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On the γ′ precipitates of the normal and inverse Portevin-Le Châtelier effect in a wrought Ni-base superalloy

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ABSTRACT

The Portevin-Le Châtelier (PLC) effects in a wrought Ni-base superalloy with different γ′ precipitates contents have been investigated. Detailed analysis on the serration type of the tensile curves indicates that the γ′ precipitates have a decisive influence on the transformation from normal to inverse PLC behavior, which is rarely proposed in other works. It is considered that the γ′ precipitates play the same role in PLC effect as temperature and strain rate for the investigated wrought Ni-base superalloy.

1. Introduction

During plastic deformation within a certain regime of temperature and strain rate, various kinds of alloys, from binary systems to very complicated engineering alloys, show a predominant feature of repeated “jerks” or “serrations” throughout the plastic regime [1–5]. Serrated flow in alloys, commonly referred to as the Portevin-Le Châtelier (PLC) effect, is generally considered to be a plastic instability associated with dynamic strain aging (DSA) [1], i.e. dynamic pinning and unpinning interactions between diffusing solute atoms and mobile dislocations during plastic flow. Based on the dependence of critical true plastic strain (ε_c) on the temperature (T) and strain rate ($\dot{\varepsilon}$), the PLC effect could be termed either “normal” or “inverse” [6]. In normal PLC effect, the critical strain for serration formation increases with decreasing temperature and/or increasing strain rate, whereas the critical strain in inverse PLC effect increases with increasing temperature and/or decreasing strain rate. Generally, many dilute solid solutions only show a normal PLC effect at high strain rates and/or low temperatures [7,8]. However, in the concentrated solid solutions with precipitates, i.e. aluminum alloys [9–11] and superalloys [2,4,12–15], a transformation from normal to inverse PLC effect is often observed with increasing temperature or decreasing strain rate [6].

Actually, most of the work conducted previously holds that normal or inverse PLC effect is only dependent on the change of temperatures and strain rates. No work has considered whether the precipitates parameters such as size, mean edge-to-edge interprecipitate distance and fraction could influence the PLC effect. Fortunately, as a main strengthening phase in Ni-base superalloys, the γ′ precipitates possess characteristics of high temperature dissolution and low temperature uniform and coherent re-precipitation in the γ matrix. Most importantly, no structure or composition change of the γ′ precipitates in the temperature range of PLC effect takes place. Consequently, Ni-base superalloys are considered as good candidate materials for examining whether the participates have an influence on the PLC effect. Hayes and Hayes [2,4] and Chen and Chaturvedi [13] studied the precipitates in Waspaloy and Inconel 718 alloy, respectively, and have considered the influence of precipitates on normal and inverse behaviors at various temperatures and strain rates. Therefore, one wonders whether the PLC effect of a Ni-base superalloy with different precipitates parameters would change between normal and inverse behavior at a constant test temperature and strain rate. In this study, a wrought Nimonic 263 Ni-base superalloy with different γ′ precipitates parameters was employed to study the dependence of PLC effect on the γ′ precipitates.

2. Experimental

The nominal chemical composition (wt%) of Nimonic 263 alloy used in this investigation is 20Co, 20Cr, 5.8Mo, 2.1Ti, 0.45Al, 0.7Fe,

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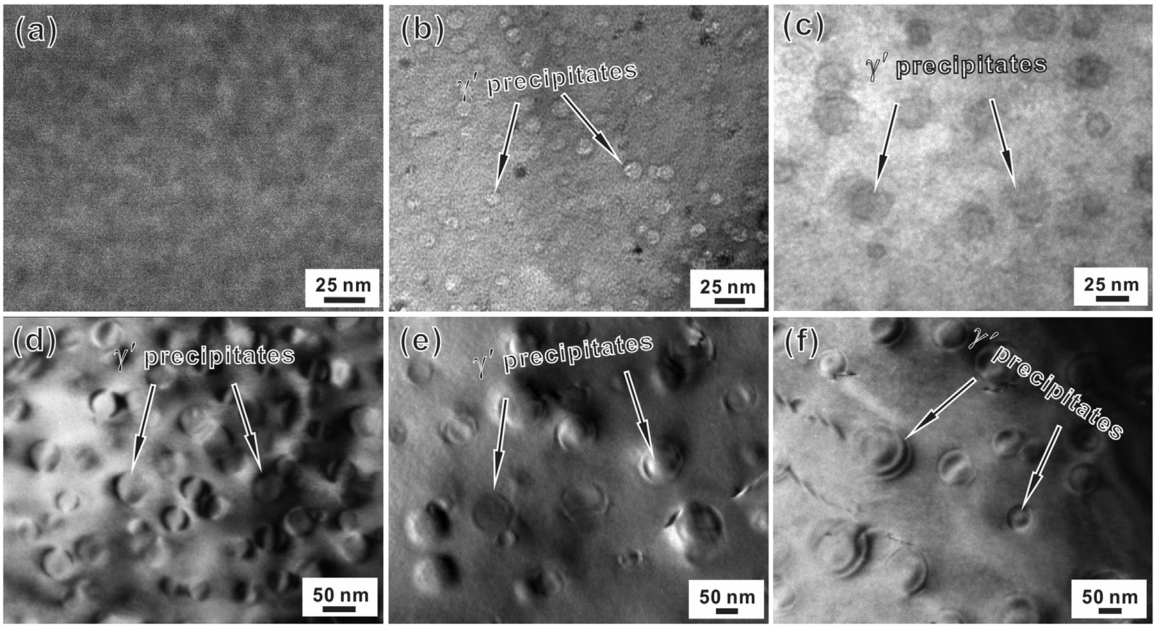

Fig. 1. Microstructures of Nimonic 263 alloy aged at 800 °C for: (a) 0.15 h, (b) 2 h, (c) 25 h, (d) 50 h, (e) 300 h, (f) 500 h.

0.06C, 0.5 Mn, 0.4Si, and balanced by Ni. Ingots were prepared by vacuum induction melting (VIM). The ingots were hot forged at 1100 °C to 30 mm (diam.) bar, in the wake of 1200 °C for 10 h homogenization, and then cut to small bars of 7 mm in diameter and 55 mm in length. The solution annealing for these small bars was conducted at 1150 °C for 1.5 h, followed by rapid water quenching to room temperature for suppressing the precipitation of the γ′ phase. The mean grain size of the as-solutioned Nimonic 263 alloy was about 200 μm. Finally, these small bars were heat treated at 800 °C for different time (0.15, 0.5, 2, 10, 25, 50, 100, 300 and 500 h) in order to achieve different γ′ precipitate parameters.

Specimens with a gage section of 3 mm in diameter and 20 mm in length were used in this work. All tensile tests were carried out on an INSTRON 5582 mechanical test machine at 500 °C with a constant strain rate of 4×10^{-4} s^{-1} in air. Here, all tests were started after holding the specimen for 10 min at the test temperature to stabilize and homogenize the temperature over the gage section. At least two samples under per aging condition were conducted, and the tensile behavior of each aging condition was well repeated. In order to measure the variation of the mean radius $\bar{r}(t)$ and mean edge-to-edge inter-precipitate distance $\bar{L}(t)$ of the γ′ precipitates, thin foils for transmission electron microscopy (TEM) were prepared by mechanical polishing and twin-jet polishing in a solution of 10% perchloric acid and 90% alcohol by volume. TEM examinations of the microstructural configurations in all samples (after aging) were carried out by TEM (JEM 2100) operating at 200 kV. Under each aging condition, at least three areas were examined. The measurement and statistics of the $\bar{r}(t)$ and $\bar{L}(t)$ of the γ′ precipitates were conducted using the Gatan Digital Micrograph software.

3. Results and discussion

3.1. Morphology of the γ′ precipitates

The microstructures of Nimonic 263 alloy after aging at 800 °C for different time are shown in Fig. 1. TEM observation shows that after aging for a short time (less than 2 h), e.g. 0.15 h, no γ′ precipitates in the γ matrix were observed even at a high magnification, as shown in Fig. 1(a). Prolonging the aging time, a large number of small spherical γ′ precipitates were observed (Fig. 1(b), aging Fig. 2h). With aging time increasing further, the size and quantity of the γ′ phase change significantly, whereas the morphology of these γ′ precipitates remains spherical, as shown in Fig. 1(c–f). The statistics of $\bar{r}(t)$ and $\bar{L}(t)$ of the γ′ precipitates with the aging time (over 2 h) are listed in Table 1. In fact, the $\bar{r}(t)$ linearly increases with increasing the aging time and agrees with the basic $\bar{r} \propto t^{1/3}$ kinetics of the Lifshitz-Slyozov-Wagner (LSW) theory [16,17]. Meanwhile, the mean edge-to-edge inter-precipitate distance $\bar{L}(t)$ remarkably increases with increasing aging time. The variation trends as above mentioned indicate that interfacial diffusion controls the growing and coarsening processes of the γ′ precipitates. Notably, a maximum volume fraction (∼17%) of the γ′ precipitates is obtained after aging for 50 h, and further prolonging the aging time does not lead to the continuous increase of volume fraction, which demonstrates that the alloy system has reached a thermodynamic equilibrium state (small γ′ precipitates dissolving and large γ′ precipitates growing up).

Table 1
Variation of the mean radius ($\bar{r}(t)$, nm) and the mean edge-to-edge inter-precipitate distance ($\bar{L}(t)$, nm) of the γ′ precipitates with the aging time.

Aging time (h)	2	10	25	50	100	300	500
$\bar{r}(t)$	6	10	14	21	27	40	48
$\bar{L}(t)$	15	25	35	48	73	105	140

3.2. PLC behavior of Nimonic 263 alloy with different γ′ precipitate parameters

According to the previous work [18,19], in light of applied strain rate, test temperature, and grain size, serration flow can be characterized by different types of strain localization, which depends upon either the propagation features of PLC bands or the morphology of the stress–strain curves. In terms of the morphological characteristics of the serrations themselves, Type A serrations rise above the general flow curve at their onset and are periodic in nature; Type B serrations fluctuate about the general level off low curve in rapid succession, whereas type C serrations fall below the general flow curve and the stress decrement are in a large magnitude. Actually, type A/B and type C serrations are also characterized by the first pinning and the first unpinning effects, and