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Effect of deformation mode on hot ductility of a γ' precipitation strengthened nickel-base superalloy

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Disparity in hot ductility of a γ' precipitation strengthened nickel-base superalloy, IN 738 LC, subjected to non-equilibrium heating and compressive and tensile stresses, was investigated. The alloy, which shows considerable hot ductility at temperatures ranging from 1160 °C to 1250 °C under compressive loading, exhibits zero ductility under tensile loading within this temperature range. The difference is attributed to the fact that while compressive loading permits plastic deformation in spite of non-equilibrium liquid phase dissolution of γ' precipitates in the alloy, the occurrence of the liquation reaction results in inhibition of plastic deformation under tensile loading. Accordingly, while grain refinement through strain-induced dynamic recrystallization occurred under compressive loading, within the same temperature range, the formation of new grains was prevented under tensile loading. This behavior is crucial during high temperature processing of γ' precipitation strengthened nickel-base superalloys.

1. Introduction

Manufacturing of components for applications in the high temperature operating environment of aero and land-based gas turbine engines requires the use of materials with excellent high temperature mechanical properties and reliable hot corrosion resistance. Nickel-base superalloys are usually used for hot-section gas turbine applications. The high temperature performance of nickel-base superalloys can be attributed to a combination of solid solution strengthening by a careful selection of alloy chemistry, precipitation strengthening of the matrix by γ' and/or γ'' phases, and improvements in grain boundary characteristics by the formation of various carbide phases. Cast nickel-base superalloys that are precipitation strengthened by γ' precipitates, such as IN 738 LC, are particularly suited for higher operating temperatures due to better microstructural stability.

Joining and forming processes are important in the manufacturing of gas turbine components. Recent developments in the joining of high temperature materials have led to the use of friction joining processes, including friction stir, linear friction, etc., which have produced excellent joints in materials that are very difficult to join by other conventional methods [1–4]. Also, forming processes, such as forging, extrusion and rolling, remain critically important in the manufacturing of gas turbine components due to the possibility of achieving high production volumes and various shapes [5,6]. Friction joining and forming involve two major events. Firstly, the materials are rapidly heated (non-equilibrium heating) to the processing temperature, which usually results in drastic changes in microstructure. Secondly, the materials undergo significant plastic deformation under externally imposed stresses. Studies have shown that plastic deformation and hot workability of γ' precipitation strengthened nickel-base superalloys depend on the dissolution behavior of the precipitates at processing temperatures [7,8]. Precipitation strengthened nickel-base superalloys, such as IN 738 LC, are usually hot and difficult to work due to their high flow stresses. This necessitates an increase in hot working temperatures, where the precipitates essentially dissolve in order to enhance workability.

One of the phenomena controlling the flow stress and the rate of crack propagation in hot-worked materials is dynamic recrystallization, which occurs in several alloys, including precipitation strengthened nickel-base superalloys [7]. During hot working deformation, several metals tend to exhibit a microstructure consisting of dislocation sub-boundaries similar to the structure obtained after softening of cold-worked materials by annealing but without recrystallization [7]. This is referred to as dynamic recovery and has been observed in different materials [7,9]. High strains and strain rates increase the density of dislocations at the sub-boundaries and cause the dislocations to become more tangled [10]. Due to greater misorientation between the sub-grains, created by the network of dislocations, new grains eventually nucleate, lowering the flow stress and resulting in a steady-state condition where deformation, recovery

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and recrystallization occur at the hot working temperature. The occurrence of such dynamic recrystallization has been observed in several materials [6,7,11]. Dynamic recrystallization and other phenomena associated with hot working of precipitation strengthened nickel-base superalloys depend significantly on the behavior of the strengthening γ' precipitates. Therefore, the behavior γ' precipitates during high temperature processing is crucial to the material's overall response to processing. The objective of this work was to study plastic deformation behavior of γ' precipitation strengthened nickel-base IN 738 LC superalloy subjected to non-equilibrium heating and compressive and tensile stresses. The results are reported in this communication.

2. Materials and experimental procedure

Cast IN 738 LC superalloy, 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 of $240 \times 60 \times 15\ mm^3$ dimension. The plates were given the standard solution heat treatment (SHT) at 1120 °C for 2 h, followed by air cooling. The ductility behavior of the alloy was studied by using a Gleeble 1500-D Thermo-Mechanical Simulation System. Cylindrical Gleeble specimens of 6 mm diameter and 115 mm length were used for the tension tests, while specimens of 6 mm diameter and 10 mm length were used for the compression tests. Gleeble simulations were performed by rapidly heating the specimens at heating rates from 111 $°C\,s^{-1}$ to 150 $°C\,s^{-1}$ and to temperatures ranging from 1100 °C to 1250 °C, while tensile or compressive stresses were applied at the peak temperatures accordingly. Certain specimens were simulated to study the effect of thermal cycle alone, without imposed stress. These were heated to the peak temperatures and held for specific holding times ranging from 0.5 s to 10.5 s, then air cooled. The temperatures of the specimens were controlled and measured during Gleeble testing by spot-welding chromel-alumel thermocouples at the longitudinal axial center of each specimen. Ductility was evaluated in terms of percent change in cross section at the midsection of each sample by measuring the diameter of fractured tensile test pieces and by measuring the diameter of transversely sectioned compression test pieces using a low magnification Nikkon optical microscope. Solution heat treated and Gleeble-tested specimens (sectioned at the location of the spot-welded thermocouples) were prepared by standard metallographic techniques and etched electrolytically in 12 mL H_3PO_4+40 mL HNO_3+48 mL H_2SO_4 solution at 6 V for 5 s for microstructural analysis. Selected samples were chemically etched by dipping in Kallings reagent for 30 s. Microstructures of the materials were examined and analyzed by using a ZEISS Axiovert 25 inverted reflected-light optical microscope equipped with a CLEMEX vision 3.0 image analyzer (Clemex Technologies Inc., Longueuil, Canada) and a 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. The grain structure of a compressed specimen was studied by carrying out electron backscatter diffraction (EBSD) -based orientation analysis using an HKL Nordlys EBSD detector (developed by Oxford Instruments), which was attached to a Philips XL 30 Scanning Electron Microscope and equipped with Oxford Instrument HKL Technology Channel 5 suite of programs. Grain mapping was carried out at a step size of 1.5 μm.

3. Results

3.1. Microstructure of the solution heat treated (SHT) material

The low magnification light-optical image of the solution heat-treated (SHT) material (Fig. 1a) reveals the grain structure of the material before thermo-mechanical processing. Fig. 1(b and c) shows SEM micrographs of the SHT material. The microstructure consisted of coarse primary γ' precipitates with sizes ranging from about 0.4–0.8 μm, fine spherical secondary γ' precipitates of about 0.1 μm in diameter, and solidification products that formed during casting of the plates, namely MC carbides and $\gamma-\gamma'$ eutectics. These microconstituents have been previously reported in SHT IN 738 LC superalloy [12,13].

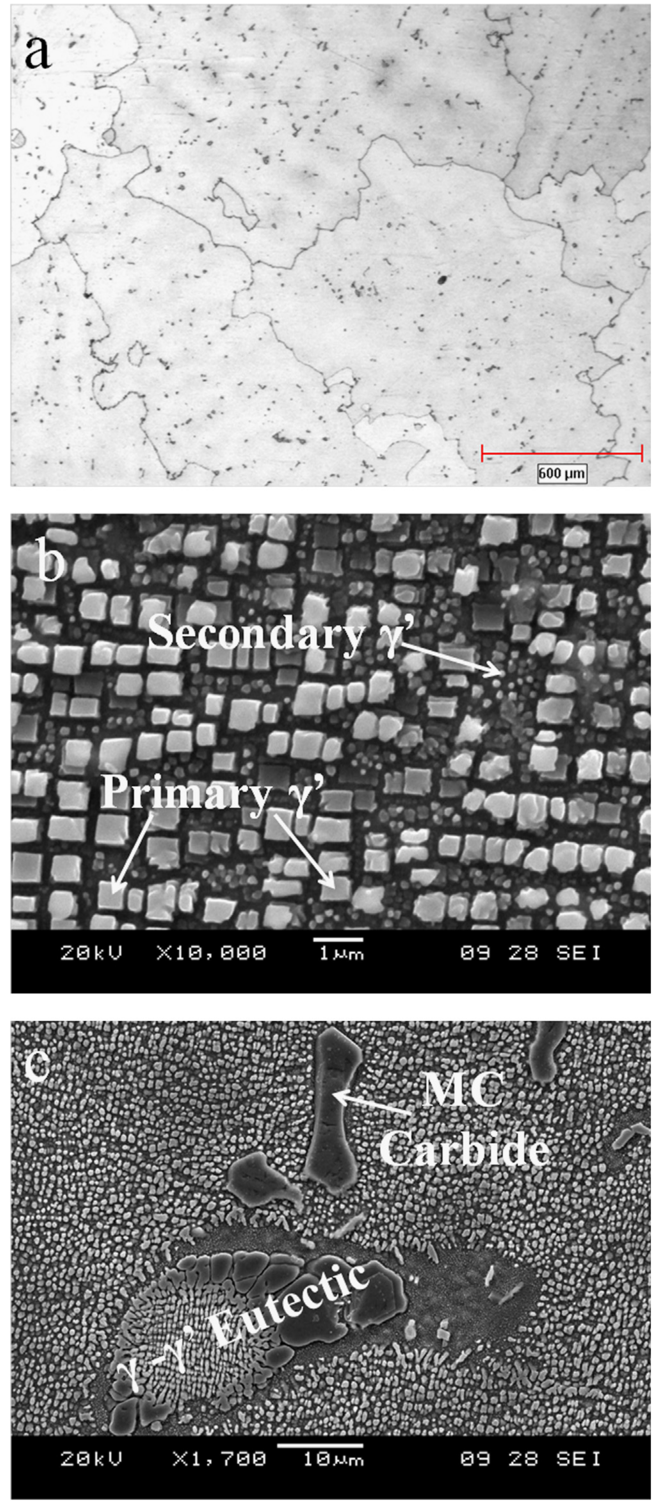

Fig. 1. (a) Low magnification light-optical image of the solution heat-treated IN 738 LC superalloy showing the grain structure, and SEM micrographs showing (b) primary and secondary γ' precipitates, and (c) $\gamma-\gamma'$ eutectic and MC carbide.