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Deformation and damage mechanisms in IN792 during thermomechanical fatigue

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Abstract

The deformation and damage mechanisms arising during thermomechanical fatigue (TMF) of the polycrystalline superalloy IN792 have been investigated. The TMF cycles used in this study are in-phase (IP) and out-of-phase (OP). The minimum temperature used in all TMF-tests is 100°C while the maximum temperature is 500° or 750°C in the IP TMF-tests and 750°, 850° or 950°C in the OP TMF-tests. The majority of the cracks are transcrystalline, except for the IP TMF-test at 750°C, where some tendency to intercrystalline crack growth can be seen. In all tests, the cracks were initiated and propagated in locations where deformation structures such as deformation bands have formed in the material. In the temperature interval 750°-850°C, twins were formed in both IP and OP TMF-tests and this behaviour is observed to be further enhanced close to a crack. Twins are to a significantly lesser extent observed for tests with a lower (500°C) and a higher (950°C) maximum temperature. Recrystallization at grain boundaries, around particles and within the deformation structures have occurred in the OP TMF-tests with a maximum temperature of 850° and 950°C and this is more apparent for the higher temperature. Void formation is frequently observed in the recrystallized areas even for the case of compressive stresses at high temperature.

Keywords: *nickel based superalloys, fatigue, shear bands, recrystallization, twinning*

1 Introduction

The nickel-base superalloy IN792 is a cast polycrystalline material which is widely used in industrial and aircraft turbines due to its high strength and excellent hot corrosion resistance [1-2]. Like most nickel-based superalloys it is intended to withstand extreme conditions of temperature and loading during many years of operation before the scheduled removal of components during engine refurbishment. IN792 belongs to an important class of material and can typically be found in critical components such as turbine blades and vanes. Even though such components sometimes are made in single crystalline form, polycrystalline materials are very common due to lower costs and easier production route [3].

The critical turbine parts are subjected to very significant and time dependent stress fields during service and the origin of the stresses is the combination of external forces and the temperature gradients which are generated during engine start-up and shut down. The accumulation of such stress and temperature cycles may eventually lead to crack initiation, a phenomenon often referred to as thermomechanical fatigue (TMF). Although material for critical hot section components have traditionally been evaluated with respect to creep, the deformation characteristics under TMF conditions are now just as important and need to be understood and quantified if component lifetime estimates are to be accurate [4-5]. Broadly speaking, TMF failure is promoted when plastic strains cannot be accommodated at low temperatures and creep deformation in combination with oxidation occurs at high temperature. Other effects like nucleation of recrystallisation caused by cyclic plastic deformation are also contributing to damage during TMF [5]. Even though the temperature and the stress/strain cycle can be completely independent of each other it is common to consider two main types of

idealized TMF cycles; (IP) in-phase TMF in which the material undergoes creep relaxation in tension at high temperature and plastic deformation in compression at low temperature, (OP) out-of-phase TMF in which the material undergoes creep relaxation in compression at high temperature and plastic deformation in tension at low temperature.

In recent studies it has been shown that the deformation and damage mechanisms in single crystal superalloys during TMF are very different from those traditionally reported to occur under creep or isothermal fatigue [5-8]. In single crystalline superalloys, highly localized twinning [5] or shearing [6] are the main deformation mechanisms operating during OP TMF cycling and recrystallization within the deformation bands has been observed if the maximum temperature is high ($>900^{\circ}\text{C}$) enough [5-6]. Unfortunately, there is still a significant lack of fundamental information regarding, for instance, the temperature dependence, the orientation dependence and the influence of cycle type on the deformation and damage mechanisms during TMF in cast nickel-based superalloys.

Regarding fatigue at high temperature of cast polycrystalline superalloys typically used for blades and vanes, it has been shown that crack initiation often depends on the oxidation of grain boundaries at the specimen surface [9]. Furthermore, it has been observed that intergranular damage caused by IP TMF loading reduces the lifetime below the respective values measured during OP TMF loading where no intergranular damage was detected [10]. Thus, due to the presence of grain boundaries, a slightly different deformation and damage behaviour can be expected for polycrystalline nickel-based superalloys compared to a single-crystalline ditto.

The research reported in the present paper focuses on the deformation and damage mechanisms in cast polycrystalline nickel-base superalloys during TMF and the influence of temperature range and cycle type have been investigated through a large number of carefully conducted TMF experiments followed by comprehensive microstructure characterizations using scanning electron microscopy (SEM). Compared to single crystalline superalloys some fundamental differences as well as similarities in the deformation and damage behaviour have been discovered, which provide new insights into the degradation process in these types of alloys under industrially relevant operating conditions.

2 Experimental procedure

All tests in this study were done on IN792, a γ' -precipitation hardened nickel-based superalloy. The chemical composition is Ni–12.4Cr–8.9Co–1.8Mo–4.0W–3.5Al–4.0Ti–4.1Ta–0.08C–0.017B–0.019Zr (wt-%). After conventional casting the material was hot isostatically pressed (HIP) at 1195°C and 150 MPa for 2 h followed by solution heat treatment at 1121°C for 2 h and ageing at 850°C for 24 h. Rotational symmetric test bars with a parallel length of 24 mm and a diameter of 6.0 mm were machined from the cast bars which had an initial diameter of 20 mm.

TMF tests can be performed with an arbitrary phase shift (φ) between the temperature and the mechanical loading. In this study, in-phase ($\varphi=0^\circ$) and out-of-phase ($\varphi=180^\circ$) tests have been considered. All tests were conducted using an MTS 810 servo-hydraulic thermomechanical fatigue machine where induction heating and forced air cooling are used in order to cycle the temperature. The minimum temperature in all tests was 100°C

while the maximum temperature was either 500°C, 750°C, 850°C or 950°C. Strain was measured by an axial extensometer and all tests were done in mechanical strain control (i.e. with a fixed total strain range compensated for thermal expansion $\Delta\epsilon = \epsilon_{\max} - \epsilon_{\min}$). The strain ratios ($R = \epsilon_{\min} / \epsilon_{\max}$) were always $R = 0$ for the IP tests and $R = -\infty$ for the OP tests. Even if significant plastic deformation occurs in the first cycle, sufficient creep relaxation during the hold time at the maximum temperature is typically necessary to establish a stable stress-strain cycle, without a drift in mean stress from one cycle to another cycle. Thus, in order to achieve a stabilized mean stress early in the tests, a 20 hour hold time was applied at the maximum temperature ($T_{\max}$) during the first cycle. For all subsequent cycles a 5 minute hold time was applied. This combination of R-ratio and longer hold time in the first cycle was chosen since it better represents the real situation for most engineering components under OP-TMF or IP-TMF loading, see reference [5] for further details.

After testing the ruptured fatigue specimens were sectioned parallel to the longitudinal axis for microstructural investigations. For comparison, also virgin (untested) material has been investigated. All samples were prepared by grinding and mechanical polishing and analysed using scanning electron microscopy. In order to obtain optimal channelling contrast in the image, an annular backscatter electron detector on a Hitachi SU70 FEGSEM operating at 10 kV was used. The contrast in such an image is associated with discontinuities in the specimen and any crystallographic defect that produces a distortion in the lattice, such as a twin, a sub-grain or dislocation can be observed. Such images are similar in appearance to transmission electron micrographs, but with lower image resolution [11].

Orientation imaging microscopy (OIM) was performed using an electron back-scattering diffraction (EBSD) system from HKL Technology. By using EBSD analysis the actual presence of twins can be confirmed since they have a misorientation of 60° compared to the surrounding material. In case recrystallization has occurred new grains are formed with an arbitrary misorientation. In the OIM figures, grain orientations are represented by different colours and the grain boundaries are coloured black or grey depending on their misorientation. In order to obtain satisfying results from the EBSD analyses noise reduction has been carried out.

3 Experimental results

3.1 *The TMF behaviour*

Results from the TMF testing can be seen in fig. 1-2. As expected, there is a clear tendency of decreasing TMF lifetime with increasing temperature. This is illustrated in fig. 1a where the mechanical strain range as a function of cycles to failure (N_f) is plotted. One can also observe that for tests with a maximum temperature of 750°C , the OP cycle seems to be more damaging than the IP cycle. If the inelastic strain range in the midlife cycle, measured as the width of the stress-strain loop at zero stress, is plotted as a function of cycles to failure, no significant temperature dependency can be seen, see fig. 1b. However, for similar number of cycles to failure there is a tendency for higher inelastic strain ranges in the IP tests compared to the OP tests.

Fig. 2 shows the maximum and minimum stress in the midlife cycle ($N_f/2$) as a function of the total number of cycles to failure. It can be seen that both the maximum and the minimum stress decreases with increasing number of cycles to failure and that tensile mean stress develops during OP TMF cycling while compressive stress develops during

IP cycling. The highest influence of temperature can be seen on the maximum stress for the IP tests and on the minimum stress for the OP tests. In both cases these stresses correspond to the stress at the maximum temperature in the TMF cycle.

Figure 1: TMF test results; (a) Mechanical strain range versus cycles to failure, (b) Inelastic strain range versus cycles to failure.

Figure 2: Maximum and minimum stress for the midlife cycle in; (a) IP TMF tests, (b) OP TMF tests.

3.2 Scanning Electron Microscopy

3.2.1 Virgin microstructure

IN792 is a conventional cast polycrystalline material and the virgin microstructure is shown in fig. 3-4. Typically the grain size is rather large and a lot of carbides are

formed at the interdendritic areas, see fig. 3. Carbides together with borides can also be observed in the grain boundaries as seen in fig. 4a. The primary γ' phase has a size about 0.6-0.7 μm and can have both an irregular and a cubic appearance as indicated fig. 4a. Much smaller secondary γ' particles are also present in the microstructure (fig. 4b), along with larger γ'/γ eutectic domains (fig 4a).

Figure 3: Backscatter electron micrograph showing typical grain structure of virgin un-tested material.

Figure 4: Typical microstructure of IN792 in virgin condition

3.2.2 IP TMF 100-500°C

Fig. 5 shows material which has been exposed to an IP-TMF test where the temperature is cycled between 100-500°C. For this test condition, the main crack propagates through the grains in a transcrystalline manner. The cracks propagate in areas, where the material is highly plastically deformed. In these zones, different types of deformation structures can be observed and in higher magnification, different types of dislocation structures, persistent slip bands and some narrow twins appear. Normally the crack propagates in a zigzag manner through the grains and often it has the same crystallographic direction as the persistent slip bands.

Figure 5: Crack appearance for an IP 100-500°C TMF tests. (a) Deformation structures along the crack, (b) Persistent slip bands. Stress axis in horizontal direction.

3.2.3 IP TMF 100-750°C

Fig. 6-9 show material which has been exposed to an IP-TMF test where the temperature is cycled between 100-750°C. For this test condition, the crack propagation is both transcrystalline and intercrystalline, see fig. 6-7. The cracks propagate in areas where the material is plastically deformed and the larger cracks often branch into different directions, see left part of fig. 6. While multiple cracking is frequently

observed, many small surface cracks seem to stop propagating after a while as seen in fig. 7b.

Figure 6: Crack appearance for an IP 100-750°C TMF tests. Twins and persistent slip bands are seen near the crack. Stress axis in horizontal direction.

Figure 7: IP 100-750°C TMF tests showing (a) intercrystalline crack propagating and deformation bands and (b) small surface cracks

In the plastically deformed zones, deformation bands are seen at lower magnification while at higher magnification, dislocation structures, twins and persistent slip bands can be observed (see fig. 6-8). Twins and persistent slip bands are often present near the cracks as indicated in fig 6. Typically, many twins also appear in the deformation bands

and slip-twin interaction frequently takes place, as illustrated in fig. 8. In fig. 9a, a twin and a deformation band can be observed and a slip-twin interaction has occurred between them (i.e. the twinning direction is changed over the deformation band). In fig. 9b, an orientation image map is shown and the misorientation profile along the y-axis in fig 9c confirms twinning since there is a misorientation of 60° between the twin and the surrounding material inside the deformation band. Generally the amount of twinning in the IP tests is higher at 750°C compared to 500°C .

Figure 8: Backscatter electron micrograph showing slip-twin interaction for an IP 100-750°C TMF tests. Stress axis in horizontal direction.

Figure 9: Interaction between a twin and a deformation band during an IP TMF 100-750°C test. (a) Backscatter electron micrograph showing twins and a deformation band. (b) EBSD-map of the area indicated in figure a. (c) Misorientation profile along the Y-axis in figure b.

3.2.4 OP TMF 100-750°C

Material exposed to an OP-TMF test in the 100-750°C temperature range is shown in fig. 10-11. The cracks propagate in a transcrystalline manner in zones where the material is plastically deformed and in these zones different deformation structures are observed. At higher magnification it is obvious that both dislocation structures and twins have been formed; see fig. 10-11. The cracks typically show two different growth behaviours. One is when the crack more or less propagates in the same crystallographic directions as the twins, see fig. 10. The other behaviour is when the crack is not following any specific crystallographic direction and instead follows the deformation structure, even if there are twins present near the crack, see fig. 11.

Figure 10: Transcrystalline crack propagation along the twinning directions during an OP TMF 100-750°C test.

3.2.5 OP TMF 100-850°C

Fig 12-13 show material which has been exposed to an OP-TMF test where the temperature is cycled between 100-850°C. As for the previous cycle types described above, the cracks propagate in areas where the material is most plastically deformed and

deformation structures can be observed at low magnification while dislocation structures and twins appear at higher magnification as seen in fig. 12-13. Crack propagation is completely transcrystalline for this test condition and cracks often branch into different directions, see the middle part of fig. 12.

Figure 11: Transcrystalline crack propagation along the deformation structures during an OP TMF 100-750°C test. Different types of dislocation structures and twins are present. Stress axis in horizontal direction.

Figure 12: Backscatter electron micrograph showing transcrystalline crack propagation for an OP TMF 100-850°C test. (middle) crack branching and deformation structures. (left) dislocation structures and twins in front of the crack tip. (right) twins along the crack. Stress axis in horizontal direction.

The cracks typically show two different growth behaviours. One is when the major part of the crack follows a specific crystallographic direction as illustrated in the left and right part of fig. 12, where there are many twins in front of the crack tip and along the crack sides. The other growth behaviour is when the crack does not follow any specific crystallographic direction but instead propagates along the deformation structure, even if there are many twins near the crack, see fig. 13. An EBSD analysis has been undertaken for the twins seen in fig. 13 and the result is shown in the orientation image map in fig. 14, which confirms that there is a misorientation of 60° between the twins and the surrounding material.

Figure 13: Transcrystalline crack propagation and deformation structures from an OP TMF 100-850°C test.

At grain boundaries and around particles, small grains are often observed, which indicate that recrystallization has occurred, see fig. 15a. This is also confirmed by an EBSD analysis of the same area and the corresponding orientation image map can be seen in fig. 15b and 15c. The size of the new grains is typically in the 1-4µm range.

Figure 14: EBSD-map and misorientation profile transverse to the twins in fig. 14

Figure 15: Recrystallization at grain boundaries during OP TMF 100-850 °C testing, (a) Backscatter electron micrograph, (b) EBSD-map, (c) misorientation profile along the Y-axis.

3.2.6 OP TMF 100-950°C

Material exposed to an OP-TMF test where the temperature is cycled between 100-950°C is shown in fig. 16-18. As for the tests conducted at lower temperatures the cracks have initiated and propagated in those areas which have the highest degree of plastic deformation and in these zones different deformation structures are observed. Recrystallization and void formation are frequently observed, especially in areas near the final fracture, as seen in fig. 16. Recrystallization can also be observed at grain boundaries and around particles as seen in fig. 17a and 18a and an EBSD analysis has been undertaken for the respective positions. The results are shown in the orientation image maps in fig. 17b and 18b respectively, and both of them confirm that new high angle grain boundaries have been formed. The size of the new grains is typically in the 2-10 μ m range which is significantly larger compared to the tests conducted at 850°C.

Figure 16: Typical microstructure near the fracture surface for an OP TMF 100-950°C test showing; (a) deformation structure and recrystallization, (b) deformation bands. Stress axis in horizontal direction.

In all OP TMF tests, independent of the maximum temperature applied, multiple transgranular surface cracks can be observed, see fig. 19. However, the majority of the surface cracks seem to have stopped propagating after a while. The appearance of the

cracks in fig. 19 also clearly indicates that the effect of oxidation becomes more important as the maximum temperature increases.

Figure 17: Recrystallization at grain boundaries during OP TMF 100-950°C testing, (a) Backscatter electron micrograph, (b) EBSD-map, (c) misorientation profile along the Y-axis

Figure 18: Recrystallization around a particle during OP TMF 100-950°C testing, (a) Backscatter electron micrograph, (b) EBSD-map, (c) misorientation profile along the Y-axis

Figure 19: Multiple transgranular surface cracks present after OP TMF testing with different maximum temperatures; (a) OP 100-750°C, (b) OP 100-850°C, (c) OP 100-950°C

4 Discussion

Thermomechanical fatigue of nickel-based superalloys is a complex process since there are many factors that influence the TMF behaviour. This topic has unfortunately not been studied extensively in the literature so far but our results provide new insights into the degradation processes occurring.

4.1 On the factors controlling the TMF lifetime of IN792

It is well known that TMF lives are significantly influenced by the maximum temperature applied in the TMF cycle since the effect of creep and oxidation increases with increasing temperature; this is clearly illustrated in fig 1a. However, as seen in fig. 1b, there is a rather good correlation between inelastic strain range and cycles to failure independent of cycle type and maximum temperature in the TMF test. This indicates that the most important factor for the TMF life is the amount of in-elastic deformation in each cycle. The inelastic strain range in each TMF cycle is the combination (in equal sizes) of plastic deformation in the cold end of the TMF cycle and creep relaxation in the hot end of the TMF cycle. Furthermore, when the stress response is considered, see fig. 2, a good correlation is found between the stress in the cold end of the TMF cycle (compressive for IP tests and tensile for OP tests) and the observed TMF life times. Both observations described above support the conclusion that the TMF life is governed by the propensity for plastic deformation in the cold end of the TMF cycle. This implies that the low temperature strength is an important factor for the TMF life of IN792. This is in good agreement with the observed influence of minimum temperature on the thermomechanical fatigue for other γ' hardened superalloys [4,12]. The characterization of the deformation structures carried out in the present work also

confirms that mechanisms associated with low temperature deformation of FCC structures are very active during thermomechanical fatigue of IN792.

4.2 Twinning

One of the main deformation mechanisms observed in this study of IN792 is twinning, which is present for both IP and OP tests and for all temperatures. Observations of twinning in nickel-based superalloys have previously been reported for OP-TMF tests [5, 7-8, 13], in situ tensile tests [14] and creep tests [15-16] in the temperature interval 600°- 850°C. This indicates that the rate of deformation is not an important parameter for the formation of twins since several different types of plastic deformation can cause twinning in the material.

In this study twins are most frequently observed for tests with a maximum temperature of 750°C and 850°C and to a lesser extent for tests with a lower (500°C) and a higher (950°C) maximum temperature. It has been reported that different types of pairwise dislocations are moving at low and high temperature [14]. At higher temperature the moving dislocation has the Burger vector $a/6\langle 112 \rangle$ and when these dislocations glide in pairs as Shockley partials in a viscous manner they will be drawing long stacking faults behind them and a twin will be created. This explains why more twins appear in tests with a maximum temperature of 750° and 850°C compared to tests with a lower maximum temperature (500°C). For tests with a maximum temperature of 950°C, the observation of the twins has strongly decreased. This can be explained by the fact that the dislocation can start to climb instead of glide at this temperature. This leads to dislocation climb over the γ' phase, as reported by Jiao et al. [17] for the temperature 950°C, and no stacking faults are created.

Interaction between twins and deformation bands when the material is exposed to compression has been reported by H. Paul et al. [18] for an Cu-8 at.% Al alloy at low temperatures. It was reported that the twins are first to be formed in the material and when the deformation bands are formed, the part of the twins that are captured within the deformation bands is inclined with an angle about 20° to the corresponding part of the twin outside the deformation band. The same observation has been made in the present study for IP TMF-test cycled in the 100° - 750°C temperature range, where the material is loaded in compression at the minimum temperature in the TMF cycle, see fig. 8.

4.3 Deformation structures

Dislocations that are formed during cyclic deformation of a FCC single crystal are typically arranged in different types of dislocation structures such as veins, persistent slip bands (PSB), labyrinth structures and dislocation cells [19]. Studies of cyclic deformation structures formed at room temperature in both single and bi-crystal copper [11, 20] and pure nickel polycrystals [21] have been reported. In copper crystals, the deformation structures consist of different types of deformation bands and slip bands. For pure nickel crystals the deformation structures consist of labyrinth structures, fragmented wall structures, bundle structures, patch structures that are formed along different crystallographic directions.

The typical deformation and dislocation structures formed during cyclic deformation in single and bi-crystal copper and, nickel polycrystal are also formed during TMF testing of IN792 in the present investigation. At lower magnification different types of deformation bands are seen for all temperature, see fig. 5-8, 10-13, 16 and 19. At higher magnification more complex dislocation structures appear, see fig. 5-8 and 10-13. In

order to really know which types of deformation and dislocation structures that are formed more detailed studies needs to be done. However it can be concluded that the dislocation structures formed during thermomechanical fatigue in IN792 are significantly different from those reported to be formed during isothermal fatigue testing of similar alloys [22]. This is because that the dislocation structure is expected to show features which are typical for both low and high temperature deformation mechanisms. Those features are typically high dislocation density in the matrix channels and pronounced cutting of the γ' -phase at low temperature and dislocation networks in the γ - γ' -interface at high temperature [22].

4.4 Recrystallization and oxidation

For the tests that had a maximum temperature of 850° and 950°C, recrystallization is observed at the grain boundaries, around particles and in the deformation structures. The damage typically associated with the recrystallization is void formation within grains as well as at the grain boundaries. This is more distinct for the highest temperature 950°C, see fig. 16 and 19c. The void formation in the present study occurs during OP-TMF testing which imply compressive stresses at high temperature. Recrystallization at the grain boundaries and around particles has previously been observed and reported for other superalloys [23-24]. However, in these cases, the material was first plastically deformed, and then annealed at temperatures in the range of about 1050°-1200°C. It is well known that plastic deformation in general will decrease the recrystallization temperature [25] and the low recrystallization temperatures found in the present study on IN792 can only be explained by a much more severe plastic deformation (at least locally) in the TMF tests compared to the other studies.

The process of recrystallization due to plastic deformation and temperature can be divided into: (1) nucleation and (2) growth of new grains. In the later stage of recrystallization, second phase particles may act as pinning points for migrating boundaries and restrict the growth of the recrystallized grains [24]. The γ' -precipitates therefore play an important role in the recrystallization behaviour of Ni-base superalloys. Naturally, the recrystallization process is significantly enhanced once the γ' -precipitates have been dissolved. However, in the present study the recrystallization has occurred at temperatures well below the γ' -solvus temperature, which according to ThermoCalc [26] predictions is approximately 1175°C for IN792. Thus, it is of interest to study the γ/γ' -structure in the recrystallized regions. The EDX-map in fig. 20 shows that the γ/γ' -structure at least partly is different from that observed in the main material. In some of the recrystallized grains, it appears that the primary γ' has dissolved, since large areas are rich in Cr but depleted of Al, while in other areas the primary γ' seems to have coarsened compared to the main material. The low temperatures at which this process occurs in the present study suggest that the recrystallization is accompanied by a coarsening process of the γ' precipitates (rather than a dissolution and re-precipitation process). The coarsening of the γ' -phase in the recrystallized grains are probably enhanced by the γ - γ' -interface dislocation networks created during the TMF cycling.

Oxidation is not that pronounced for temperatures below 850°C and the γ' depleted zones at the surface and around the cracks are therefore very narrow. However, for specimens tested with a maximum temperature of 850°C and 950°C, the oxidation is more significant and the γ' depleted zones are growing at the surface and around the cracks. Especially at 950°C, the material is strongly attacked and the oxidation causes

large pits in the material, which can act as an initiation point for TMF cracks, see fig.

19c.

Figure 20: EDX-mapping of a recrystallized area after OP TMF 100-950°C testing. The γ' is black in the Cr-map and white in the Al-map

4.5 Crack propagation

The majority of the cracks observed for both IP and OP tests and for all temperatures are transcrystalline which is consistent with the transgranular crack propagation behaviour of other Ni-based superalloys at intermediate temperatures [27]. However, for the IP TMF tests with a maximum temperature of 750°C some tendency to intercrystalline cracking can be observed and it is believed that this behaviour will be enhanced with increasing maximum temperature in the IP TMF tests [9]. However, in the present study no IP tests with a higher maximum temperature than 750°C have been conducted.

All cracks observed in this study have propagated in areas where the material is most plastically deformed, such as in deformation bands. Cracks propagating in deformation bands have also been observed by Zhang et al. [11] and Li et al [20]. They argue that the deformation bands act as nucleation sites for the cracks and that the crack propagation follows the deformation bands. In the present study, twinning is common in the areas, where cracks have propagated and several different observations have been made. One

is when the crack is following the same crystallographic direction as the twins and another is when there are many twins in front of the crack tip, see fig. 10 and 12. The observation of twinning in front of cracks has been reported in a study of a single crystal nickel base superalloy exposed to an OP TMF-test [8] and in a study of pure aluminium exposed to an in situ tensile-test [28]. In case twinning occurs in front of a crack tip in nickel base superalloys some criteria must be satisfied; that is that the crack coincides with a $\{111\}$ planes and that Shockley partials ($a/6\langle 112 \rangle$) are formed in front of the crack. The fact that the cracks and twins have the same crystallographic direction is probably a consequence of that the twins are formed in front of the crack, i.e. the crack itself (surrounded by a plastic zone) promotes twinning.

The last observation is when the crack and the twins have different crystallographic directions, see fig. 6, 11 and 13. This indicates that the twins have been formed before the crack and that this occurs can clearly be seen in fig. 8-9. It has been reported by Zhang [8], that twins can act as the nucleation site for a crack and this has also been observed in this study. However, in most cases the deformation structures act as crack nucleation sites in this study.

5 Summary

The behaviour of a polycrystalline nickel-based superalloy IN792 during thermomechanical fatigue testing can be summarized as follows.

- The majority of the cracks are transcrystalline for both IP and OP TMF-tests. The only exception is for IP TMF-test at 750°C, where some tendency to intercrystalline crack growth can be seen.

- All cracks which are formed in the material have initiated and started to propagate in deformation structures such as deformation bands formed in the most plastically deformed areas of the material. There are also many small surface cracks formed, which typically stop after some growth.
- In the temperature interval 750°-850°C, twins are formed in both IP and OP TMF-tests and this behaviour is observed to be further enhanced ahead of the crack tip. Twins are to a significantly lesser extent observed for tests with a lower (500°C) and a higher (950°C) maximum temperature.
- Recrystallization has occurred at grain boundaries and around particles for tests with a maximum temperature in the interval 850°-950°C. For the temperature 950°C, recrystallization has also occurred within deformation bands. Void formation is frequently observed in the recrystallized areas even for the case of compressive stresses at high temperature.
- Oxidation has a large influence on the damage mechanism for tests with a maximum temperature of 950°C.

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