Lamellar Coarsening in Fe₂₈Ni₁₈Mn₃₃Al₂₁ and its Influence on Room Temperature Tensile Behavior

Ian Baker and Fanling Meng

¹Thayer School of Engineering, Dartmouth College, Hanover NH 03755

Corresponding author: Ian Baker, 14 Engineering Dr, Dartmouth College, Hanover, NH, 03755, U.S.A.; telephone number: 1-603-646-2184; email: Ian.Baker@dartmouth.edu

Abstract

As-cast nanostructured eutectoid Fe₂₈Ni₁₈Mn₃₃Al₂₁, which consists of ~50 nm wide (Ni, Al)-rich B2 and (Fe, Mn)-rich f.c.c. aligned phases, was annealed at 1173 K for times up to 250 h in order to produce a wide range of phase widths (up to 2.5 μm) over which to evaluate the mechanical properties. The coarsening kinetics over this two orders of magnitude change could be described by $r^2 - r_0^2 = kt$, where r is the phase width after time t at 1173 K, r_0 is initial phase width, and k is a constant. The yield strength, σ_y , was found to follow a Hall-Petch-type relationship, i.e. $\sigma_y = \sigma_0 + k_y \lambda^{-1/2}$, where λ is the f.c.c. phase width, σ_0 is the strength for the single-phase material, and k_y is a constant. Interestingly, the work-hardening rate also roughly followed a $\lambda^{-1/2}$ relationship, while the elongation to failure increased as λ increased from failure before yielding in the as-cast material to ~7% for the material with the largest λ values

Keywords: Two-phase alloys, Nanostructure, Mechanical Properties, Eutectoid Transformation

Introduction

In earlier work, we showed, using a combination of scanning electron microscopy, transmission electron microscopy and atom probe tomography, that the as-cast microstructures of $\text{Fe}_{30}\text{Ni}_{20}\text{Mn}_{30}\text{Al}_{20}$ and $\text{Fe}_{25}\text{Ni}_{25}\text{Mn}_{30}\text{Al}_{20}$ (all compositions in at. %), consist of 65-85 nm wide (Ni, Al)-rich B2 (ordered b.c.c.) and (Fe, Mn)-rich f.c.c. aligned plates with a 50-100 nm wide f.c.c. phase was present along all the grain boundaries [1]. A few, very small (~5 nm) f.c.c. precipitates with the same composition as the matrix f.c.c. phase were present in the B2 phase. The two alloys, $\text{Fe}_{30}\text{Ni}_{20}\text{Mn}_{30}\text{Al}_{20}$ and $\text{Fe}_{25}\text{Ni}_{25}\text{Mn}_{30}\text{Al}_{20}$, had high hardness values of 454 HV and 413 HV, respectively.

The idea of adding a ductile f.c.c. phase to a stronger but brittle B2 phase is not new. It has been examined in the Ni-Al-Fe system [2-4], where the resulting multiphase alloys showed a good balance of strength and ductility. Similarly, the lamellar-structured f.c.c./B2 alloy $\text{Fe}_{30}\text{Ni}_{20}\text{Mn}_{35}\text{Al}_{15}$ also shows a good balance between room temperature strength and ductility [5-8]. Thus, the presence of a f.c.c. phase both in the matrix and along the grain boundaries in the FeNiMnAl alloys suggests that they may show some ductility even if the B2 phase is brittle.

In this paper, we show how the mechanical properties of the *eutectoid* alloy $\text{Fe}_{28}\text{Ni}_{18}\text{Mn}_{33}\text{Al}_{21}$, which has a similar microstructure to $\text{Fe}_{30}\text{Ni}_{20}\text{Mn}_{30}\text{Al}_{20}$ and $\text{Fe}_{25}\text{Ni}_{25}\text{Mn}_{30}\text{Al}_{20}$, depends on the f.c.c. phase width over a range of two orders of magnitude (produced by annealing at 1173 K). It is well-known that decreasing the lamellar spacing can substantially increase the yield strength, a feature demonstrated in duplex TiAl alloys [9-12], pearlitic steels [13-15], two-phase Mg-Al-Cr-Ca alloys [16], and two-phase FeNiMnAl alloys [6, 17] amongst others. However, the effect of lamellae width on ductility is more nuanced with decreases in lamellar spacing producing increases in ductility in some studies [16, 18], reductions in ductility in some other studies [11], and having little effect in yet other studies [15]. In contrast, the effect of changes in lamellar spacing on the work hardening rate has rarely been reported.

Experimental

Ingots with atomic composition of $\text{Fe}_{28}\text{Ni}_{18}\text{Mn}_{33}\text{Al}_{21}$ were produced by arc melting under an argon atmosphere pieces of 99.97% Fe, 99.95% Ni, 99.9% Mn, and 99.9% Al: an

additional 5 wt. % of Mn was added to each melt since this element is often lost during melting due to its high vapor pressure. The ingots were melted three times and flipped over between each melting. After melting the ingots were allowed to cool in the water-cooled copper crucible of the arc melter resulting in grain sizes of $\sim 50\text{ }\mu\text{m}$. The compositions of the ingots, measured using a JEOL 8500F electron probe microanalyzer, were found to be close to the target composition.

The as-cast material was annealed at 1173 K for a variety of times from 0.25-250 h, in order to coarsen the microstructure, and cooled in air. The phases present and resulting microstructures were examined by X-ray diffraction (XRD), and by using a scanning electron microscope (SEM), and a transmission electron microscope (TEM) including energy dispersive X-ray spectroscopy (EDS). Details of specimen preparation and instrument operation can be found elsewhere [19].

Four types of mechanical testing were undertaken. Hardness tests were performed on the as-cast alloy using a TIME TH-713 Microhardness Tester with a load of 200 g (1.96 N) and a dwell time of 15 s.

The hardness of the individual phases after they had been coarsened after a 100 h anneal at 1173 K were determined using a Hysitron Ubi-1 nanoindenter using a procedure described elsewhere [6].

Room temperature tensile tests were performed on rectangular cross-section, dog-bone-shaped specimens with a gauge length of 10 mm, width of 2.54 mm and thickness of $\sim 1.27\text{ mm}$ at an initial strain rate of $5 \times 10^{-4}\text{ s}^{-1}$ on specimens after each anneal (three tests per annealing condition). The specimens were polished using silicon carbide papers and finished with $0.3\text{ }\mu\text{m}$ alumina powder to remove any surface cracks. The gauge length was measured before and after tensile testing using an optical microscope and the elongation to fracture determined from these measurements.

For the as-cast material, which showed fracture before yielding under tension, cuboidal ($\sim 3 \times 3 \times 8\text{ mm}$) compression specimens were tested at an initial strain rate of $5 \times 10^{-4}\text{ s}^{-1}$, again with three specimens. These tests were stopped prior to fracture.

The fracture surfaces from the tensile tests were examined using secondary electron (SE) imaging on a FEI XL30 field emission gun scanning electron microscope.

Results

As-cast Microstructure

SE images of as-cast $\text{Fe}_{28}\text{Ni}_{18}\text{Mn}_{33}\text{Al}_{21}$ are shown in Figure 3 with the grain boundary (GB) labeled. The two-phase nature of the matrix with a fine, aligned microstructure is clearly evident. The grain boundary region has a different structure compared to the matrix, suggesting a different phase transformation process [1].

Figure 2(a) shows a bright field (BF) TEM image of as-cast $\text{Fe}_{28}\text{Ni}_{18}\text{Mn}_{33}\text{Al}_{21}$ taken along [100] of the B2 phase. The widths of the phases are ~50 nm. Convergent beam electron diffraction patterns (CBED) from the two phases showed that they were B2 and f.c.c., see Figures 2(b) and (c). The BF TEM image in Figure 2(d) shows the grain boundary phase and Figure 2(e) is a CBED pattern showing that this phase is also f.c.c. In other words, the microstructure is very similar to that of the alloys $\text{Fe}_{30}\text{Ni}_{20}\text{Mn}_{30}\text{Al}_{20}$ and $\text{Fe}_{25}\text{Ni}_{25}\text{Mn}_{30}\text{Al}_{20}$ that were characterized in detail previously, although the phase widths are slightly smaller [1].

Annealing Behavior

The phases present and microstructure after annealing at 1173 K for a variety of times up to 250 h followed air cooling were examined using XRD and a SEM. Figure 3 shows XRD patterns of $\text{Fe}_{28}\text{Ni}_{18}\text{Mn}_{33}\text{Al}_{21}$ after annealing at 1173 K for 1 h followed by quenching into water and after annealing at 1173 K for 20 h and air cooling. In both cases only B2 and f.c.c. peaks are present.

Examination in the SEM showed that the fine aligned, two-phase microstructure of the as-cast $\text{Fe}_{28}\text{Ni}_{18}\text{Mn}_{33}\text{Al}_{21}$, see Figure 4(a), had coarsened into a two-phase, lamellar structure that is no longer aligned even after only a anneal for 0.25 h at 1173 K, see Figure 4(b). The grain boundary f.c.c. is also no longer evident. The width of both phases after the 0.25 h anneal had increased from 50 nm to ~350 nm. Figures 4(b-j) show SE images of $\text{Fe}_{28}\text{Ni}_{18}\text{Mn}_{33}\text{Al}_{21}$ annealed at 1173 K for additional times up to 250 h, at which time the phase width was ~ 2.5 μm . All of them exhibited lamellar-like structure with increasing phase widths as the annealing time increased.

The increase in phase size is plotted on Figure 5 according to the relationship:

$$r^2 - r_0^2 = kt \quad (1)$$

on log scales, where r is the phase width at time t at 1173 K, r_0 is initial phase width, and k is a constant, and shows a good linear fit (correlation coefficient $R^2 = 0.97$).

EDS data for the individual phases, obtained from thin foils in the TEM, is shown in Table 1 for the as-cast condition and after annealing at 1173 K for 20 h, 100 h and 250 h. For all cases, the B2 phase is rich in Ni and Al, while the f.c.c. phase is rich in Fe and Mn. The grain boundary phase has a composition similar to that of the matrix f.c.c. phase. These observation are consistent with the fact that Mn is an f.c.c. stabilizer in iron-rich phases, while Al is a b.c.c. stabilizer [1, 20]. It is interesting to note that after annealing, the B2 phase contains more Fe and Mn, and less Ni and Al than the as-cast material.

There is a clear change in chemistry between the as-cast material and the annealed specimens. This is presumably because the phases in the as-cast material reflect the temperature at which they formed, while the annealed specimens reflect the equilibrium phase compositions at the temperature of the anneal. For the annealed specimens, it is worth noting that with increasing annealing time, the Fe and Al content of the f.c.c. phase decrease while the Ni and Mn content increase with annealing time.

Mechanical Properties

The hardness of as-cast $\text{Fe}_{28}\text{Ni}_{18}\text{Mn}_{33}\text{Al}_{21}$ was measured to be 467 ± 18 HV, a value comparable to that of the similar alloy $\text{Fe}_{30}\text{Ni}_{20}\text{Mn}_{30}\text{Al}_{20}$ at 454 ± 6 HV noted earlier [1].

Nanoindentation was performed on the alloy after annealing for 100 h at 1173 K, which was used to coarsen the phases. This gave values of the hardness of 5.0 ± 0.3 GPa for the f.c.c. phase and 7.3 ± 0.2 GPa for the B2, i.e. the B2 phase is significantly harder than the f.c.c. phase. Similar results were observed for a f.c.c./B2 lamellar-structured eutectic FeNiMnAl alloy [6].

As-cast $\text{Fe}_{28}\text{Ni}_{18}\text{Mn}_{33}\text{Al}_{21}$ was annealed at 573 K for 2 h to reduce any as-cast residual stresses. Tensile tests at room temperature at a strain rate of $5 \times 10^{-4} \text{ s}^{-1}$ were performed on both the as-cast material and after the 573 K anneal. All specimens fractured before any sign of plastic deformation showing failure by transgranular fracture, see Figure 6 (a,b).

In order to obtain a yield strength for the as-cast material, compression tests were also performed on the as-cast $\text{Fe}_{28}\text{Ni}_{18}\text{Mn}_{33}\text{Al}_{21}$ at a strain rate of $5 \times 10^{-4} \text{ s}^{-1}$. The yield stress

of as-cast $\text{Fe}_{28}\text{Ni}_{18}\text{Mn}_{33}\text{Al}_{21}$ in compression is $\sim 1180 \pm 30$ MPa (average of three tests), which is higher than that of f.c.c./B2 two-phase lamellar-structured $\text{Fe}_{30}\text{Ni}_{20}\text{Mn}_{35}\text{Al}_{15}$ at ~ 800 MPa [5].

Room temperature tensile tests at an initial strain rate of $5 \times 10^{-4} \text{ s}^{-1}$ were performed on $\text{Fe}_{28}\text{Ni}_{18}\text{Mn}_{33}\text{Al}_{21}$ annealed at 1173 K for various times up to 250 h corresponding to the microstructures shown in Figure 4. Three tests were performed for each heat treatment condition. Figure 7 shows typical stress strain curves for each annealing time, where it is evident that as the annealing time increased, i.e. as the microstructure became coarser, the yield strength, work-hardening rate and fracture strength decreased, while the ductility increased.

A more ductile fracture mode was found in the annealed $\text{Fe}_{28}\text{Ni}_{18}\text{Mn}_{33}\text{Al}_{21}$, see for example Figure 6(c,d), which is consistent with the increased elongation compared to that of as-cast $\text{Fe}_{28}\text{Ni}_{18}\text{Mn}_{33}\text{Al}_{21}$.

TEM examination of $\text{Fe}_{28}\text{Ni}_{18}\text{Mn}_{33}\text{Al}_{21}$ annealed for 100 h at 1173 K and strained to failure ($\sim 7\%$ elongation) showed that dislocations were present in both the f.c.c. and B2 phases, see Figure 8, indicating that both phases underwent plastic deformation.

Discussion

Lamellar Coarsening

Particle coarsening kinetics are controlled particularly at high temperature, by lattice diffusion which can often be described by the Litshitz, Slyozov and Wagner (LSW) theory [21, 22] where the particle radius increases, r , increases with increasing time, t , according to

$$r^3 - r_0^3 = kt \quad (2)$$

where r_0 is the initial particle size and the constant, k , depends on a variety of parameters including the interfacial free energy and the chemical diffusion coefficient amongst others. Since a plot of equation (2) only has a correlation coefficient, R^2 , of 0.94 versus a R^2 of 0.97 for a plot of $r^2 - r_0^2 = kt$ (Figure 5), the use of the cubic exponent is not immediately compelling. The lack of agreement with LSW theory is perhaps not surprising since that theory was derived for spherical precipitates with a zero volume fraction in a supersaturated matrix.

A model that may be more relevant to the current situation is that developed by Ardell and Ozolins [23] and later by Ardell [24-26] which is based on trans-interface, diffusion-controlled coarsening between ordered and disordered phases with coherent interfaces. This model, which was developed for L1₂-structured (ordered f.c.c.) Ni₃Al precipitates in a disordered Ni-Al matrix, predicts that the average particles size $\langle r \rangle$ growth can be described by $\langle r \rangle^n \propto t$, where $n = 2.4$. Ardell [25] reanalyzed data for a number of alloys and showed that in some cases R^2 was greater for plots of $\langle r \rangle^{2.4} \propto t$ than for the more traditional $\langle r \rangle^3 \propto t$ [21, 22]. The differences in correlation coefficient were often quite small.

Following the approach of Ardell [25], r^n was plotted versus t for several values of n between 2 and 3, and the correlation coefficient R^2 was determined. A plot of R^2 versus n is shown in Figure 9. It is evident that $n = 2$ has the highest correlation coefficient and $n = 3$ has the lowest, although it is worth noting that the range of R^2 values is only 0.94-0.97. As noted by Ardell [25], picking the exponent n based on the correlation coefficient alone is not a good approach. Thus, we note here that even though $n = 2$ has the highest R^2 value, n values from 2-3 could reasonably be used to describe the coarsening data.

Yield Strength

Figure 10 shows plots of the yield strength, the fracture stress and the work-hardening rate measured at 1% strain versus $\lambda^{-1/2}$, where λ is the f.c.c. lamellae spacing. The reason for plotting these parameters against some parametrization of λ is that previous research has shown that in eutectic f.c.c./B2 FeNiMnAl alloys the B2 phase is much stronger than the f.c.c. phase at room temperature. Thus, *initial* plastic deformation occurs *only* in the f.c.c.. Hence, the parameter controlling the strength is λ . Interestingly, once plastic deformation occurs, the f.c.c. phase work hardens and its flow strength can reach the yield strength of the B2 phase, at which point plastic deformation occurs in both phases. This is consistent with the observation of dislocations in both phases of the annealed material that shows significant ductility when strained to fracture (Figure 8).

For most metallic materials [27], the yield strength, σ_y , is related to the grain size, d , through the Hall-Petch relationship [28, 29], i.e.

$$\sigma_y = \sigma_0 + k_y d^{-1/2} \quad (3)$$

where σ_0 is the lattice resistance and k_y is a constant. The relationship is generally applicable well into the nanocrystalline regime [30, 31]. A Hall-Petch relationship also describes the yield strength of nanolaminates in which layers of two different materials are present [32].

In the present work, a Hall-Petch-type plot (Figure 10) of the yield strength versus $\lambda^{-1/2}$ has a correlation coefficient R^2 of 0.93, suggesting the relationship works well over the two orders of magnitude change in λ studied here. Thus, the yield stress of the alloy is consistent with the stress necessary to move dislocations in the f.c.c. phase, which increases with interlamellar spacing refinement [16].

It is of interest to examine whether this relationship is observed in other lamellar materials. Previous work on the *eutectic* alloy $\text{Fe}_{30}\text{Ni}_{20}\text{Mn}_{35}\text{Al}_{15}$ showed a $\lambda^{-1/2}$ dependence of the hardness (which is equivalent not to the yield strength but to the flow stress at a strain of ~8%), but over a shorter range of λ values (approximately one order of magnitude), and for generally larger lamellae sizes (0.5-10 μm) than those studied here. A $\lambda^{-1/2}$ dependence of the yield strength has also been observed in hot-extruded duplex TiAl, which consists of γ -TiAl (L1_0 structure) and α_2 -Ti₃Al (DO_{19} structure) phases, over ranges of lamellar sizes ranging from 30 to 400 nm [33, 34] and from 20-590 nm [12], in directionally-solidified crystals of TiAl with lamellar spacing of 0.37-1.8 μm [10], and in two-phase Fe-Mn-Al-C alloy over a λ range of 60-500 nm [35]. Similarly, a $\lambda^{-1/2}$ dependence of the yield strength is a reasonable fit for various eutectoid and ultrahigh carbon steels [15, 36], although the range of lamellar spacings studied was often quite small, e.g. 125-225 nm [36].

In contrast, a number of researchers [16, 37-39] have found that the relationship $\sigma_y = \sigma_0 + k_y \lambda^{-1}$ was a better fit for the yield strength in a number of pearlitic steels in which the onset of yielding occurred in the ferrite. Similarly, λ^{-1} is also a better description for the yield strength for *eutectic* FeNiMnAl alloys over a range of Al contents [17] in which initial deformation occurs in the f.c.c. phase. For the latter materials [16-17, 37-39] a plot

of σ_y versus $\lambda^{-1/2}$ produces a non-physical, negative value for σ_0 . It is worth noting that the range of λ values examined for the steels and the eutectic *eutectic* FeNiMnAl alloys [16-17, 37-39] were much less than that examined for the materials noted above that showed a $\lambda^{-1/2}$ dependence. Recently, Armstrong [40, 41] has argued that the observed change from an exponent of -1/2 to -1 reflects a change in deformation mechanism from having pile-up at a grain boundary to having a single dislocation expand against a grain boundary.

Work-hardening Rate

The Hall-Petch relationship has been extended to describe the relationship between the flow stress, $\sigma(\epsilon)$, and the grain size at some strain, ϵ , i.e.

$$\sigma(\epsilon) = \sigma_0(\epsilon) + k(\epsilon)d^{-1/2} \quad (4)$$

where the constants $\sigma_0(\epsilon)$ and $k(\epsilon)$ depend on the strain [27,36,42]. Thus, plots of the room temperature flow stress measured at different strains versus $d^{-1/2}$ produced straight lines in aluminum [43] and copper [44] amongst other materials [27], at least up to strains of 20%.

The effect of grain size on the work-hardening rate, defined as:

$$\theta = \frac{d\sigma}{d\epsilon} \quad (5)$$

has received much less attention [36]. Kovács, Chinh, and Kovács-Csetényi [36] developed a phenomenological model for the dependence of θ on grain size and compared the resulting constitutive equations to the work-hardening behavior of 99.99% aluminum strained under tension. They showed that at low strains ($\leq 10\%$), θ could be related to the grain size over the grain size range 40-300 μm by

$$\theta \propto d^{-1/2} \quad (6)$$

In contrast, at large strains θ becomes independent of grain size. In other words, at low strains the deformation within grains is strongly affected by the deformation at the grain boundaries, whereas at larger strains, when the density of statistically stored dislocations is higher, the deformation in the grains becomes independent of the deformation occurring at the grain boundaries.

The work here shows a similar relationship between θ , measured at 1% strain, and $\lambda^{-1/2}$ over a range of λ of two orders of magnitude from 50 nm to 2.5 μm ($R^2 = 0.82$) to that observed by Kovács, Chinh, and Kovács-Csetényi [36] in aluminum, i.e.

$$\theta \propto \lambda^{-1/2} \quad (7)$$

There is very little data on the effect of lamellar spacing on θ . However, in contrast to the observations here, Dollar et al. [16] reported that θ in the pearlitic steels that they studied was not affected by λ over the range 0.13-0.53 μm . Similarly, Karlsson and Lindén [39] reported that the work-hardening rate was independent of the scale of the microstructure for both pearlite and spheroidite.

Ductility

Prior research on the lamellar eutectic f.c.c./B2 alloy $\text{Fe}_{30}\text{Ni}_{20}\text{Mn}_{35}\text{Al}_{15}$ showed that while initial plastic deformation occurred in the f.c.c. phase, fracture was initiated in the B2 phase, which is also more brittle than the f.c.c. phase [6]. In the present work, the elongation to failure increased almost linearly with increasing λ up to $\sim 1 \mu\text{m}$ after which it appeared to be largely independent of λ . Liu and Maziasz [33] found that the room-temperature elongation to failure of dual-phase TiAl alloys depended not on lamellar spacing but on the colony size. Hyzak and Bernstein [14] found that over the λ range 0.12-0.27 μm , the ductility of a pearlitic steel was only very weakly dependent on λ , depending primarily on the prior austenite grain size, a result also found by Lewandowski and Thompson [45] in pearlitic steels. In contrast, Bae, Lee and Nam [36] reported that the reduction in area at the neck in tensile-tested pearlitic steels decreased with both increasing prior austenite grain size *and* lamellar spacing, although the lamellar spacing range studied was only 125-225 nm, and Dollar, Bernstein and Thompson [16] showed that for a fixed grain size of 30 μm and fixed pearlite colony size of 8-10 μm , the elongation to failure in pearlitic steel decreased from 51% to 27% with increasing lamellar spacing over a larger lamellar spacing range of 130-530 nm. In none of the above comparisons did the ductility increase with increasing lamellar spacing as observed in $\text{Fe}_{28}\text{Ni}_{18}\text{Mn}_{33}\text{Al}_{21}$.

Conclusions

The microstructure and mechanical properties of as-cast nanostructured eutectoid $\text{Fe}_{28}\text{Ni}_{18}\text{Mn}_{33}\text{Al}_{21}$ was studied both before and after annealing at 1173 K for times up to 250 h. It was found that:

1. The phase size increase from 50 nm to 2.5 μm after annealing at 1173 K is well described by $r^2 - r_0^2 = kt$, where r is the phase width after time t , r_0 is initial phase width, and k is a constant.
2. The yield strength, σ_y , can be described, over a two orders of magnitude range of λ , by the Hall-Petch-type relationship $\sigma_y = \sigma_0 + k_y \lambda^{-1/2}$, where λ is the f.c.c. phase width, σ_0 is the strength for the single phase material, and k_y is a constant,.
3. The initial work-hardening rate followed a $\lambda^{-1/2}$ relationship.
4. The elongation to failure increased almost linearly with λ from fracture before yield in the as-cast state up to 7% at $\lambda \sim 1 \mu\text{m}$, after which it appeared to be largely independent of λ .

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Table Captions

Table 1: EDS data obtained from TEM specimens showing the composition (in at. %) of $\text{Fe}_{28}\text{Ni}_{18}\text{Mn}_{33}\text{Al}_{21}$ in both the as-cast state and after annealing at 1073 K. Each is an average of three measurements. Note the compositions do not necessary to add up to 100% due to rounding errors.

Figure Captions

Figure 1: SE images of (a) as-cast $\text{Fe}_{28}\text{Ni}_{18}\text{Mn}_{33}\text{Al}_{21}$ showing aligned microstructure with the grain boundary labelled, and (b) the aligned microstructure enlarged.

Figure 2: (a) BF TEM image of as-cast $\text{Fe}_{28}\text{Ni}_{18}\text{Mn}_{33}\text{Al}_{21}$ viewed along [100] of the B2 phase, and (b, c) CBED patterns from the two phases. Note the (001) diffraction spot in (b) indicates the ordered nature of this phase; (d) BF TEM image and (e) CBED pattern of the grain boundary phase in as-cast $\text{Fe}_{28}\text{Ni}_{18}\text{Mn}_{33}\text{Al}_{21}$.

Figure 3: XRD patterns from bulk $\text{Fe}_{28}\text{Ni}_{18}\text{Mn}_{33}\text{Al}_{21}$ annealed at 1173 K for 1 h followed by quenching into water and for 20 h followed by cooling in air. Both show only B2 and f.c.c. peaks.

Figure 4: SE images of $\text{Fe}_{28}\text{Ni}_{18}\text{Mn}_{33}\text{Al}_{21}$ (a) in the as-cast state, and after annealing at 1173 K for (b) 0.25 h; (c) 0.75 h; (d) 0.75 h; (e) 3 h; (f) 5 h; (g) 10 h; (h) 20 h; (i) 100 h and (j) 250 h.

Figure 5. Plot of $r^2 - r_0^2 = kt$, where r is the phase width after time t at 1173 K, r_0 is initial phase width, and k is a constant.

Figure 6. SE images of the fracture surfaces of $\text{Fe}_{28}\text{Ni}_{18}\text{Mn}_{33}\text{Al}_{21}$ (a,b) in the as-cast condition, and (c,d) after annealing for 20 h at 1173 K.

Figure 7: Tensile stress-strain curves for $\text{Fe}_{28}\text{Ni}_{18}\text{Mn}_{33}\text{Al}_{21}$ annealed at 1173 K for different times.

Figure 8. Bright field TEM image of $\text{Fe}_{28}\text{Ni}_{18}\text{Mn}_{33}\text{Al}_{21}$ annealed for 100 h at 1173 K and strained to failure. Note the dislocations in both the f.c.c. and B2 phases.

Figure 9. Plot of correlation coefficient, R^2 , versus the exponent n in the equation $r^n = kt$

Figure 10: Plots of (a) yield stress and fracture stress and (b) work-hardening rate at 1% strain as a function of $\lambda^{-1/2}$, and (c) elongation to failure as a function of λ for $\text{Fe}_{28}\text{Ni}_{18}\text{Mn}_{33}\text{Al}_{21}$. The red lines are fits to the yield strength and working hardening rate data. Each point is the average of three tests. The error bars indicate the standard deviations from three tests.

Table 1: EDS data obtained from TEM specimens showing the composition (in at. %) of $\text{Fe}_{28}\text{Ni}_{18}\text{Mn}_{33}\text{Al}_{21}$ in both the as-cast state and after annealing at 1073 K. Each is an average of three measurements. Note the compositions do not necessary to add up to 100% due to rounding errors.

Alloy	phase	Fe	Ni	Mn	Al
as-cast	f.c.c.	36.5±1.5	9.9±0.8	36.6±0.7	17.1±1.0
	B2	8.6±0.5	36.8±0.3	14.2±0.5	40.4±0.7
	GB phase	37.8±0.8	9.0±0.5	38.0±0.3	15.3±0.5
1073 K, 20 h	f.c.c.	42.7±0.7	3.7±0.4	37.8±0.4	15.9±0.7
	B2	10.6±0.1	40.6±0.2	14.7±0.3	34.3±0.5
1173 K, 100 h	f.c.c.	40.2 ±0.4	5.6±0.1	39.6±0.5	14.6±0.9
	B2	14.5±0.7	31.4±1.3	18.6±1.0	35.7±2.1
1173 K, 250 h	f.c.c.	39.1±0.3	6.5±0.3	40.5±0.5	13.8±0.7
	B2	15.9±0.3	31.4±0.2	19.7±0.1	32.9±0.2

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