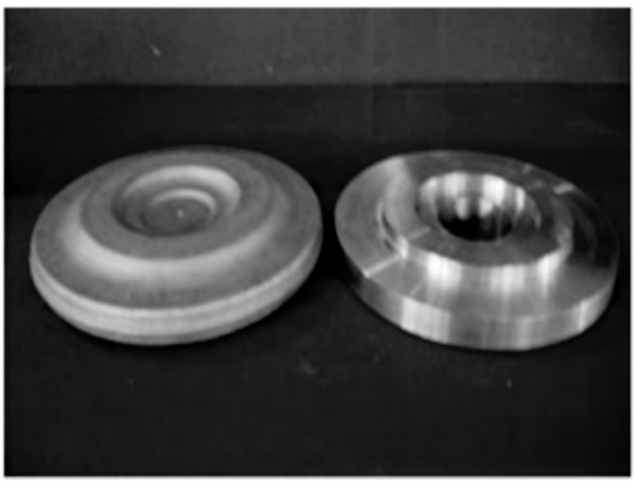

Figure 5. As forged and sonic machined disc forging of LT26A.

4.1 Material

Tripple vacuum, arc-melted and hot-forged titanium alloy BT5-1 (equivalent to IMI 317) was procured from M/s MIDHANI, Hyderabad, in the form of 100 mm dia machined bars. The alloy was melted indigenously, using tripple vacuum, arc-melting technique to obtain the desired homogeneity in the material. The alloy had a nominal composition of *Ti*-5.69*Al*-2.73*Sn*-0.04*Fe*-0.09*C*. The as-received microstructures of the material in the longitudinal and transverse directions of the bar are shown in Fig. 6. The microstructure in longitudinal direction consists of bimodal distribution of long and equiaxed primary α grains in the

Table 2. Processing parameters for LT26A

Parameter	Preforming	Disc Forging
Billet dimension (mm)	ϕ 180 × 345	ϕ 232 × 208
Billet temperature (°C)	950	1000
Die temperature (°C)	920	920
Ram Speed (mm/s)	0.2	0.2
Deformation (per cent)	40	Till 2 mm flash

Table 3. Properties obtained from the disc forgings

Property	Specification	Results obtained	
		Radial	Tangential
Tensile (room temperature)			
Proof strength (Mpa)	≥ 850	922 - 967	971- 987
Ultimate tensile strength (Mpa)	≥ 990	995 - 987	1017 - 1056
Elongation (%)	≥ 6	7 - 13	8 - 13
Reduction in area (%)	≥ 15	13 - 24	14 - 22
Tensile (520 °C)			
Proof strength (Mpa)	≥ 480	486 - 550	497 - 555
Ultimate tensile strength (Mpa)	≥ 620	625 - 654	628 - 711
Elongation (%)	≥ 9	9 - 14	11 - 17
Reduction in area (%)	≥ 20	30 - 39	14 - 22
Notch tensile			
N/P ratio	≥ 1.35	1.51 - 1.68	1.50 -1.68
% Creep strain			
(450 °c /450 Mpa /100 h)	0.15 Max	0.10 - 0.13	0.10 -0.14
Post creep tensile			
Proof strength (Mpa)	≥ 850	971 - 1026	1009-1048
Ultimate tensile strength (Mpa)	≥ 990	995 -1071	1026-1071
Elongation (%)	≥ 6	6-9	7-9
Reduction in area (%)	≥ 10	13-20	12-24
Low cycle fatigue			
(85 to 850 Mpa)	8000 cycles	8297-15001	8742-15522
Cycle: 10-20-10-20 s			
High cycle fatigue			
440 mpa	-	0.66×10^7 - 1.62×10^7	1.0×10^7 - 1.71×10^7
Fracture toughness			
K_{1c} (Mpa $m^{1/2}$)	-	67-84	

transformed β matrix. The alignment of primary grains in the direction of working can also be seen. In the transverse direction, the material has predominantly elongated primary α grains (lath structure) randomly distributed in the transformed β matrix. The β transus of the material was estimated to be at 1020 °C.

4.2 Multiaxial Isothermal Forging

Two billets of dimension 70 x 70 x 70 mm^3 were extracted from the as-received material with a dia of 100 mm and designated as samples A and B. The sides along the orthogonal direction were labeled as longitudinal (L), transverse (T) and short transverse (ST). The longitudinal direction of the cube was labeled in such a way that it coincided with the longitudinal axis of the as-received bar. Billets were coated with Deltaglaze 347M to avoid oxidation and also to reduce friction. Multi-axial isothermal forging experiments were designed to achieve the desired grain refinement.

To carry out multi-axial isothermal forging, the billets were heated to the required temperature using a muffle furnace whereas the forging dies were heated to the required temperature using a line frequency induction heater. Based on the conventional thumb rule of 1/2 h soaking time for every 1 inch of ruling section thickness, appropriate soaking time give to the billets. The billets were then transferred to the forge press. The transfer time was ~ 30 s. A deformation speed of 1mm/s was used to deform the billets. The processing temperature was decreased by 100 °C during subsequent processing stages for both the billets. A deformation of 25 percent was imparted to the material in each direction at each processing temperature. After imparting deformation in one direction, the material was rotated into the next direction and the required amount of deformation was imparted. The procedure was repeated for the third direction as well. At the last stage of multi-axial isothermal forging, it was at 650 °C for sample A and 750 °C for sample B. In the final stage of multi-axial isothermal forging, a deformation of about 75 percent was imparted along short transverse direction to reduce the thickness of the billet to 12 mm to facilitate pack rolling. The process schedule used is listed in Table 4. The microstructures of samples A and B obtained after each stage of deformation in all the three directions are shown in Fig. 7. The grain size was measured in each stage using standard planar intercept method. The grain size was calculated using the following equation:

$$Avg.\,grain\ size\ (\mu m) = \left(\frac{lenght}{average}\,intercept\right) * \left(\frac{1}{magnification}\right)$$

A minimum of 10 readings were taken to get a representative figure of grain size in the material. The mean grain size values are listed in Table 5. In stage-I for both samples A and B, the material exhibited elongated primary α in a transformed β matrix in the longitudinal direction. A mixture of equiaxed and elongated primary α was observed in the other two directions. However, the size of the primary α has got reduced when compared to the as-received structure. Refinement in grain size in sample B was higher than that of sample A. It may be due to minor variations in the microstructure of the starting material along the length of the bar. In both the samples, a substantial reduction in the grain size (from 180 μ to 50/40 μ) was observed. During isothermal forging, grain refinement is mainly due to dynamic recrystallisation of the material wherein the lamellar structure gets refined into equiaxed grains. However, the structure in the longitudinal

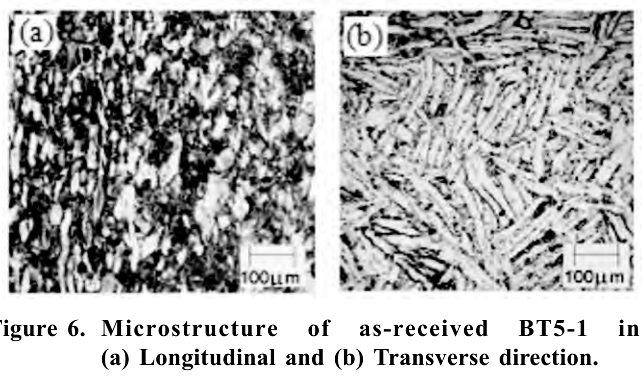

Figure 6. Microstructure of as-received BT5-1 in: (a) Longitudinal and (b) Transverse direction.

Table 4. Process Schedule used for multiaxial isothermal forging

Sample	Temperature (°C)		Speed (mm/s)	Soaking time (min)	Deformations in three directions		
	Billet	die			L	T	ST
A	950	920	1	120	25%	25%	25%
	850	850	1	120	25%	25%	25%
	750	750	1	90	25%	25%	25%
	650	650	1	60	25%	25%	75%
B	950	920	1	120	25%	25%	25%
	850	850	1	120	25%	25%	25%
	750	750	1	90	25%	25%	75%

Table 5. Grain size after multiaxial isothermal forging

Sample	Average grain diameter (μm)	ASTM Grain size number
As received sample	180	2
Sample A		
At 950 °C	58	5
At 850	44	6
At 750	28	7
At 650	14	9
Sample B		
At 950 °C	40	6
At 850	34	7
At 750	26	7

direction was found to be lamellar α, which indicated that the recrystallisation process was not complete.

Microstructure of the samples after stage-II processing essentially consists of equiaxed primary α distributed uniformly in all the three directions in sample B, indicating the absence of grain directionality. However, few pockets of elongated α grains were seen in sample A. Uniformity of microstructure in all the directions indicates the effect of deformation by multi-axial isothermal processing. The size of the α grain was reduced from 59 μm to 44 μm in sample A and from 40 μm to 35 μm in sample B. It was possible to deform at such a low temperature (850 °C) mainly due to improvement in workability as a result of grain refinement. The recrystallisation temperature of the material was also reduced, resulting in an equiaxed structure.

From the microstructure obtained after stage-III processing at 750 °C in sample A was found that the microstructure had refined further. The grain size was found to be reduced from 44 μm to 29 μm. Sample B was subjected to a different deformation schedule; initially 25 per cent reduction was imparted in two directions and in the third direction (ST), 75 per cent of deformation was imparted to the sample. Average grain size in the sample was estimated to be 27 μm. In the final stage of processing, Sample A was isothermally forged at 650 °C for 25 per cent deformation in two directions, and 75 per cent deformation in the third direction. Average grain size in the sample was found to be 15 μm. The final thickness of sample A and B in the short transverse direction after multi-axial isothermal forging was about 12 mm.

4.3 Pack Rolling

Rectangular samples with regular dimensions were extracted from both samples A and B. The extracted samples were then suitably packed in a mild steel plate 6 mm thick. The edges of the pack were welded using a steel electrode. Titanium alloy samples were initially coated (before encapsulation) with glass lubricant on all the surfaces to prevent oxidation. Canned samples were coated with glass lubricant to reduce friction and heated to the rolling temperature using a muffle furnace. The rolling temperature selected was equivalent to the forging temperature in the last stage for each sample to avoid grain coarsening, i.e., 750 °C in the case of sample B and 650 °C in the case of sample A. The samples were soaked for one hour in the furnace. Deformation was restricted to about 25 per cent in each pass. The material was then reheated for 10 min at the same temperature and again rolled to the required deformation. Deformation was continued till a pack thickness of 3.5 mm was obtained. The thickness of the BT 5-1 alloy sheet after pack rolling was 3 mm. Part of this sheet sample was repacked (with 6 mm mild sheets at both the sides) and further rolled at the same temperature to obtain a packed thickness of 3.5 mm. The thickness of BT 5-1 alloy sheet at this stage in the pack was 0.7 mm. Microstructural evaluation was also carried out on this fine sheet. It was also observed that under both these conditions the deformation was smooth without any opening of the pack or cracking at the edges.

Sample A was rolled down from 12 mm plate (obtained during forging) to 3 mm thickness sheet at 650 °C. The microstructure of this rolled sheet is given in Fig. 8. The microstructure resembles a deformed structure with wavy boundaries. However, there is no evidence of recrystallisation under these conditions and the grain size was estimated to be 12 μm. This can be attributed to the reduction in material temperature, as the deformation is by conventional rolling and not by isothermal rolling. Experiments on this sheet material for static recrystallisation may indicate further grain refinement. Sample B was rolled at 750 °C initially to 3 mm thickness (from the 12 mm thick forged plate). The microstructure resembles a deformed structure in the transverse direction with wavy boundaries. In the longitudinal and long transverse directions, the microstructure was a mixed one with elongated and equiaxed α grains. Equiaxed grains were also seen between the deformed layers. The average grain size of the α under this stage was 16 μm. The reduction in the grain size can be attributed to dynamic recrystallisation during processing. A part of this sample (3 mm sheet of sample B) was further packed and rolled down to a pack thickness of 1mm.The thickness of the BT5-1 sheet was 0.7 mm. The microstructure obtained in the 0.7 mm sheet in Fig. 9 shows the presence of fine equiaxed grains uniformly distributed in the matrix.

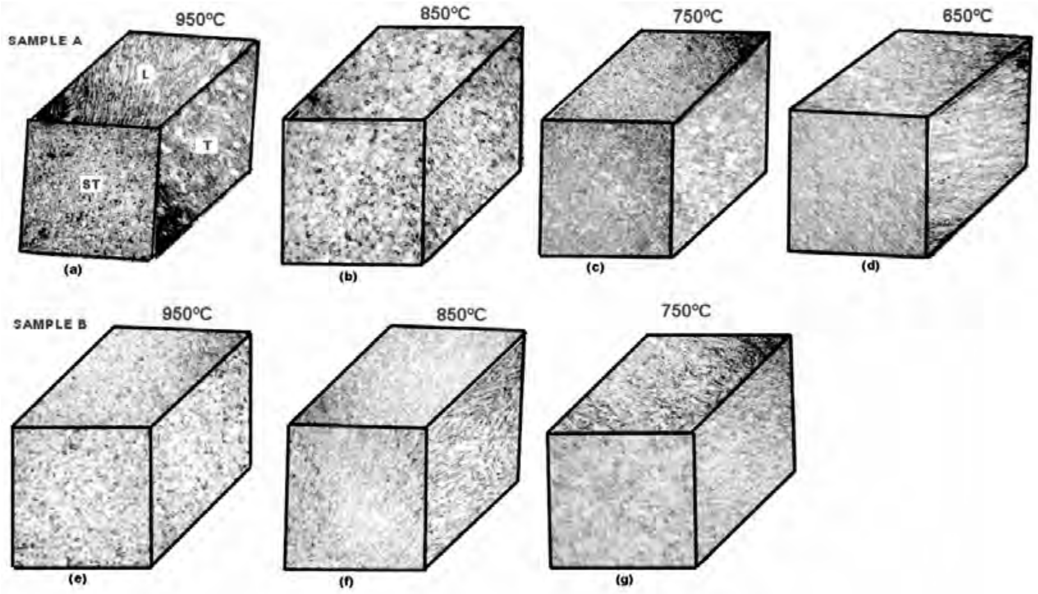

Figure 7. Microstructure after each processing stage in sample A (a, b, c, d) and B (e, f, g).

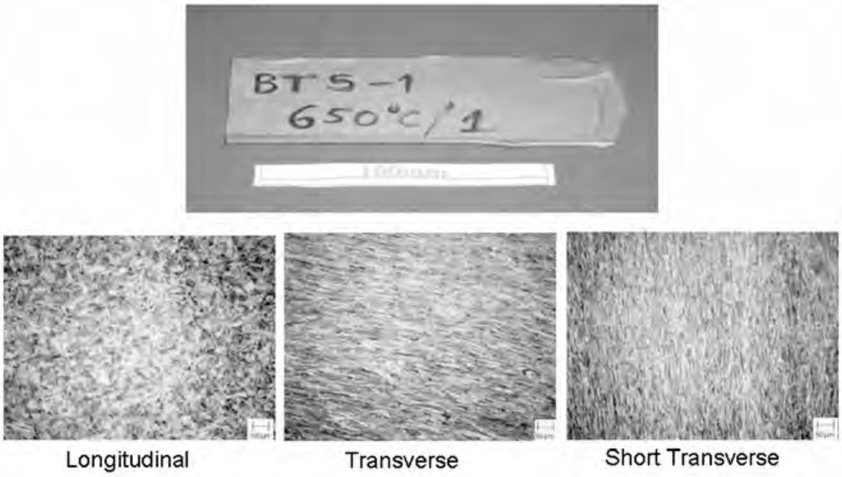

Figure 8. Sample A pack rolled to 3 mm thickness at 650 °C.

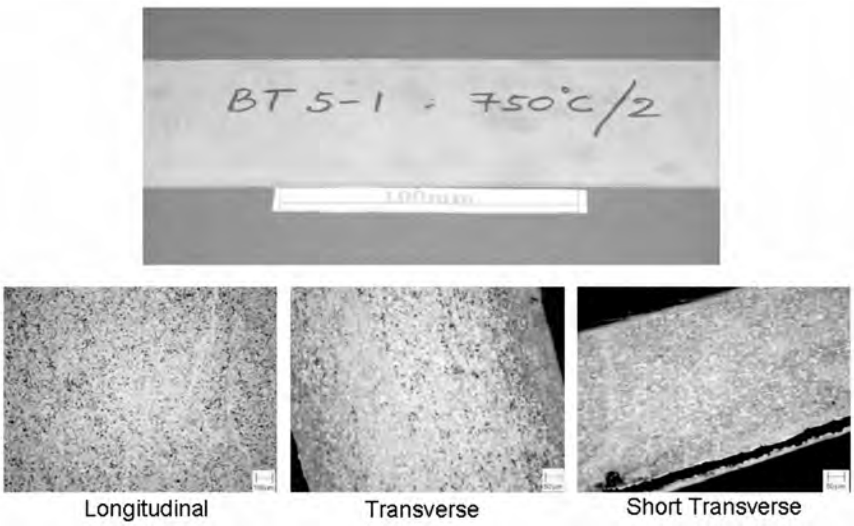

Figure 9. Sample B pack rolled to 0.7 mm thickness at 750 °C.

The structural refinement under these conditions can be attributed to the subsequent heating and deformation processing at this temperature. The grain size was 10 µm. If static recrystallisation was carried, the grain size can be refined further. Thus by combining multiaxial isothermal forging and pack rolling sheets with SMC, microstructures can be produced Multi-axial isothermal forging with progressively lower temperatures alone can refine the grain size. In the current study, the grain size has been refined by an order of magnitude, from 180 µm in the starting material to about 15µm. These sheets with duplex structure and fine grain size have shown enhanced superplasticity at low temperatures and high strain rates. These sheets can be used for producing components like air bottles using superplastic forming techniques.

5. CONCLUSION

Advanced forming techniques such as isothermal and near-isothermal forging, multi-axial isothermal forging, and multi-axial isothermal forging, combined with pack rolling, superplastic forming and diffusion bonding help to exercise a high degree of control over the deformation behaviour, microstructure evolution, and hence, control the properties that can be achieved.

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