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Crack-Free Welding of IN 738 by Linear Friction Welding

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Abstract. Inconel 738 (IN 738), like other precipitation-hardened nickel-base superalloys that contain a substantial amount of Al and Ti, is very difficult to weld due to its high susceptibility to heat-affected zone (HAZ) cracking during conventional fusion welding processes. The cause of this cracking, which is usually intergranular in nature, has been attributed to the liquation of various phases in the alloy, subsequent wetting of the grain boundaries by the liquid and decohesion across one of the solid-liquid interfaces due to on-cooling tensile stresses. In the present work, crack-free welding of the alloy was obtained by linear friction welding (LFW), notwithstanding the high susceptibility of the material to HAZ cracking. Gleeble thermomechanical simulation of the LFW process was carefully performed to study the microstructural response of IN 738 to the welding thermal cycle. Correlation between the simulated microstructure and that of the weldments was obtained, in that, a significant grain boundary liquation was observed in both the simulated specimens and actual weldments due to non-equilibrium reaction of second phase particles, including the strengthening gamma prime phase. These results show that in contrast to the general assumption of LFW being an exclusively solid-state joining process, intergranular liquation is possible during LFW. However, despite a significant occurrence of liquation in the alloy, no HAZ cracking was observed, which can be partly related to the nature of the imposed stress during LFW.

Introduction

IN 738 is a nickel-based superalloy with a significant volume fraction of an ordered intermetallic $Ni_3(Al,Ti)$ γ' phase, possessing an excellent high-temperature strength and remarkable corrosion resistance. This makes it a suitable material for manufacturing of hot-section components in aero-engines and in power generation turbine applications where they are exposed to severe operating conditions. Welding techniques are usually employed for both the fabrication and repair of these components. IN 738, like other precipitation-hardened nickel-base superalloys that contain a substantial amount of Al and Ti, is very difficult to weld due to its high susceptibility to heat-affected zone (HAZ) cracking during welding and strain age cracking during post-weld heat treatment (PWHT) [1]. The cause of this cracking during conventional fusion welding processes, which is usually intergranular in nature, has been attributed to the liquation of various phases in the alloy, subsequent wetting of the grain boundaries by the liquid and decohesion across one of the solid-liquid interfaces due to on-cooling tensile stresses. Detailed mechanism of grain boundary liquation cracking in the alloy has been provided elsewhere [2].

To address the problem of liquation cracking in weldments, a recent trend has involved the use of supposedly solid-state welding techniques, such as, friction welding for joining crack susceptible structural alloys. Friction welding generally involves the rubbing together of two components under the influence of an axial force, generating heat at the interface between these components and resulting in the plasticizing of the material at the interface. The friction causing motion is subsequently stopped after a sufficiently plasticized layer is formed. A forging force is then applied to consolidate the weld. Linear Friction Welding (LFW) is a variation of the friction welding process which makes use of heat generated under a reciprocating linear motion. LFW has been reported to be of particular interest for aircraft engine applications, with potential to produce efficient joints in new components, such as, for the fabrication of aero-engine rings and bladed-integrated-disk (Blisk) assemblies, as well as for repair applications [3, 4, 5]. The process has been considered, like other friction welding processes, to be a solid state joining technology. S. Vardhan Lalam et al [6] suggested that since the welding is performed in solid state, extensive migration of elements does not take place, and the welds are also free from segregation, porosity and liquation cracking that are common in conventional fusion welding. M. Karadge et al. [7] specifically mentioned that friction welding avoids melting and re-solidification processes and gives rise to formation of a narrow HAZ due to extremely localized heat generation. Friction welding has also been regarded as a hot forging process in which local melting at the grain boundaries and microfissuring in the HAZ can be avoided [8]. Due to the success of friction welding processes in producing crack-free joints in some aerospace alloys, the present work was initiated to investigate the weldability of the difficult-to-weld IN 738 superalloy by LFW and to study concomitant microstructural changes in the alloy due to the welding process.

Experimental Procedures

Cast IN 738 with a nominal composition of (wt pct) 0.11C, 15.84Cr, 8.5Co, 2.48W, 1.88Mo, 0.92Nb, 0.07Fe, 3.46Al, 3.47Ti, 1.69Ta, 0.04Zr, 0.012B and balance nickel was received in the form of plates having dimensions of 238 mm long × 58 mm wide × 14 mm thick plates. 12.8 mm x 11.1 mm and 17.7 mm high specimens for LFW were machined from these plates. These were given the standard solution heat treatment (SHT) of 2 hours at 1120°C followed by air cooling. The microstructure of the SHT pre-weld IN 738 consisted of coarse primary γ' precipitates with sizes ranging from about 0.4 to 0.8 μm, fine spherical secondary γ' precipitates of about 0.1 μm in diameter, and solidification products that formed during casting, namely MC carbides and γ-γ' eutectic. Figure 1 shows an SEM micrograph of the SHT pre-weld material. The LFW was performed at ambient temperature under prevailing atmospheric conditions using a MTS Linear Friction Welding Process Development System (PDS) located the National Research Council of Canada's Institute for Aerospace Research. Details on the technical specification of the equipment are described elsewhere [9]. The as-welded specimens were sectioned in the axial direction and polished using standard metallographic techniques. Cylindrical specimens, 10 mm in height by 6 mm in diameter, for the Gleeble thermomechanical simulation were cut from the plates using a Hansvedt Model DS-2 travelling wire Electro-Discharge Machine. These specimens were also given the standard SHT of 2 hours at 1120°C followed by air cooling. The LFW simulation was carried out using a Gleeble 1500 – D thermomechanical simulator located at the University of Manitoba. During the simulation process, cylindrical test specimens were heated at a fast heating rate of 150°C/second to temperatures ranging from 1110°C to 1250°C, and held for 0.5 seconds followed by air cooling. All Gleeble simulated specimens were also sectioned in the radial direction and polished using standard metallographic techniques. Microstructures of the welded and Gleeble simulated specimens were examined and analyzed by an inverted reflected-light optical microscope equipped with a CLEMEX vision 3.0 image analyzer (Clemex Technologies Inc., Longueuil, Canada), JSM 5900 scanning electron microscope (SEM) equipped with an Oxford (Oxford Instruments, Oxford, United Kingdom) ultrathin window energy-dispersive spectrometer (EDS) and Inca analyzing software, and JOEL JAMP-5900F Field Emission Scanning Auger Microprobe operating as a high resolution SEM. Prior to microscopic examination, polished specimens were etched electrolytically in 12 mL H_3PO_4 + 40 mL HNO_3 + 48 mL H_2SO_4 solution at 6 volts for 5 seconds. In order to reveal the grain boundaries, one section of the welded specimen was etched with Kallings reagent (2 grams $CuCl_2$ + 33 mL HCl + 33 mL Methanol). Micro-hardness profile across the weld was determined using a Buehler micro-hardness tester.

Results and Discussion

Microstructure of the Welded Material

Microscopic examination of linear friction welded (LFWed) IN 738 showed a sound and crack-free interfacial region, as illustrated by the optical and scanning electron micrographs given in Figures 2a and 2b. An overview of the LFWed joint revealed three different microstructural regions, as indicated in the optical micrograph of Figure 3a. The weld zone (region 1), which formed at the interface between the two LFWed work pieces and extended to about 300 μm into the material on either side of the joint, showed characteristics considerably different from the SHT microstructure (Figure 1) due to the thermomechanical processing that occurred during LFW. SEM analysis of this region (Figure 3b and a higher magnification in 3c) revealed that γ' precipitates, including γ-γ' eutectic constituents, completely dissolved during LFW, while MC carbides survived to the weld line. Another microstructural characteristic was the occurrence of recrystallization within the weld zone of the LFWed IN 738. A recrystallized microstructure along the weld line has been previously obtained in different Ni-based superalloys welded by frictional processes [3, 6, 10]. It has been suggested that the recrystallization occurs dynamically due to the thermomechanical conditions imposed during friction welding, namely the combination of strain at elevated temperatures and high strain rates. Figure 3d shows an optical micrograph of the recrystallized microstructure observed in the weld zone of IN 738. In particular, the maximum amount of plastic deformation would have occurred at the weld line which resulted in the formation of fine fully recrystallized grains. Traversing away from the weld line, the microstructural evolution increasingly exhibited regions of deformed or partially recrystallized grains, such that beyond 300 μm the recrystallization was localized along prior grain boundaries. Overall the observed marked changes in the microstructural characteristics (precipitates, carbides and grain structure) are related to the steep gradients in the temperature and deformation conditions within the weld zone.

The thermomechanically affected zone (TMAZ), region 2, which is between 300 and 600 μm from the weld line, also exhibited a complete dissolution of the γ' precipitates and γ-γ' eutectic constituents, while the MC carbides remained. Hence, the coarse and secondary γ' precipitates that were present in the pre-weld SHT IN 738 dissolved completely in both the weld zone and TMAZ (i.e. 600 μm from the weld line on either side of the joint) and the main microstructural difference is related to the recrystallization characteristics as mentioned above. A higher magnification SEM study suggested the presence of very fine γ' precipitates in both the weld zone and TMAZ. Figure 3e is a high resolution electron micrograph of the weld zone taken with a scanning Auger spectrometer, with an inset showing very fine γ' precipitates of less than 0.1 μm diameter. The occurrence of very fine γ' particles in the weld zone and TMAZ can be related to the fast cooling rate after LFW. This rapid reprecipitation of very fine γ' particles during cooling of the LFWed IN 738 contributed significantly to an increase in micro-hardness of the material close to the weld line, as indicated in Figure 4. Overall, the measurements show an increase in micro-hardness from that of the base material at distances of up to about 2 mm from the weld line. The region within 600 μm from the weld line (weld zone and TMAZ), where reprecipitated very fine γ' particles were observed by SEM examination, correlates with the region where the highest hardness values were recorded. The hardness then decreased gradually with distance away from this region into the base material. These findings are in corroboration with previous work on inertial friction welding of γ' strengthened alloy RR1000, and γ' and γ'' strengthened superalloy IN 718 that also reported an increase in hardness due to the re-precipitation of very fine tertiary γ' particles during weld cooling [11]. In the HAZ, region 3, secondary γ' precipitates that were produced by the pre-weld SHT dissolved completely, while the primary γ' precipitates dissolved partially (Figure 3f). A significant microstructural observation that is generally neither expected nor has been reported in LFWed materials is grain boundary liquation. Nevertheless, grain boundary liquation in the HAZ of the LFWed IN738 was observed to occur, and is discussed next.

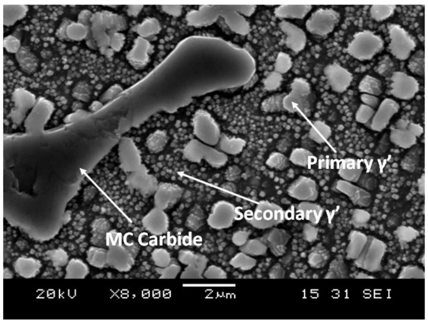

Fig. 1 – SEM micrograph of solution heat treated IN 738 showing primary and secondary γ' precipitates, and MC Carbides

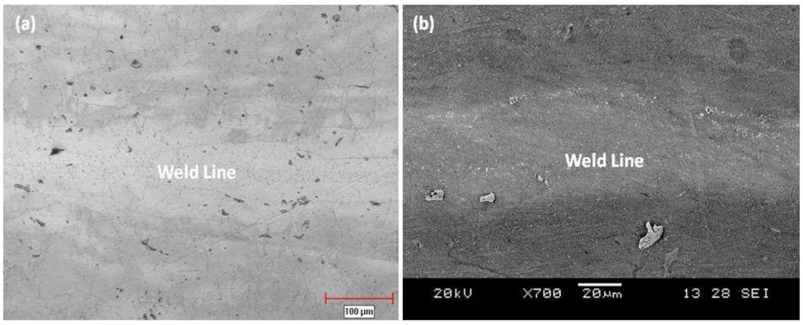

Fig. 2 – (a) Optical micrograph of an LFW joint (b) SEM micrograph of an LFW joint

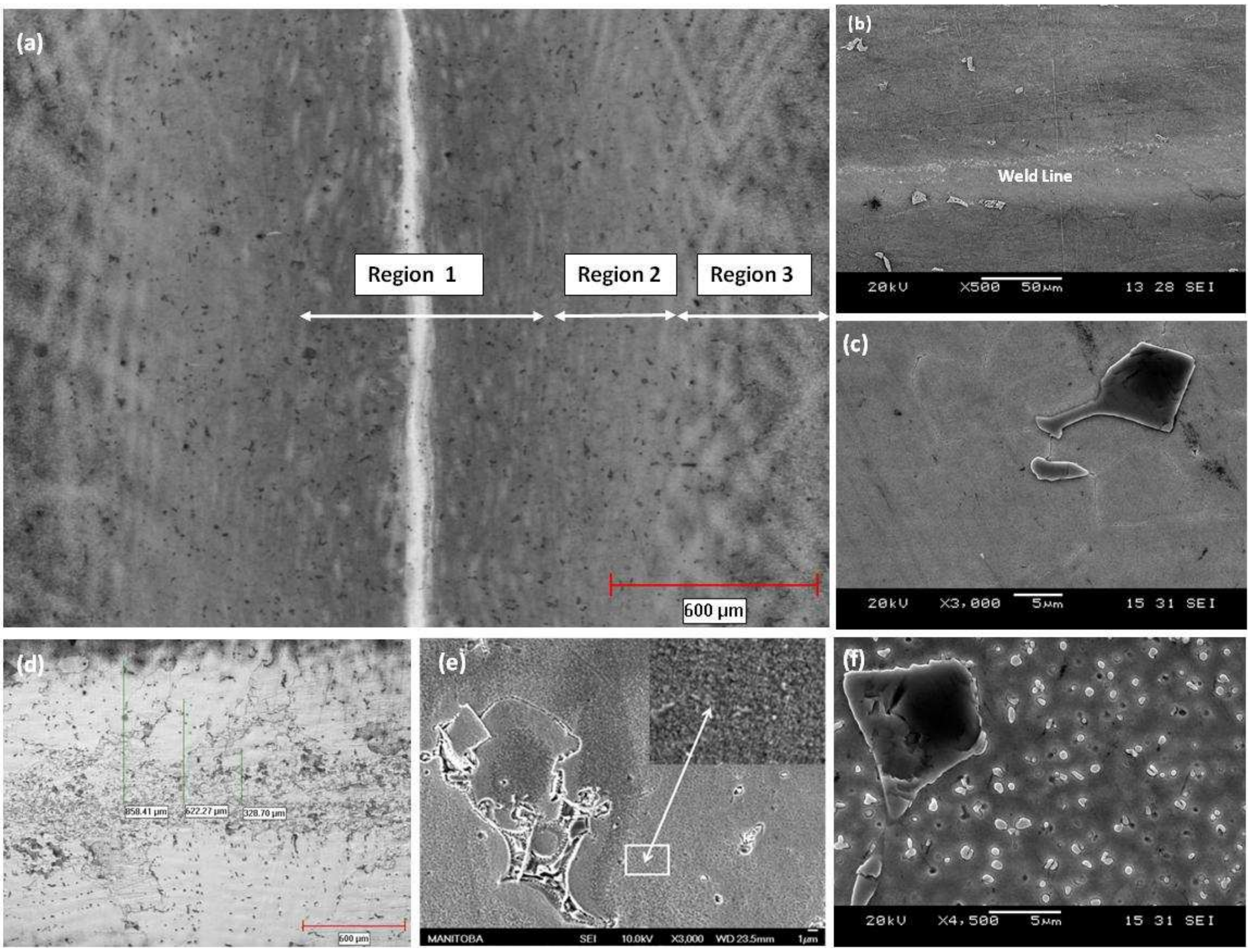

Fig. 3 – (a) Optical micrograph showing an overview of the joint microstructure (b) SEM micrograph showing microstructure around the weld line (c) Higher magnification SEM micrograph of the weld zone, region 1 (d) Optical micrograph showing recrystallization across the joint (e) SEM micrograph showing the very fine γ' precipitates in the weld zone and (f) SEM micrograph of HAZ, region 3

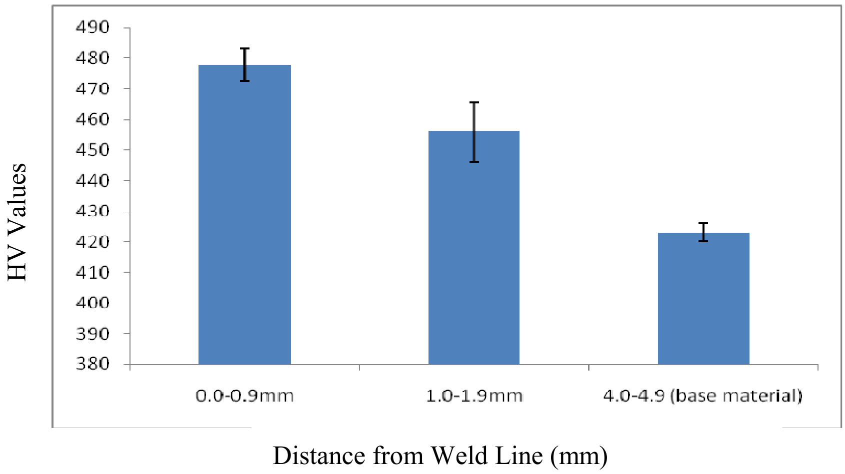

Fig. 4 – Micro-hardness values across the welded joint, which are average of values within a distance of 0 to 0.9 mm, 1.0 to 1.9 mm and 4.0 to 4.9 mm from the weld line. The region within the 4.0 to 4.9 mm was the base material.

Grain Boundary Liquation During LFW

Thermal cycle during LFW often results in dissolution of second phase precipitates and has generally been assumed to occur by solid-state dissolution process, due to the fact that the peak temperature reached during joining is usually below the equilibrium solidus temperature of the base material. Rapid heating during LFW, however, can lead to non-equilibrium phase reactions below the solidus temperature. Notable among these is a non-equilibrium eutectic-type reaction between second phase precipitates and the matrix phase to produce a metastable liquid through a phenomenon also known as constitutional liquation. γ' precipitates, which are the main strengthening phase of precipitation hardened nickel-based superalloys, have been generally reported to undergo solid-state dissolution during friction welding [12]. The degree of particle dissolution depends on interplay between the heating rate and the initial particle size. An attempt has been made to model particle dissolution under rapid heating condition by analytical technique based on an isokinetic and additivity concept and also by control volume numerical calculations [13]. The results of the two methods, which were found to be in good agreement, showed that the degree of particle dissolution depends on interplay between the heating rate and the initial particle size. Soucail et al. [12] in a separate work, studied dissolution of γ' phase in a nickel-based superalloy at equilibrium and under rapid heating condition. Following a different analytical approach, developed by Ashby and Easterling [14], they derived a particle dissolution model and verified it with experimental results. Their analytical results, which are reported to be in good agreement with the experimental results, are qualitatively consistent with those of Bjorneklett et al. [13]. They showed that there was a significant departure from equilibrium under rapid heating condition, in that the temperature of complete dissolution of γ' particles increased with increasing heating rate and the extent of the departure was dependent on the initial particle size. The increase in complete dissolution temperature was found to be more pronounced with an increase in particle size. An increase of about 120°C in the complete γ'dissolution temperature at a heating rate of 8°C/s was reported for γ' precipitates with an initial size of 0.8 μm. Similar effect of heating rate on γ' solvus temperature has also been reported by DTA studies [15]. Therefore, depending upon the initial particle size and heating rate, limited integrated time available for homogenisation by diffusion process during continuous heating can cause γ' precipitate particles in nickel-based superalloys to persist to temperatures well above their equilibrium solvus temperature. Pepe and Savage [16] have proposed that existence of such second phase intermetallic particles at a temperature where they could react with the matrix by a eutectic-type reaction will result in constitutional liquation. There exists a range of temperature in γ' precipitation hardened nickel-based alloys within which γ-γ' eutectic reaction occurs and persistence of γ' particles to this temperature range during continuous heating could result in their constitutional liquation [17].

In IN 738, γ-γ' eutectic transformation has been reported to occur over a range of temperatures, which could be below 1180°C [18]. In the present work, Gleeble simulation of the welding thermal cycle was used to investigate the dissolution behavior of γ' precipitates in IN 738 superalloy. The effect of peak temperature and holding time on transformation of γ' precipitates were studied. The results showed that γ' particles survived well above their solvus temperature to peak temperatures of up to 1250°C, which consequently resulted in their constitutional liquation and subsequent contribution to grain boundary liquation. The extent of liquation increased with increase in the peak temperature (Figures 5a and 5b). A careful microstructural study revealed the occurrence of γ' and grain boundary liquation in the welded material (Figures 3e and 6), which is in agreement with the result of the Gleeble simulation. Therefore, in contrast to the generally held view and reports, grain boundary liquation can occur during LFW. However, despite the occurrence of intergranular liquation, grain boundary liquation cracking was not observed as is usually the case during conventional fusion welding processes.

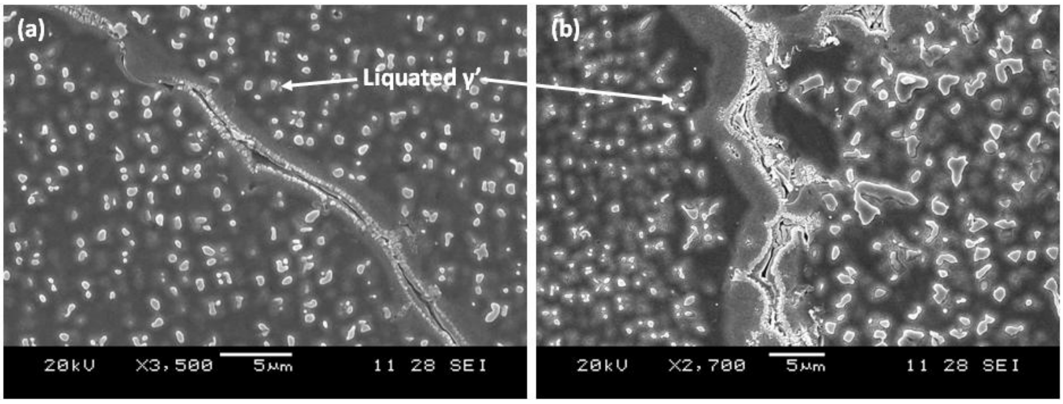

Fig. 5 – SEM micrographs showing grain boundary liquation in Gleeble-simulated specimens at (a) 1170°C (b) 1200°C

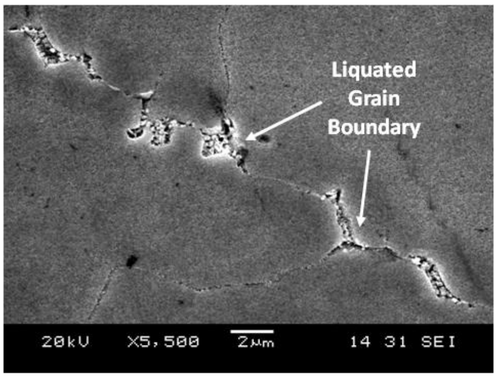

Fig. 6 – SEM micrographs showing liquation along grain boundaries of the LFWed material

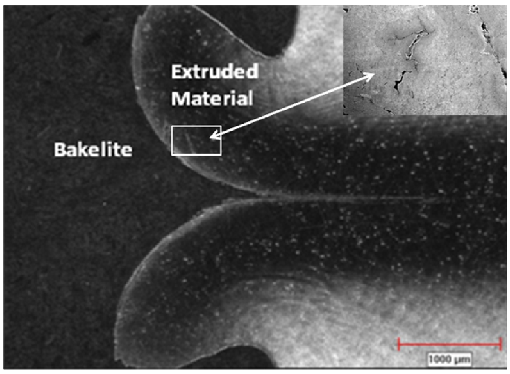

Fig. 7 – Optical micrographs of the extruded flash material, with an inset showing cracks in the material where tensile stresses are present

The presence of a liquid film along the grain boundary is a necessary, but not a sufficient, condition for intergranular liquation cracking, since cracking requires that on-cooling tensile stresses exceed the local strength at one of the solid-liquid interfaces to cause de-cohesion across the interface [2]. In other words, HAZ liquation cracking requires the co-existence of both liquated grain boundary and tensile stresses. The generation of tensile stress is critical for the liquation cracking to occur and must be present in the alloy at a temperature where continuous intergranular liquid film persists. Absence of cracking during LFW, as observed in the present work can be partly related to the state of stress within the work-piece during welding. Numerical simulation of the stress distribution in a material subjected to compressive axial loading, similar to that imposed during LFW, has shown that most of the material is under compressive stresses, while tensile stress is mainly localized to regions extruded away from the joint area [19]. This suggests that imposing compressive load during the forging stage of LFW is not only capable of eliminating one of the two required conditions for liquation cracking, but can also provide resistance to crack formation. The counter-crack-formation nature of compressive loading is evident in a reported experimental work where compressive stress did not only result in crack prevention but also caused healing of short cracks in nickel-based superalloy Rene88DT [20]. Therefore, despite the occurrence of grain boundary liquation, application of compressive load during joining appears to have contributed to preclusion of liquation cracking in the LFWed IN 738 by attenuating the driving force for cracking, which is the presence of tensile stresses during joining. This is supported by the presence of cracks observed in the extruded flash material (Figure 7), displaced from the joint area under the influence of the compressive force, where the state of stress is expected to be predominantly tensile in nature. Therefore, crack-free joining of IN738 by LFW can be considered as an exception to the well know mechanism of HAZ grain boundary liquation cracking whereby the presence of liquation on the grain boundaries ultimately leads to HAZ cracking.

Summary and Conclusions

1. Linear friction welding (LFW) produced a sound and crack-free joint in IN 738, an alloy generally considered to be very difficult to weld due to its high susceptibility to heat affected zone liquation cracking during welding.
2. The LFW of IN 738 was accompanied by significant microstructural changes across the joint, including complete dissolution of the strengthening γ' precipitates within about 600 μm on either sides of the weld line. Very fine γ' particles reprecipitated within this region during weld cooling, which contributed to the increased micro-hardness observed closer to the weld line.
3. In variance to the generally held view and reports, grain boundary liquation occurred during LFW of IN 738. It was observed to occur as a result of non-equilibrium eutectic-type liquation reaction involving the main strengthening phase of the alloy, γ' precipitates.
4. The occurrence of intergranular liquation, however, did not result in liquation cracking in linear friction welded IN 738 material. This was partly due to the compressive state of the stresses within the joint region during welding, which have a tendency to heal rather cause cracking.

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