ANISOTROPY OF ROOM TEMPERATURE DUCTILITY IN HAYNES®282® FORGINGS

Ceena Joseph[1], Magnus Hörnqvist[2], Christer Persson[1]

[1]Chalmers University of Technology, Department of Materials and Manufacturing Technology, S-41296 Göteborg, Sweden
[2]Chalmers University of Technology, Department of Applied Physics, S-41296 Göteborg, Sweden

Keywords: Haynes ®282®, tensile ductility, microstructure

Abstract

Haynes®282® is relatively new nickel base superalloy, offering excellent mechanical properties and attracting interest from the gas turbine and aerospace industry. One of the criteria for meeting specification requirements is the room temperature ductility.

In this paper specimens from forgings were tested for room temperature ductility in the longitudinal transversal (LT) and short transversal (ST) directions. The results show large anisotropy in ductility, varying from approximately 25% in LT to 12% in the ST direction. Fractographic and metallographic analysis show the presence of MC and M_6C carbide banding. Failure in both ST and LT direction was ductile with intergranular fracture, but in the ST direction the failure was preferentially due to cracking of carbides. The orientation of carbide bands in preferred direction and their excessive segregation affecting ductility in ST is discussed in this work.

Introduction

Nickel base superalloys are widely used in manufacturing of gas turbine components which are required to have high temperature strength, fatigue and creep resistance for high temperature applications [1]. One of the recent additions to this class of alloys is Haynes®282® , which offers excellent mechanical properties like thermal stability [2]. It is presently attracting interest from the industry for application in aero engines and industrial gas turbines where the versatility in terms of high temperature properties and fabricability has made this alloy superior to other competent alloys like 718 and Waspaloy.

In many applications it is desirable to use Haynes®282® in forged form. In order to be an attractive candidate material, the associated manufacturing process needs to show a stable result, which guarantees conformance to the material specification. This is usually tested in terms of room temperature strength and ductility, as well as high temperature creep strength and ductility. But very little information is available on room temperature tensile deformation and fracture behaviour of Haynes®282® forgings. In this study, the room temperature tensile deformation behaviour of Haynes®282® forging specimens in the longitudinal transversal (LT) and short transversal (ST) has been investigated along with its damage and fracture mechanisms in relation to microstructural features. It is observed that commercial forgings can show large anisotropy with respect to tensile ductility ranging from 12 % in ST to 25 % in LT direction.

Experimental procedure

The composition of γ' strengthened superalloy Haynes®282® used in this investigation is shown in Table I. The alloy was melted in a vacuum induction furnace and remelted in an electroslag remelt furnace. Billet with a nominal diameter of 152 mm were forged perpendicular to the billet axis down to a varying thickness (see Table II). The forgings were solutions and precipitation heat treated with similar parameters as for manufacturing of plate according to AMS4951.

Screw-head tensile specimens with 25mm in gauge length and 6.4 mm in diameter were machined from the forgings with tensile axes parallel to (ST) or perpendicular (LT) to the pressing direction. The specimens discussed in this paper along with their room temperature tensile properties are shown in Table II. Uni-axial tensile tests were carried out in a universal testing machine at room temperature with strain rate of 0.005/min to determine the yield stress, and then 0.05/minute until failure. Specimens for analysis of microstructure and deformation were carried out using optical microscope and scanning electron microscope (SEM). Typically specimens were sectioned near the fracture of the tensile specimens, mechanically polished and etched electrolytically with a solution consisting of oxalic acid at 1.8V.

Table I Composition (wt %) of Haynes®282® alloy

Specimen ID	**Ni**	**Cr**	**Co**	**Mo**	**Ti**	**Al**	**Fe**	**Mn**	**Si**	**C**	**B**
A and B	Bal	19.44	10.22	9.42	2.15	1.44	0.92	0.06	0.07	0.067	0.004
C	Bal	19.48	10.30	8.50	2.19	1.56	0.42	0.03	<0.05	0.062	0.004

Table II. Mechanical properties of discussed specimens.

Specimen ID	**Specimen Direction**	**Forging thickness (mm)**	**Elongation 4D(%)**	**UTS(MPa)**	**YS(MPa) (0.2 % offset)**
A	LT	40	23	1170	720
B	ST	100	16	1080	685
C	ST	60	12	1060	765

Results and Discussion

Microstructure of heat treated Haynes®282® superalloy

A typical optical micrograph of the microstructure of Haynes®282® in the forged and heat treated condition is shown in Figure. 1. Primary MC (Ti and Mo rich) carbides are distributed within the matrix which has an average grain size of 200 µm. Secondary carbides M_6C which are preferentially Mo rich and $M_{23}C_6$ (Cr-rich) carbides are dispersed inside the grain and at the grain boundaries. An SEM image of a grain boundary with presence of both primary and secondary carbides, along with γ' is shown in Figure. 2. As can be seen in this image the area around the grainboundary seems to be devoid of large γ' precipitates. The grain boundaries usually contain globular $M_{23}C_6$ carbides along with other secondary carbide precipitates.

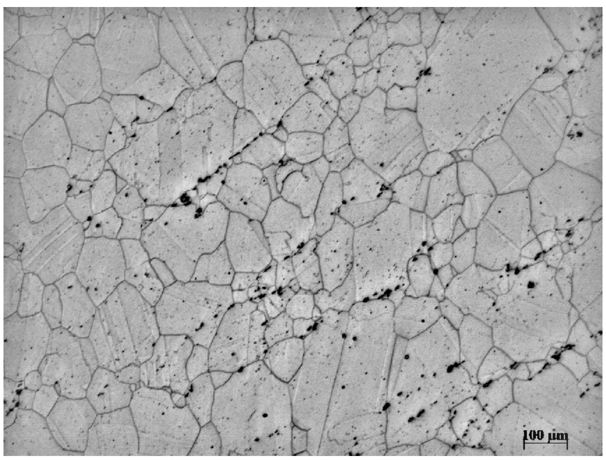

Figure 1. Optical image of forged Haynes®282®

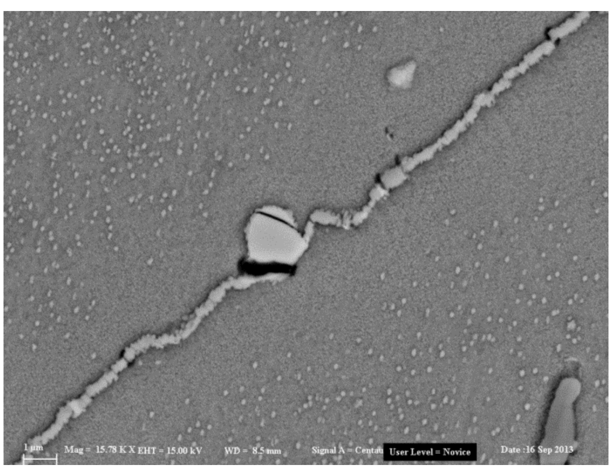

Figure 2. SEM image of Haynes®282®

The optical microstructures of the tested specimens A, B and C in the LT and ST directions are as shown in Figure 3a,b and c. The microstructures are taken from region just below the fracture surface and in the direction of tensile axis. As can be seen in these images the orientation of MC and M_6C carbides with respect to the tensile axis is different in the LT and ST direction. In specimen A with elongation 23% along LT direction, the carbides are oriented along the direction of the tensile axis. In specimen C with elongation 12% in the ST direction the carbides are oriented perpendicular to the tensile axis. Other tested specimens in LT direction indicated that the carbide orientation varied between 0-45° to the tensile axis, while that in the ST direction was preferentially 90° to the tensile axis. One of the exceptions in ST is shown in Figure 3b wherein the carbides are oriented along the direction of tensile axis and also observed to be heavily distributed within the matrix. The region segregated with carbides show presence of smaller grains, as these MC and M_6C carbides inhibit the grain growth. This leads to a more pronounced banded structure in case of short transverse specimens.

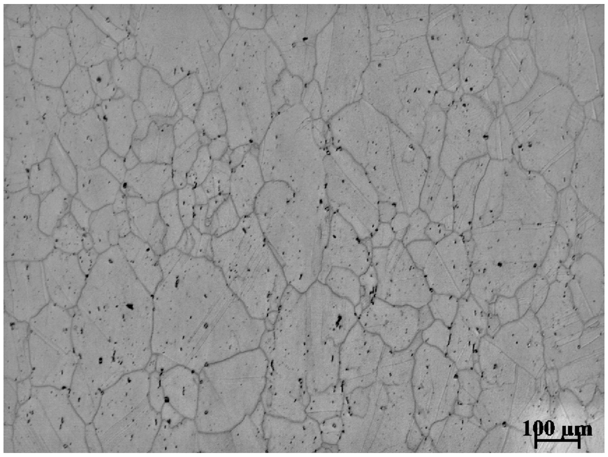

Figure 3a. Specimen A in LT direction with 23% elongation

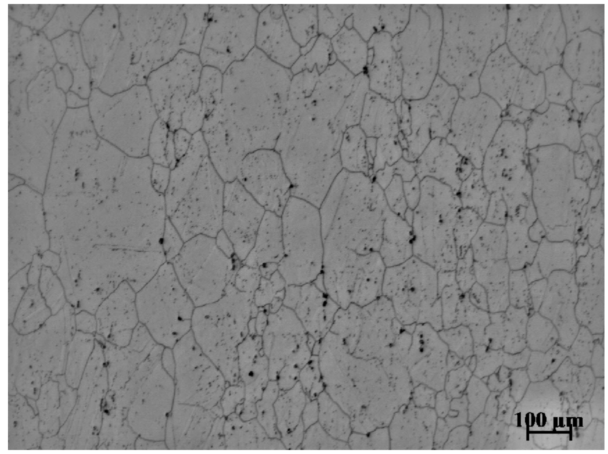

Figure 3b. Specimen B in ST direction with 16% elongation

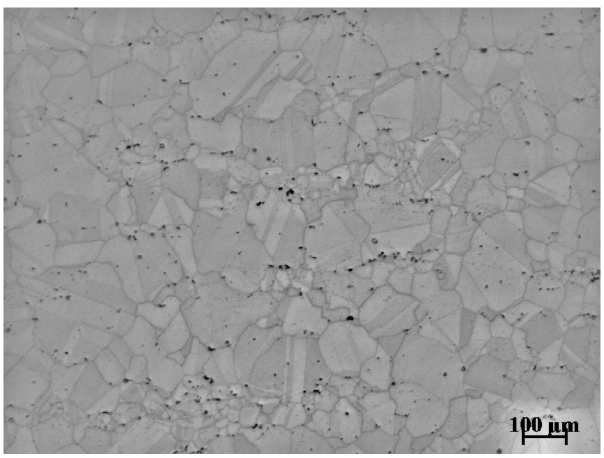

Figure 3c. Specimen C in ST direction with 12% elongation

$M_{23}C_6$ carbides are precipitated at the grain boundaries as well as in small globular forms near the MC carbides. Extensive work has been carried out to determine the role of size distribution and form of $M_{23}C_6$ carbides in superalloys. It has been suggested that formation of secondary carbides is also dependent on the size, form and dispersion MC carbides [3-5]. In ST specimen presence of excess and segregated form of MC carbides can lead to formation of lamellar $M_{23}C_6$ carbides along with brittle M_6C carbides thereby decreasing the strength at the grain boundaries. This might affect the strength properties in short transvers direction. However in LT direction specimen, the finely dispersed MC carbides form serated $M_{23}C_6$ carbides which might contribute to the strength in grainboundaries and hence better tensile properties at room temperature. This indicates that the size and distribution of MC carbides in the forging stock plays a very important role to its strength properties of the subsequent forgings.

Fracture surface morphology

The morphology of the fracture surface of tensile specimen A-(LT, 23% ductility) is shown in Figure. 5a. It shows an intergranular failure along with presence of dimpled granular structure. Figure. 5b shows the occurrence of micro-void coalescence leading to intergranular failure. The MC carbides also act as crack initiation sites as shown in Figure. 5c. The presence of such MC carbide particles near the grain boundaries is detrimental to the strength of the alloy. However in this case, both MC and M_6C carbides are not preferentially located at the grain boundaries and the predominant failure is micro-void coalescence giving higher elongation in LT direction, usually between 23-31% even at room temperature. It is also possible that the presence of micro-void coalescence is related to the occurrence of a region devoid of large γ' around the grain boundaries, creating a locally softer region.

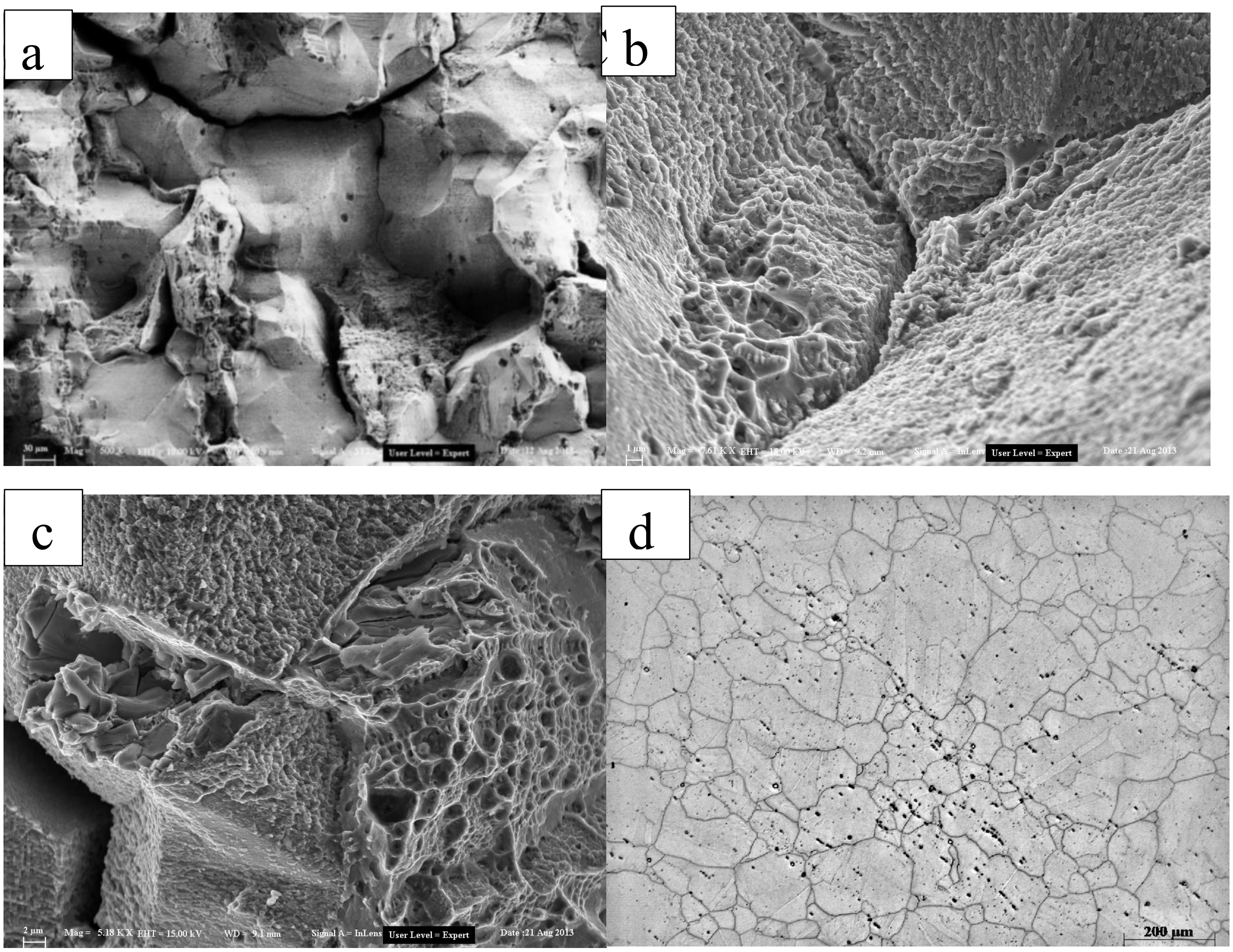

Figure 5. (a) SEM image showing intergranular failure in specimen A (LT, 23 % ductility) (b) SEM image indicating micro-void coalescence in the intergranular failure regions (c) SEM image indicating cracking of MC carbides (d) Optical image of cross section of specimen A

In specimen B (ST, 16% ductility), the fracture morphology shows both intergranular failure and heavy segregation of MC and M_6C carbides giving rise to banded structure as seen in Figure.6a and d . Figure 6b and 6c show presence of segregated MC and M_6C carbides respectively. Cracking in these segregated MC and M_6C carbides gives rise to cracks instantaneously between matrix and carbide particles causing fracture with cleavage characteristics. However, regions showing intergranular failure show presence of microvoids as in the LT specimen. Hence the elongation in ST direction is reduced compared to the LT direction due to the distribution and cracking of carbides.

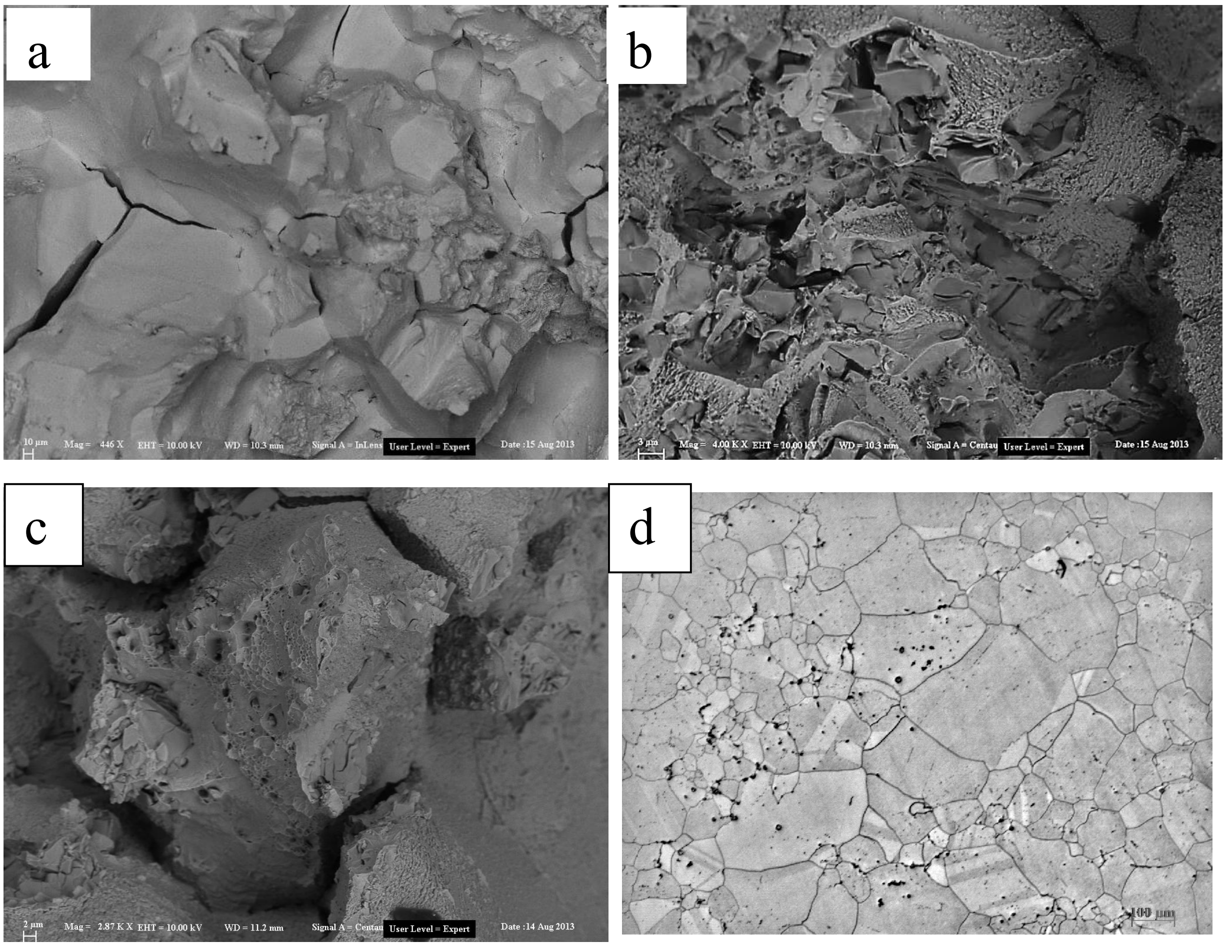

Figure 6. (a) SEM image showing intergranular failure and heavy segregated region of carbides in Specimen B(ST, 16 % ductility) (b) SEM image showing cracks segregated MC carbide (c) SEM image indicating segregated M_6C carbides (d)Optical image of cross section of specimen B

In specimen C (ST, 12% ductility), the fracture morphology shows both intergranular failure and heavy segregation of MC and M_6C carbides as seen in Figure 7a. Figure 7b and c show presence of segregated MC and M_6C carbides respectively, while the Figure 7d show banded structure. Cracking in these segregated MC and M_6C carbides gives rise to cracks instantaneously between matrix and carbide particles, giving rise to facture of cleavage characteristics. However, regions showing intergranular failure show presence of microvoids. Hence the elongation in ST direction is relatively lower than that of LT direction.

The optical image in Figure 5d, 6d and 7d also indicates a grain size variation from around 10µm to 400µm. Such a large variation in grain size can also impact on ductility of the material. The ductility in ST specimens can also be improved by having smaller and uniform size distribution.

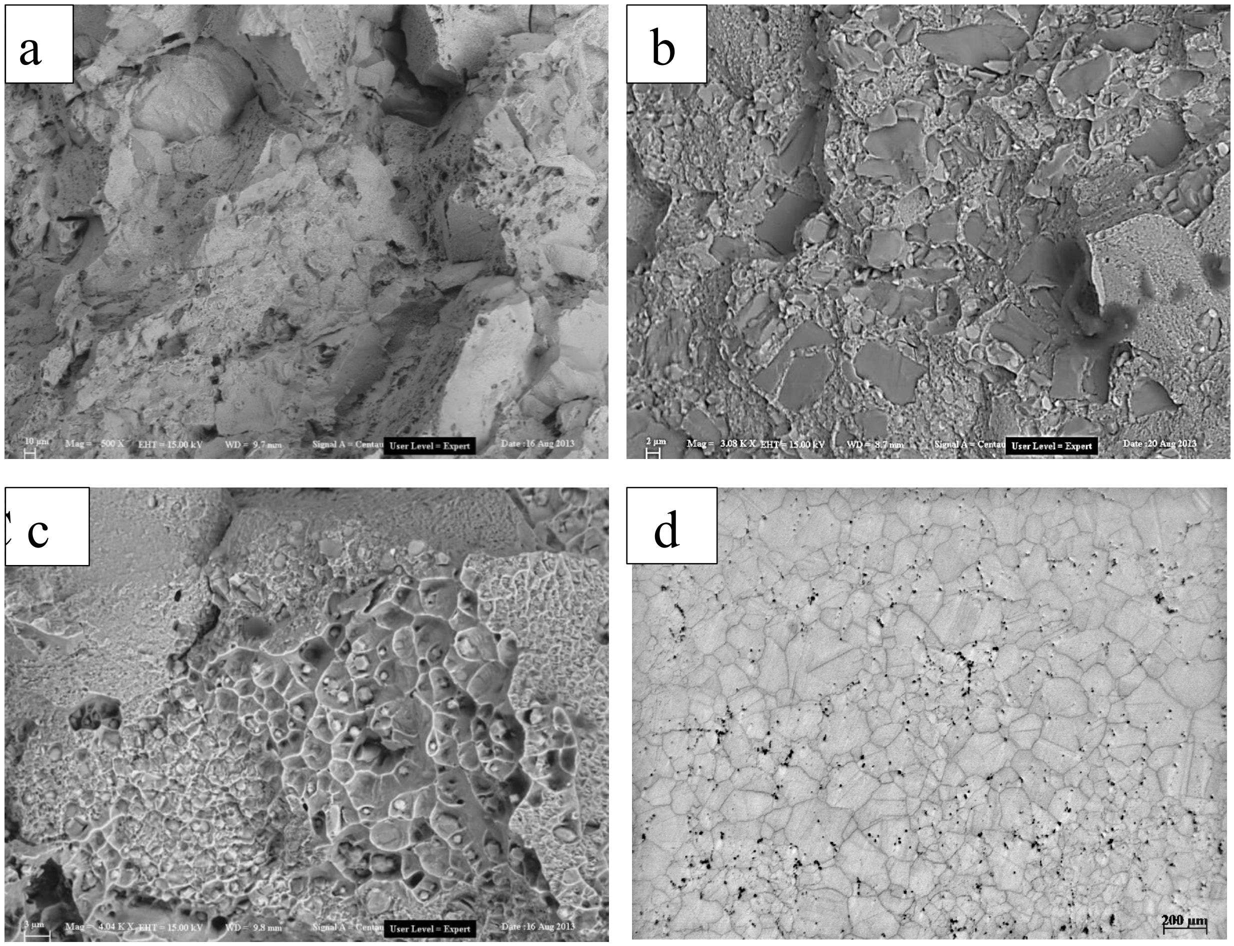

Figure 7. (a) SEM image of the fracture surface showing segregated carbides in specimen C (ST, 12 % ductility) (b) SEM image showing cracks in segregated MC carbide (c) SEM image indicating cracked M_6C carbides within dimpled structure. (d) Optical image of cross-section below the fracture surface in specimen C.

Tensile Properties

Room temperature tensile test results are summarized in Table II and Figure 8. The main observations on the trends of tensile properties are as follows:

- Both ductility and ultimate tensile strength (UTS) are higher in the LT direction, and there is a co-variation between these two values (see Figure 8). Presumably, the loss of both ductility and UTS are caused by the earlier on-set of cracking due to the distribution of carbides in the ST specimens.
- The orientation dependence of the YS is less systematic. Specimen C (ST, 12 % ductility) shows the highest YS at 765 MPa, followed by specimen A at 720 MPa (LT, 23 %) and specimen B (ST, 16 %) at 685 MPa. This may be a result of differences in the carbide distribution between the specimens, although carbides are not expected to have a major strengthening effect.

There have been extensive studies of carbides in superalloys, indicating their beneficial or detrimental effects on the mechanical properties depending on its type, size, distribution and morphology [6,7]. M_6C carbides and some of the MC carbide particles can be detrimental to room temperature tensile properties of superalloys if they are excessively segregated in small regiona. Hence, the distribution of carbides plays an important role for the tensile properties in forgings, as the carbides fracture and act as crack initiation sites. Figure 9 shows a typical example of the severe cracking in MC and M_6C carbides.

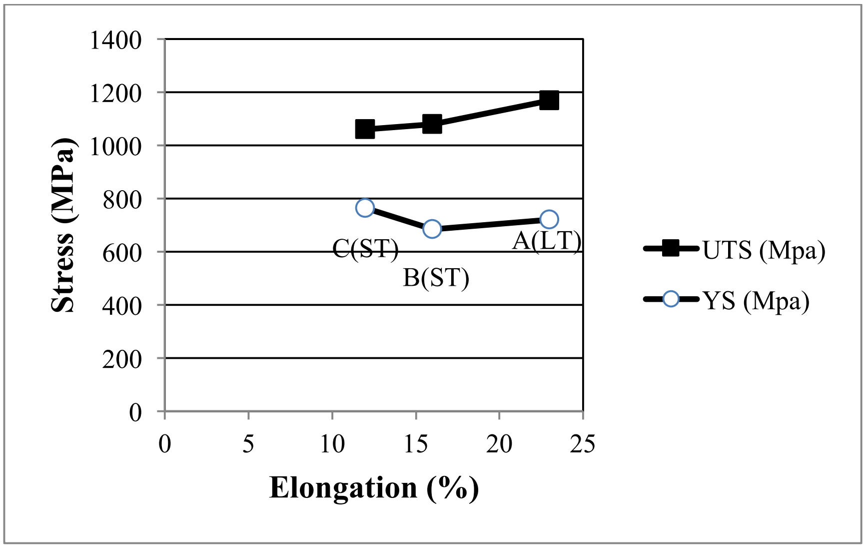

Figure 8.Variation of YS (open symbols) and UTS (Closed symbols) for the different speciemsn.

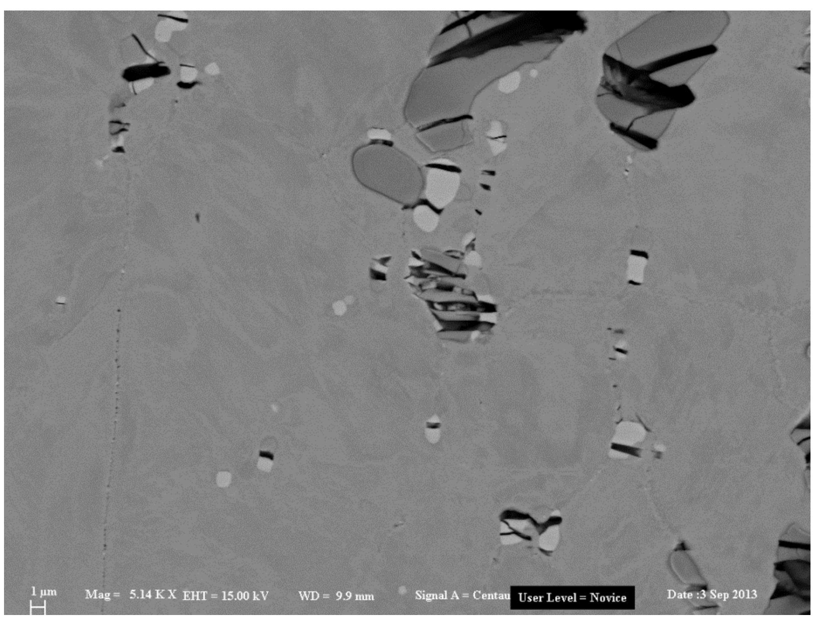

Figure 9. Extensive cracking of MC and M_6C carbides perpendicular to the loading direction

Conclusions

Tensile and fracture behaviour of Haynes®282® forgings in ST and LT directions has been investigated in this work and it can be concluded that

- Fracture is due to micro-void coalescence at grain boundaries in LT direction. However in ST direction it is due to both the intergranular micro-void coalescence as well as cracking in segregated MC and M_6C carbides.
- Banding of carbides with respect to its orientation 0-90° to tensile axis plays a very important role in determining the room temperature strength and ductility. In ST specimens, an elongation of 12% was due to heavy segregation of carbides 90° to tensile axis and 16% due to carbide segregation 0 – 45° to tensile axis.
- Excessive segregation of MC carbides from the forging stock should be avoided to improve its room temperature ductility.

References

[1] Roger C Reed, *The Superalloys Fundamentals and Applications*, Cambridge University Press, (2008), 1-25.

[2] L.M.Pike, "Development of a fabricable gamma-prime strengthened superalloy", Superalloys 2008

[3] Wei. C.-N, et al. "The effects of carbon content on the microstructure and elevated temperature tensile strength of a nickel-base superalloy." *Materials Science and Engineering: A* 527(16-17. (2010)):3741-3747.

[4] Bai. G , et al. "Effect of temperature on tensile behavior of Ni–Cr–W based superalloy." *Materials Science and Engineering: A* 528 (4-5) (2011) 1974-1978.

[5] J. Larson, *Metall. Mater.Trans* A7 (1976) 1497-1502

[6] Yang. J , et al. "Relative stability of carbides and their effects on the properties of K465 superalloy." *Materials Science and Engineering: A* 429(1-2) (2006). 341-347.

[7] J.Chen, et al. "MC carbide formation in directionally solidified MAR-M247 LC superalloy, Mater.Sci.Eng A247(1998)113-125