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ORIGINAL ARTICLE

Potential of twin-belt-cast EN AW 6082 blanks for the manufacture of wishbone suspension forgings

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Abstract Twin-belt-cast (TBC) strip is offered as forging stock for relatively flat, 2-dimensional forgings such as wishbone suspension. When forged, the fragmented eutectic cells of the TBC blank and the very fine precipitates are aligned in the forging direction with a section grain structure where much of the section is predominantly fibrous. The fibrous grains of the forged component undergo recrystallization and grain growth and are finally replaced by elongated grains in the plastic flow direction during solution heat treatment. Selection of the solutionizing temperature is claimed to be critical. The yield and tensile strength, elongation and hardness all decrease steadily with increasing solutionizing temperature due to grain coarsening. However, both the static and dynamic mechanical properties of the TBC EN AW 6082 blank compare favourably with the extruded forging stock, in spite of coarse grains across the entire section of the forging. After all, both suffer from coarse grains on the surface of the forging. It is thus concluded that the TBC EN AW blanks can be used for the manufacture of relatively flat, 2-dimensional automotive parts such as wishbone suspension components.

Keywords Aluminium alloys · Processing · Forming · Microstructure · Heat treatment

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1 Introduction

Forging is a forming process in which a uniform blank is shaped into a final product by pounding it under high pressure between shaped or flat dies in one or several stages [1]. The automotive and aerospace industries have been using an increasing volume of aluminium forgings instead of castings in highly demanding structural applications [2, 3]. Aluminium forgings offer near net shape, minimum further machining, outstanding mechanical properties and surface finish and are thus favoured in highly stressed parts [4]. A very attractive combination of mechanical properties and corrosion resistance has made age-hardened EN AW 6082 the most popular lightweight aluminium forging alloy for the manufacture of automotive suspension and steering components [http://www.alueurope.eu/wp-content/uploads/2011/11/AAM-Applications-Chassis-Suspension-2-Suspension-parts.pdf, http://european-aluminium.eu/media/1541/aam-products-4-forged-products.pdf].

The most widely used EN AW 6082 forging stock is the round bars which are extruded from DC-cast billets [5]. The DC billets are cast at diameters above 150 mm and have to be extruded to reduce their diameters to dimensions suited to those of the components to be forged from these bars. While very popular, this type of forging stock suffers from several drawbacks owing to quality issues and production costs arising from an extra extrusion step that adds to the processing costs. Friction and high shear strains in the contact zone between the billet and the extrusion die lead to small recrystallized surface grains that tend to grow abnormally when exposed to high temperatures. The extruded EN AW 6082 forging stock thus suffers from a nonuniform structure with fine recrystallized surface grains that are very susceptible to the formation of peripheral coarse grain formation [6–12]. This is a major concern since 6082 forgings are almost always submitted to a solution heat treatment before ageing to the T6 temper. The coarse surface grains that form during the solution heat treatment not only degrade the surface quality but also reduce the impact toughness of the suspension components.

Another type of EN AW 6082 forging stock is also produced with the DC casting process but is not extruded [13–18]. This cast stock not only offers to take care of the coarse peripheral surface grains but also reduces the processing costs owing to the elimination of the intermediate extrusion step. However, lack of commercial cast stock in small diameters for relatively small forging components is a major drawback. This makes an intermediate extrusion process inevitable to manufacture small forgings. The casting-forging route is thus suitable only for heavy forgings since the DC-cast billets are seldom produced at less than 150-mm diameter. Besides, the hot forging of the cast stock with equiaxed grains does not suffice to produce the extent of fibering typically achieved with the forging of the extruded counterpart. Recently, continuously cast rods with less than 50-mm diameter have been tested for their potential as forging stock in the manufacture of lightweight suspension components with favourable results regarding their applicability [19]. DC-cast billets produced with the recently developed near net shape casting process are also available [20–22]. The section shape is the nearest preform shape that is most suited to the final shape of the component to be forged. The blanks sectioned from these billets are readily forged into small automotive components with a minimum number of forging operations.

Forging of relatively small automotive components such as suspension forgings requires cast feedstock smaller in size than those commercially available DC-cast billets. Blanks punched from twin-belt-cast (TBC) EN AW 6082 strips in desired dimensions and shapes could provide an alternative. Twin-belt casters produce continuously cast slab that is generally fed into a hot-rolling mill for deformation into re-roll sheet using the residual heat of the as-cast section [23]. The strip exiting the TBC at 500 °C is rolled through hot-rolling mills and is manufactured as hot coil at the desired thickness.

Table 1 Chemical composition of the TBC EN AW 6082 alloy forging stock (wt%)

Si	Fe	Mn	Mg	Cu	Cr	Ti	Zn	Al
0.804	0.429	0.458	0.737	0.007	0.004	0.033	0.029	97.457

TBC essentially produces discs for deep drawing and slugs for impact extrusion applications for the manufacture of metal tubes. While its use has been mostly confined to less demanding applications where mechanical properties are not at a premium, TBCs may be a viable supplier of aluminium forging stock for the manufacture of relatively simple, flat components. TBC strips can be readily stamped to extract suitable preforms/blanks as forging stock for relatively flat, 2-dimensional suspension parts such as wishbone suspension (Fig. 1). The present work was undertaken to explore the potential of commercial TBC EN AW 6082 strips as forging stock in the manufacture of lightweight wishbone suspension parts for the first time.

2 Experimental

The EN AW 6082 alloy used in the present work (Table 1) was cast industrially with a Hazelett twin-belt caster in the form of a 600-mm-wide, 38.5-mm-thick strip and subsequently hot rolled to 30 mm and finally homogenized at 500 °C for 8 h and cooled slowly to room temperature. Blanks, 30 × 30 × 80 mm, were cut from the TBC strip and were used for industrial-scale forging experiments. The EN AW 6082 alloy blanks were preheated to approximately 520 °C and were forged on a 1600-ton forging press into an experimental part (Fig. 2), before they were immediately quenched in water. All forgings were produced in the standard T6 temper. They were solutionized at three different temperatures for 2.5 h to identify the optimum solutionizing conditions: 510, 520 and 530 °C, and quenched in water before they were artificially aged at 185 °C for 3.5 h.

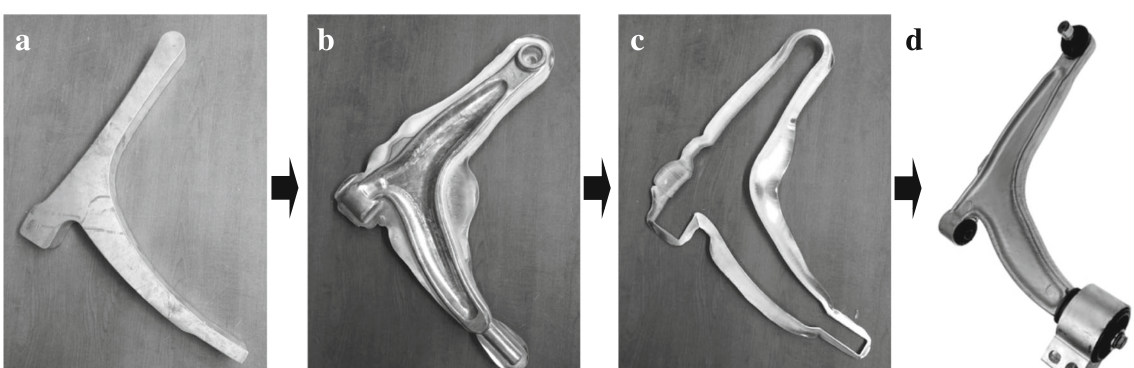


Fig. 1 Typical forging sequence of a wishbone for twin-belt-cast EN AW 6082 blank. **a** Preform cut from a TBC blank. **b** Wishbone part forged from TBC blank. **c** Flash produced. **d** Finished wishbone suspension component

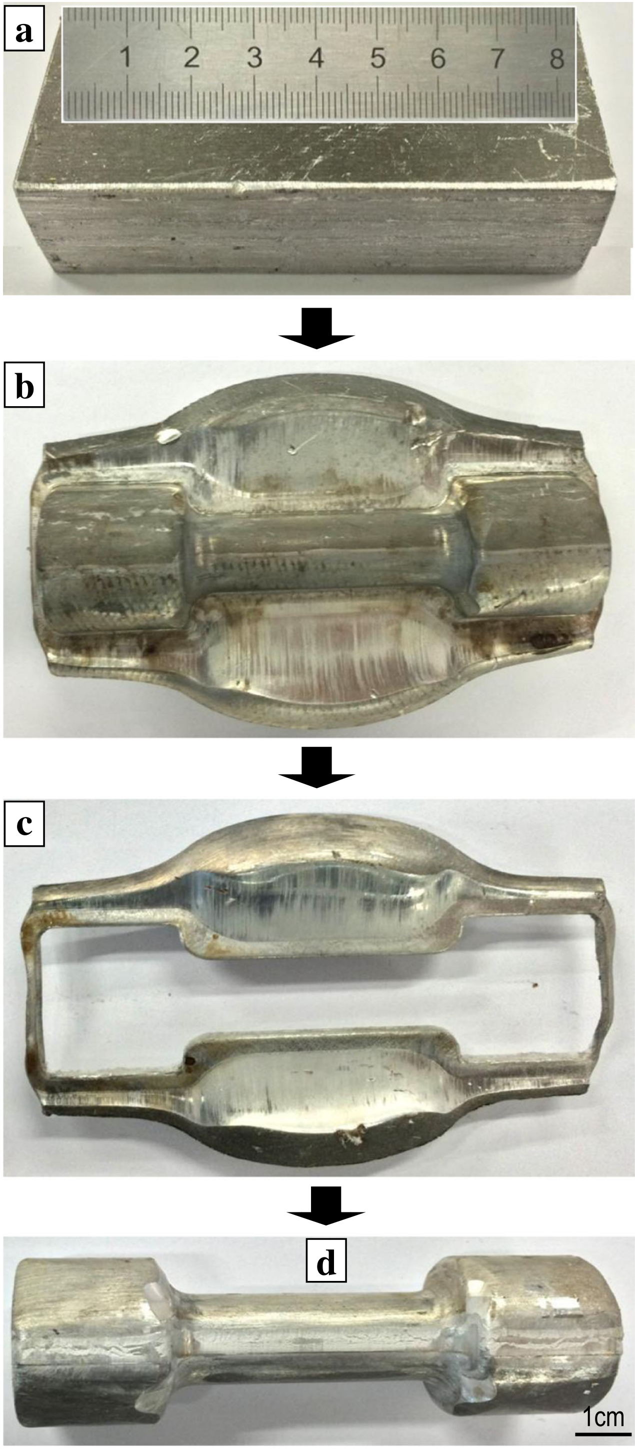


Fig. 2 Forging sequence employed in the present work for the experimental part. **a** Preform cut from TBC EN AW 6082 blank with the following dimensions: 30 × 30 × 80 cm. **b** Experimental part forged from the TBC blank. **c** Flash and **d** forging produced

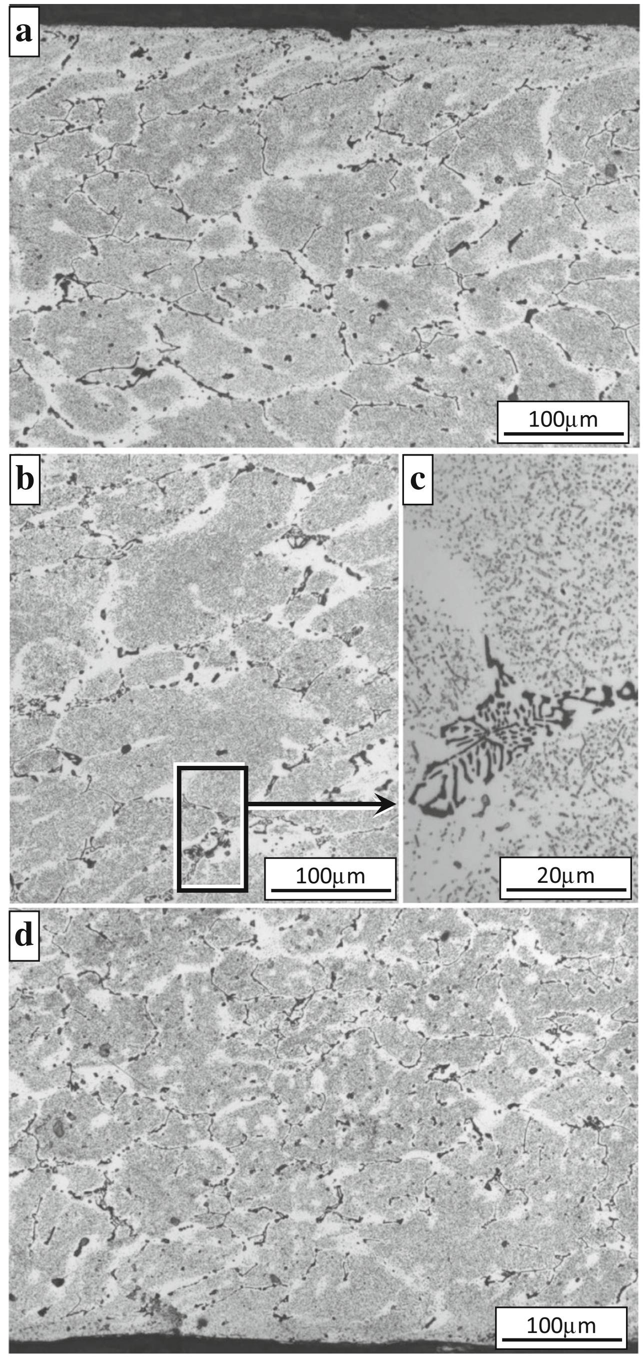


Fig. 3 Microstructure of the TBC EN AW 6082 alloy across the section of the blank. **a** Near the top surface. **b**, **c** The centre. **d** Near the bottom surface. **c** Higher magnification view of the *marked region* in **b**

Samples sectioned from the as-received TBC blanks and the forgings both in the forged and heat-treated states were prepared with standard metallographic techniques: ground with SiC paper, polished with 3-μm diamond paste and finished with colloidal silica. Their microstructures were examined after etching with a 0.5% HF solution using a Nikon MA200 model optical microscope. A second set of samples were etched in a mixture of 32% HCl, 32% HNO_3, 32% H_2O and 4% HF to observe grain structures across different sections of the forgings. These samples were also anodized in Barker's solution, 5 ml HBF_4 (48%) in 200 ml water, and then examined with an optical microscope under polarized light. X-ray

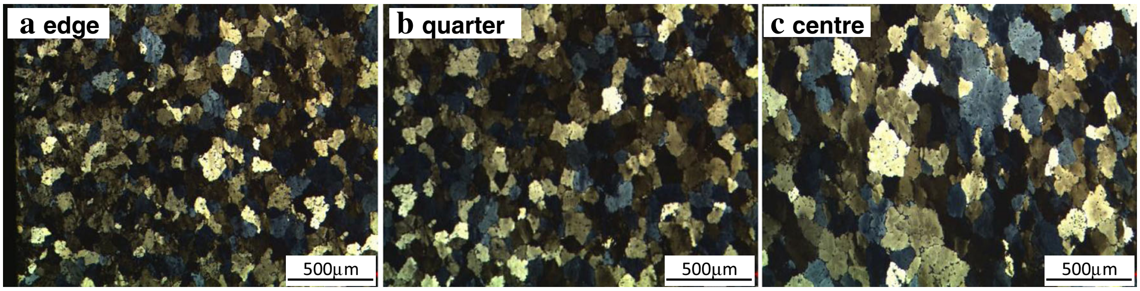


Fig. 4 Grain structure of the TBC EN AW 6082 alloy across the section of the blank. **a** Near the top surface. **b** The quarter depth. **c** Near the centre

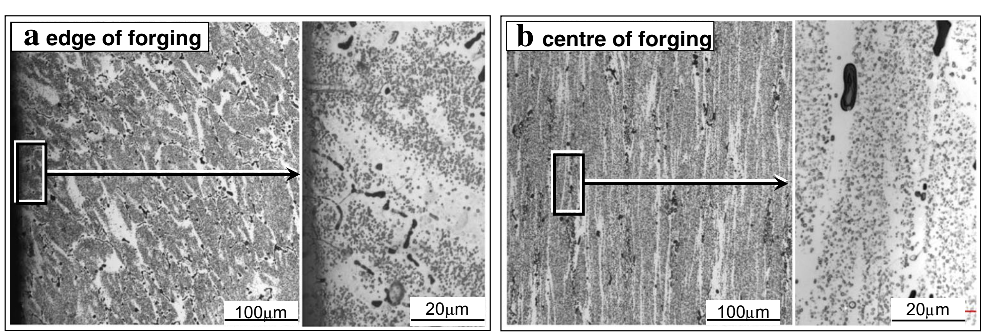
diffraction (XRD) patterns were recorded with a Rigaku D/Max 2200/PC diffractometer equipped with CuK_α radiation to identify the intermetallic particles.

Hardness of the TBC blanks and the forgings before and after heat treatment was measured with a Brinell hardness tester under a load of 250 kgf using a 5-mm-diameter steel ball with a dwell time of 10 s. The tensile tests were performed using a screw-driven ZWICK/ROELL Z250 tensile testing machine in air at room temperature at a cross-head speed of 1 mm/min. An extensometer with a measuring length of 36 mm, attached to the sample, was employed to measure the strain. The 0.2% proof stress was reported as the yield stress. A 300-J pendulum impact testing machine (DMC 6705CE, UK) was used in the Charpy mode at an impact velocity of 5.24 m/s to measure the impact energy values. Fatigue tests were conducted on a MTS Landmark desktop model fatigue testing unit, which is capable of operating at 100 Hz with a maximum load of 15 kN. The fatigue test samples were prepared according to ASTM E466 and cycled at a stress amplitude of 175 MPa, at a frequency of 30 Hz with a minimum to maximum stress ratio of −1. The fatigue test results were reported as the average number of cycles to fracture from five different fatigue tests.

3 Results and discussion

Hardness of the as-received TBC blanks, 40.0 ± 1.7 HB, suggest that it is in the fully soft state owing to the homogenization cycle the TBC sheet has received before shipping. The transverse section of the TBC blanks is shown in Fig. 3. One can see Fe-based intermetallic particles both inside the grains and at grain boundaries and a heavy precipitation inside the grains near the upper surface of the sheet (Fig. 3a). The former were identified by XRD to be of the monoclinic β-Al_5FeSi phase as expected due to an unusually high Fe content. The Fe concentration of the present alloy is nearly twice that of the typical EN AW 6082 forging stock (Table 1). Si is known to be an effective Fe precipitator [24]. Silicon of the present alloy is also in excess of that can be bound in the Mg_2Si phase. The relatively low Mn level of the present alloy and lack of Cr are also responsible for the predominance of the monoclinic β-Al_5FeSi phase. In their work addressing the effects of Mn/Fe ratio and cooling rate on the modification of Fe intermetallic compounds in cast A356 alloy with different Fe contents, Zhang et al. have claimed that both Mn and Cr favour α_c-$Al_{12}(Fe,Mn,Cr)_3Si$ over β-Al_5FeSi [25]. β-Al_5FeSi particles in aluminium wrought and foundry alloys are typically in the form of needles and plates. Their shape is responsible for their negative impact on the mechanical properties. The mechanical properties have been shown to deteriorate as a result of an increase in the size of β-iron intermetallics and explained their results in terms of the β-Al5FeSi platelet size [26]. These β-Al_5FeSi particles are replaced by coarse eutectic cells at the centre plane (Fig. 3b, c). Such eutectic colonies have been identified as the centre plane segregates and are typical of continuously cast strips. The structure near the bottom plane is similar to that near the top plane (Fig. 3d). However, the number of intermetallic particles is higher due to the gravity segregation. The relatively heavier Fe-based particles sink to

Fig. 5 Microstructure of the EN AW 6082 forging. **a** Near the edge. **b** Near the centre. Micrographs on the *right-hand side* are higher magnification views of the *marked regions* in micrographs on the *left-hand side*



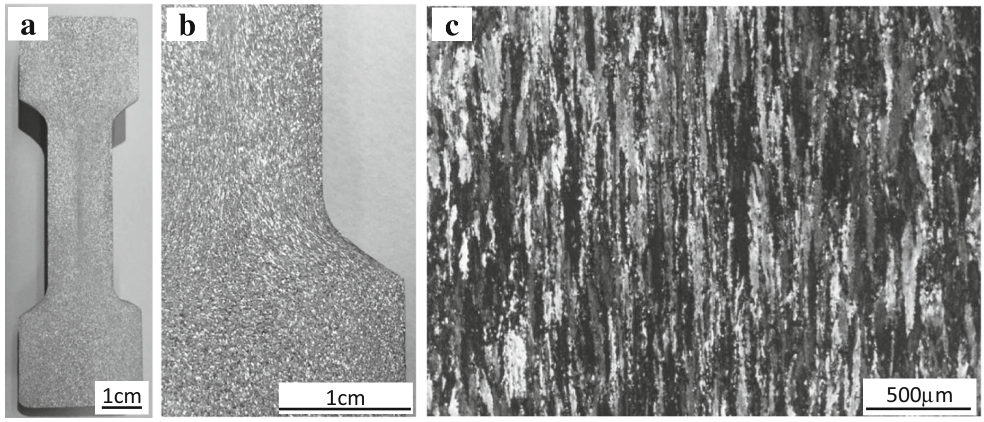


Fig. 6 a Section, **b** lower right corner of the section and **c** the anodized section of the EN AW 6082 forging showing the macrostructure and grain structure

the bottom of the melt pool by the time solidification kicks off, and the majority of these particles end up in the bottom portion of the cast sheet. The intragranular precipitates were identified by XRD to be β-Mg_2Si. The post-homogenization cooling was apparently too slow to retain the Mg and Si in solution.

The grain structure across the section of the TBC sheet is shown in Fig. 4. The grain structure is predominantly

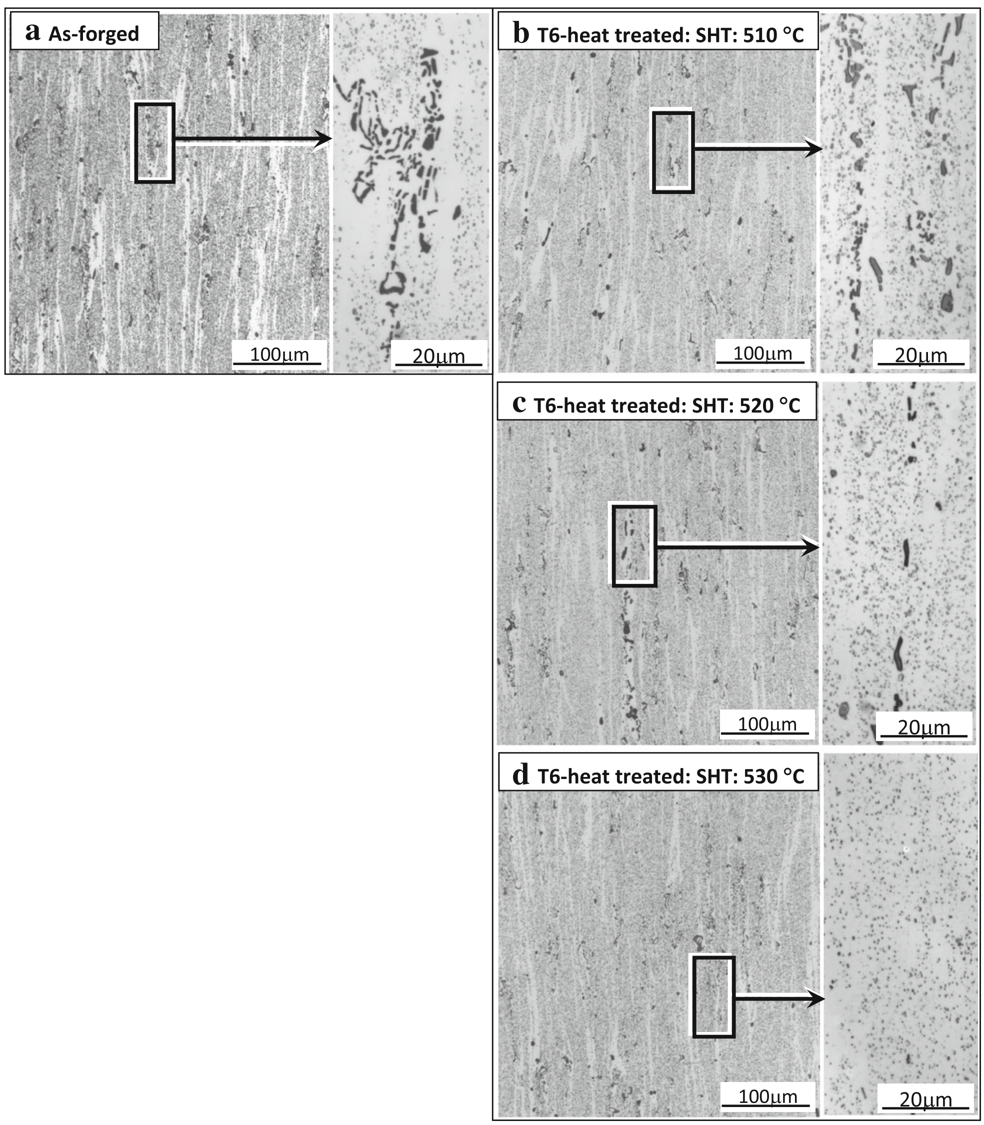


Fig. 7 Microstructure of the EN AW 6082 forging: **a** as-forged and after T6 heat treatment at different SHT temperatures: **b** 510 °C, **c** 520 °C, **d** 530 °C. Micrographs on the *right-hand side* are higher magnification views of the *marked regions* in micrographs on the *left-hand side*. They illustrate the gradual solutionizing of the coarse eutectic phases with increasing SHT temperature

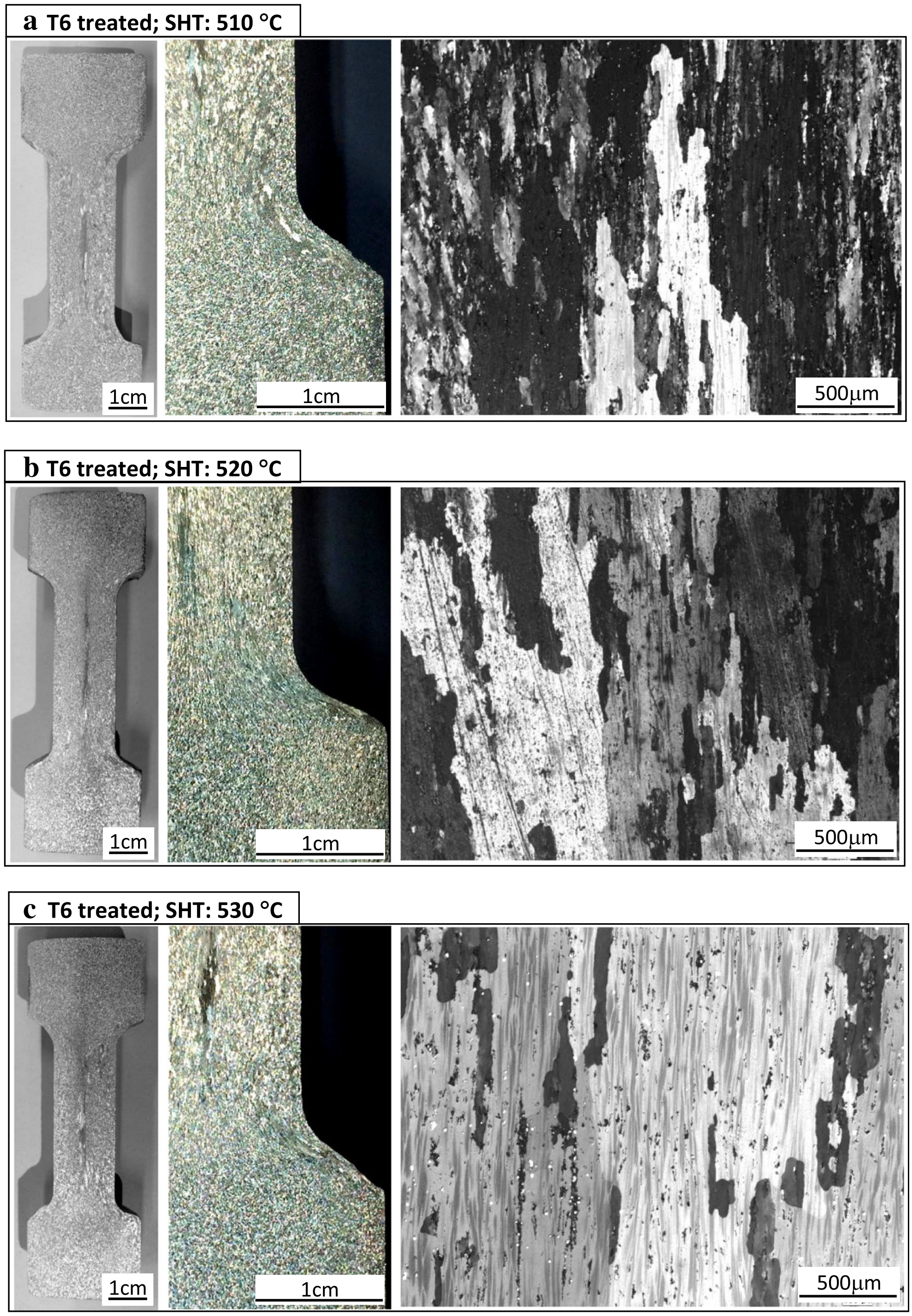


Fig. 8 Section (*left*), lower right corner of the section (*middle*) and the anodized section (*right*) of the EN AW 6082 forging showing the macrostructure and grain structure after the T6 heat treatment at different solution heat temperatures: **a** 510 °C, **b** 520 °C, **c** 530 °C

equiaxed. This is in marked contrast to the twin-roll-cast (TRC) counterparts where the grains were shown to be aligned at some angle to the casting direction owing to the substantial hot rolling and extrusion deformations encountered in the roll gap during TRC. It is thus fair to claim that the TBC blanks are relatively strain-free at the start of the forging sequence. A gradual increase in the average grain size from the surface to the centre of the section is evident. The average grain sizes were estimated to be 85 ± 6 and 113 ± 15 μm, on the surface and at the centre, respectively. This slight coarsening across the section is due to the cooling rate gradient in TBC where the surfaces in contact with the belt solidify faster than the centre.

Major structural changes are noted when the TBC blank is forged into the experimental part shown in Fig. 2d (Fig. 5). The majority of the eutectic cells are broken into relatively smaller compound particles. These particles and the very fine precipitates are aligned in the direction of plastic flow

Table 2 Tensile test results and hardness measurements after T6 heat treatment at different solution heat treatment temperatures

SHT (°C)	R_p (MPa)	R_m (MPa)	A_5 (%)	Hardness (HB)
510	312 ± 7	336 ± 5	10.3 ± 0,6	116 ± 3
520	289 ± 5	319 ± 8	9.8 ± 0. 8	110 ± 2
530	281 ± 12	312 ± 9	9.7 ± 1.3	104 ± 1

associated with hot forging. We infer from the distribution of the precipitate-free zones that decorate the grain boundaries that the alignment is stronger at the centre of the section. This has led to a section grain structure where the grains near the surfaces are pancake-shaped while the rest of the section is predominantly fibrous (Fig. 6). This is typical of forged components where the grains on the surface are not free to flow as the centre grains due to the sticking friction that acts along the contact zone between the die and the work piece. The hardness of the forging at the centre is 66 HB, confirming the deformation hardening of the forging during hot forging. Friction effects in microforging processes were investigated in detail by Ghassemali et al. recently [27–30].

Microstructures of the heat-treated components are largely similar to those of the forged components with a heavy population of precipitates aligned in the plastic flow direction (Fig. 7). However, coarse Fe-based intermetallic particles, evidently in large numbers at the solution heat treatment (SHT) of 510 °C, were reduced in number with increasing SHT temperature. The solutionizing of these compound particles is apparently encouraged with increasing temperatures, which, in turn, offers higher solute levels for precipitation during a subsequent ageing cycle. Increased number of very fine precipitates with increasing SHT is evidenced by the gradually increasing intragranular contrast (darkening).

Grain structures of the heat-treated forgings are shown in Fig. 8. There is no evidence of coarse surface grains, typical of solution-heat-treated EN AW 6082 forgings manufactured from extruded stock. This is in agreement with the results of a recent work where the absence of abnormally coarse surface grains has been shown to be an attribute of forging from cast stock [11]. Instead, there are coarse grains right at the centre of the section. This is believed to be due to the relatively coarser dendritic structure at the centre where the spacing of the interdendritic compound particles is relatively large. This, in turn, reduces the pinning effect on the moving boundaries leading to coarse grains at the centre plane during preheating the stock to the forging temperature as well as during hot forging. While it is not directly evident from the section macrographs (Fig. 8), the metallographic analysis of the anodized samples clearly shows that there is grain coarsening across the entire section of the heat-treated forgings (Fig. 8). The fibrous grains of the forged component are replaced by coarser grains elongated in the plastic flow direction. These new grains have apparently formed through recrystallization and subsequent growth of these fibrous grains during SHT since the strain energy inherited from the forging step suffices to kick off a recrystallization reaction. The forging deformation and the temperatures above 500 °C apparently encourage recrystallization and grain growth. This process occurs readily in the present alloy since the dispersoids in typical EN AW 6082 alloy stock for extrusion and forging help to impede the motion of grain boundaries, thus avoiding recrystallization and grain growth. However, these dispersoids are largely missing in the present alloy because of its low Mn and very low Cr contents (Table 1). Nevertheless, there is still some growth anisotropy provided by the Fe-based intermetallic

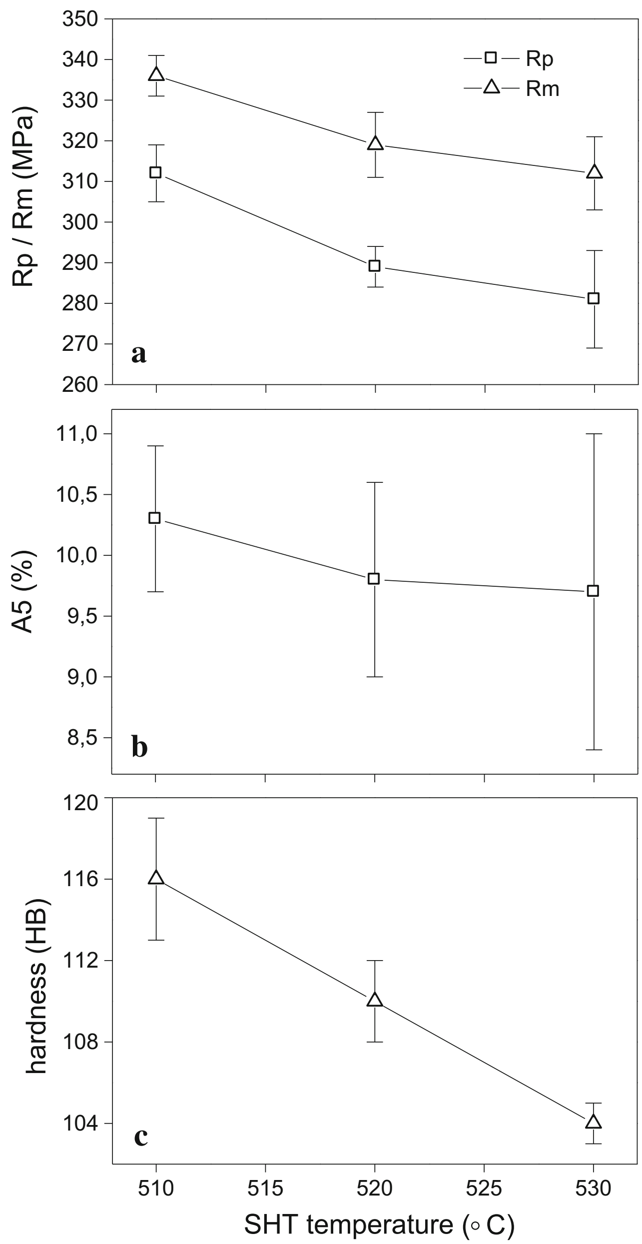


Fig. 9 Change in **a** yield (*Rp*) and tensile strength (*Rm*), **b** elongation and **c** T6 hardness with increasing SHT temperature

Table 3 Tensile test, hardness, Charpy impact energy and number of fatigue cycles to fracture results measured on the forgings produced from twin-belt-cast EN AW 6082 alloy blanks and on forgings manufactured from standart extruded forging stock

Forging stock Minimum required	R_p (MPa) >260	R_m (MPa) >310	A_5 (%) >8	Hardness (HB) >95	Impact energy (J)	Cycles to fracture (N_f)
TBC blank	312 ± 7	336 ± 5	10.3 ± 0.6	116 ± 3	14 ± 1	255,351 ± 87,541
Extruded stock	313	342	8.6	106	14	204,625 ± 128,756

particles and Mg-Si precipitates as inferred from the shape of the recrystallized grains. One can see the further growth of the selected recrystallized grains at the expense of smaller ones with increasing SHT temperature (Fig. 8c). Even an increase of 10 °C in the SHT temperature makes a marked impact.

The tensile properties, hardness measurements, fatigue properties and the impact energy values of the experimental part forged from the TBC blank are listed in Table 2. The yield and tensile strength, elongation and hardness values all decrease steadily with increasing SHT temperature (Fig. 9). The drop in tensile properties and hardness values with increasing SHT temperature is clearly linked with grain coarsening that is inevitable at higher SHT temperatures (Fig. 8). While standard T6 processing with a SHT between 510 and 520 °C readily meets the minimum values identified by the customer, solutionizing at 530 °C leads to tensile properties and hardness values slightly below these limits. Hence, the effect of SHT temperature needs to be considered when TBC blanks are processed conventionally and supplied in the T6 temper. Of the three SHT temperatures employed, 510 °C is regarded as the optimum.

It is fair to claim from Table 3 that both the static and dynamic mechanical properties of the TBC EN AW 6082 blank compare favourably with the extruded forging stock. In fact, the preliminary results regarding the number of fatigue cycles to fracture with the TBC stock appears to be superior. While further fatigue testing is underway, this may be attributed to the lack of coarse surface grains, in spite of some coarsening near the centre of the forgings. Fatigue properties are known to be very sensitive to the surface conditions through crack initiation, and the absence of coarse surface grains may be an advantage. It is concluded from the foregoing that the TBC EN AW blanks can be used for the manufacture of relatively flat, 2-dimensional automotive parts such as wishbone suspension components.

4 Conclusions

One of the cast forging stock options for the manufacture of lightweight suspension components is the TBC strip that can be readily stamped to cut suitable preforms/blanks and shaped into relatively flat, 2-dimensional forgings such as wishbone suspension.

The heat treatment of suspension components forged from TBC ENAW 6082 blanks is critical. The yield and tensile strength, elongation and hardness all decrease steadily with increasing SHT temperature due to grain coarsening during the high-temperature solutionizing step. Nevertheless, both the static and dynamic mechanical properties of the TBC EN AW 6082 blank compare favourably with the extruded forging stock, in spite of some grain coarsening near the centre of the forging.

It is thus concluded that the TBC EN AW blanks can be used for the manufacture of relatively flat, 2-dimensional automotive parts such as wishbone suspension. Industrial-scale production of wishbone forgings from TBC EN AW 6082 stock is underway at the leading manufacturer of suspension components in Turkey.

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