JOURNAL OF MATERIALS SCIENCE 25 (1990) 673–676

Effect of nickel addition on the toughness of 18Cr–1Mo stainless steels

HIROFUMI MIYAJI, AKIHIKO HOSHINO, HIROOKI NAKAJIMA
National Research Institute for Metals, 2-3-12, Nakameguro, Meguro-ku, Tokyo 153, Japan

The toughness of ferritic stainless steels has been thought to deteriorate due to carbide and nitride precipitation and also due to grain coarsening. The effect of martensite dispersion on the toughness of 18Cr–1Mo stainless steel was studied by adding nickel up to 4%, by which the γ-loop was extended and an α–γ transformation was introduced. To precipitate austenite homogeneously, proper thermomechanical processing was conducted: rapid hot-rolling without recrystallization in the temperature range just above the two-phase region, followed by precipitation treatment of austenite at the temperature of the two phase region. The ferrite single phase of grain size 100 μm was developed by nickel addition up to 2%, and a micro-duplex structure composed of ferrite and martensite of 10 μm was developed by above 3%. The impact transition temperature in the ferritic single phase decreased remarkably only with 1 to 2% nickel addition and water quenching from solution treatment temperature. In the micro-duplex structure, the transition temperature decreased remarkably with 4% nickel addition and tempering treatment. Martensite has the effect of stopping the crack propagation, and the transition temperature depends mainly on the fracture facet size, which decreases with the amount of martensite.

1. Introduction

In high-purity ferritic stainless steels characterized by low carbon and low nitrogen contents, the deterioration in toughness caused by the precipitation of carbide and nitride [1–4] or by grain coarsening [5] has been well recognized.

Nevertheless, an improvement in toughness would be expected in 18Cr–1Mo stainless steel, because the homogeneous micro-duplex structure of ferrite and martensite can be developed when the proper thermomechanical processing is conducted, and α–γ transformation is introduced by extending the γ-loop with the addition of austenite-forming elements [6].

Among austenite-forming elements, nickel addition is attractive because it improves the toughness of ferritic stainless steels [7–9] although its effect on the duplex structure has not been clarified. The present investigation aims to examine whether the structural change in 18Cr–1Mo steels due to the addition of nickel up to 4% would improve the toughness of these steels.

2. Experimental procedure

According to the phase diagram of Fe–18Cr–Ni alloy [10] as shown in Fig. 1, the two-phase region composed of ferrite and austenite develops at elevated temperatures with nickel addition of a few per cent.

18Cr–1Mo steels containing up to 4% nickel were prepared by melting high-purity raw materials in an induction vacuum furnace. The chemical compositions of five alloys are shown in Table I.

Secondly, it is considered that the dispersion state of the γ-phase precipitated in the α-ferrite matrix varies considerably with thermo-mechanical processing, as shown in Fig. 2. This figure shows that the deformed ferrite must be brought into the two-phase region before the recrystallization occurs to achieve homogeneous dispersion of the γ-phase. For this purpose, 60 mm^2 ingots were heated to 1150° C, just above the two-phase region, for 1 h, and hot-rolled rapidly to 7 mm thick. Then they were heated to 850° C for 1 h, at which maximum amounts of γ-phase should precipitate. The specimens were then cooled by water quenching and air cooling, respectively. The tempering treatment was conducted for the specimens containing martensite at 600° C for 1 h, just below the two-phase region, where the softening extent of the structure reaches maximum, as shown in Fig. 3.

The degree of toughness was decided by measuring the transition temperature of absorption energy by the

TABLE I Chemical composition of specimens (wt %)

Steel no.	C	Si	Mn	P	S	Cr	Ni	Mo	N
1	< 0.005	0.004	0.002	0.003	0.007	17.97	0.010	1.00	0.0048
2	< 0.005	–	–	–	0.006	18.39	0.98	0.99	0.0045
3	< 0.005	–	–	–	0.006	17.96	2.02	0.98	0.0048
4	< 0.005	–	–	–	0.007	18.24	2.84	0.99	0.0049
5	< 0.005	0.007	0.003	0.003	0.006	18.45	3.92	1.00	0.0048

0022-2461/90 $03.00 + .12

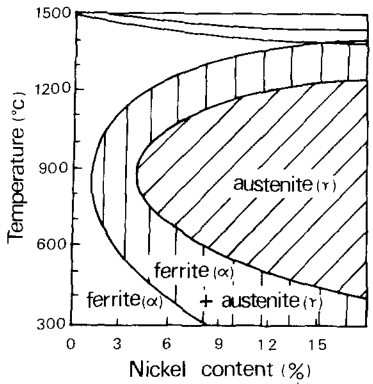


Figure 1 Fe–18Cr phase diagram containing nickel elements [10].

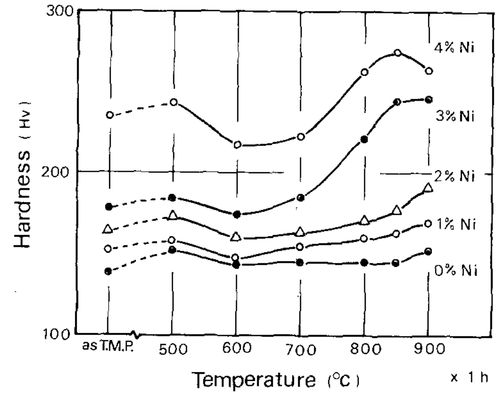


Figure 3 Vickers hardness of the thermo-mechanical processed specimens after quenching from reheated temperatures.

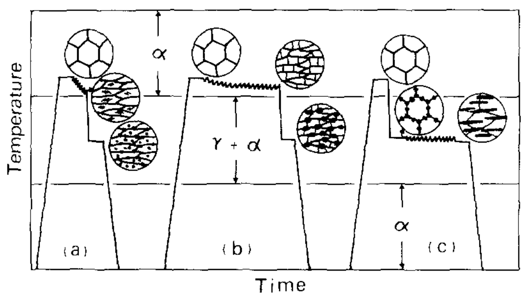


Figure 2 Schematic diagram of various precipitation states of γ-phase due to thermo-mechanical processing. (a) Homogeneous, (b) inhomogeneous and (c) elongated.

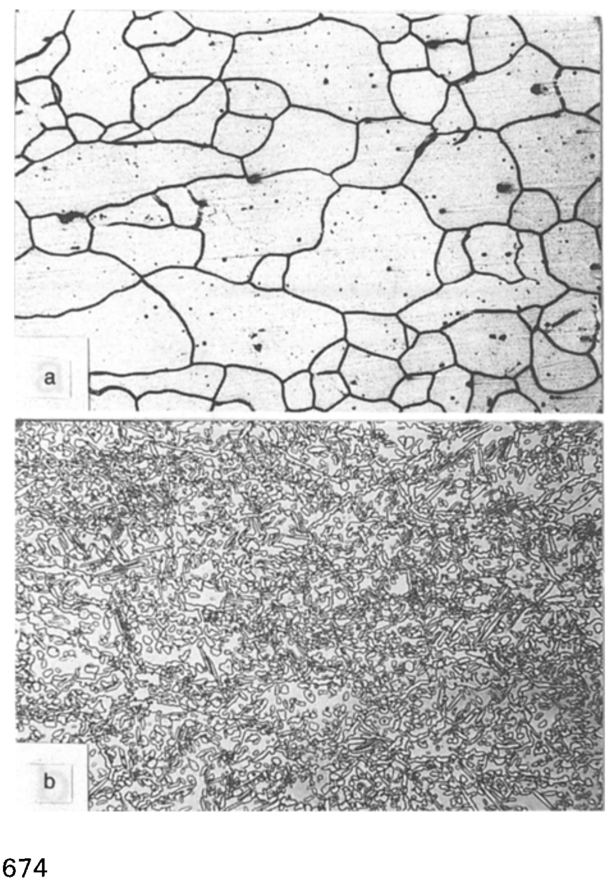
Charpy impact test, using V-notched half-size impact test specimens. The fracture surface was also observed with scanning electron microscopy.

3. Results and discussion

As shown in Fig. 4, the structure of the specimens changed remarkably with the nickel content. The ferrite single phase of grain size 100 μm developed when the nickel content was below 2%, but a duplex structure composed of ferrite and martensite developed above 3% nickel content. At 4% nickel addition, a micro-duplex structure of about 10 μm in the respective phase was developed.

Figure 5 shows the relationship between the transition temperature of each specimen and the nickel content of steels. At first, in the ferrite single phase, the effect of solute nickel appeared obvious in water-quenched specimens. The transition temperature decreased remarkably with nickel addition up to 2%. However, in air-cooled specimens, the effect of nickel addition was not recognized at all. In the duplex structure of ferrite and martensite of the steels containing above 3% nickel, the transition temperature was not lower than that of ferrite single phase in the water-quenched specimens. However, in the air-cooled specimens it decreased more obviously than in the ferrite single phase, but the level was nearly the same as that of the water-quenched specimens. The effect of the cooling rate was not recognized in this case.

On the other hand, the transition temperature of tempered specimens decreased remarkably with 4% nickel addition, but increased with 3% nickel addition. A typical example of a fracture surface in the Charpy impact test is shown in Fig. 6. Figure 6a is for 3% nickel, and 6b for 4% nickel alloy. Both specimens showed cleavage fracture (transgranular fracture), but the fracture facet size between them differed considerably. The fracture facet size L_c was measured for many



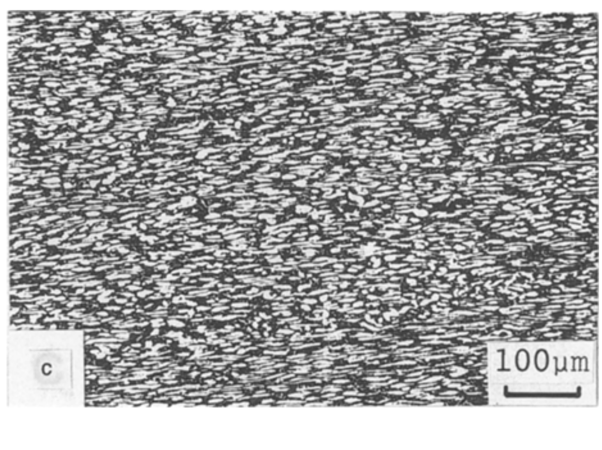


Figure 4 Effect of nickel content on the microstructure after thermo-mechanical processing. Ni = (a) 2%, (b) 3% and (c) 4%.

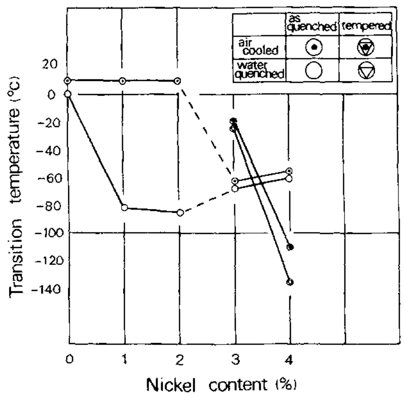


Figure 5 Effect of nickel content and heat treatment on the energy transition temperature.

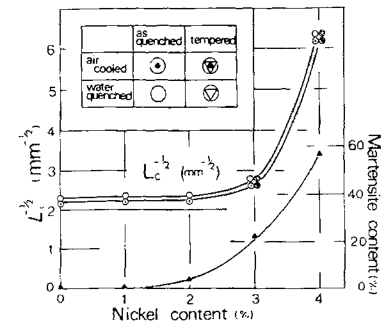


Figure 7 Relation between $L_c^{-1/2}$ and martensite content against nickel content.

specimens. The relationship between $L_c^{-1/2}$ and martensite content was plotted against nickel content, as shown in Fig. 7, which shows that $L_c^{-1/2}$ does not depend on the cooling rate from the solution treatment temperature or on tempering treatment, but depends on the martensite content.

Figure 8 shows a section of nickel-plated fracture surface in this latter case. It is seen from this figure that the martensite phase has the effect of stopping crack propagation, and the fracture facet size seems to become much smaller.

In the ferrite single phase, the transition temperature depends remarkably on the cooling rate from the solution treatment temperature, though the value of $L_c^{-1/2}$ did not change at all with the cooling rate. The reason for this seems to relate to the precipitation of carbide and nitride on the grain boundary [1–4]. The carbon + nitrogen level of 100 p.p.m. in these alloys was responsible for embrittlement. The crack would initiate at these boundaries and nickel addition is not expected to have an effect in this case, even when the matrix is strengthened by solute nickel.

On the other hand, in the as-quenched duplex structure of ferrite and martensite the transition temperature did not depend on the cooling rate. It suggests that austenite transformed to martensite with carbon and nitrogen dissolved completely in itself, and the precipitation of carbide and nitride to the grain

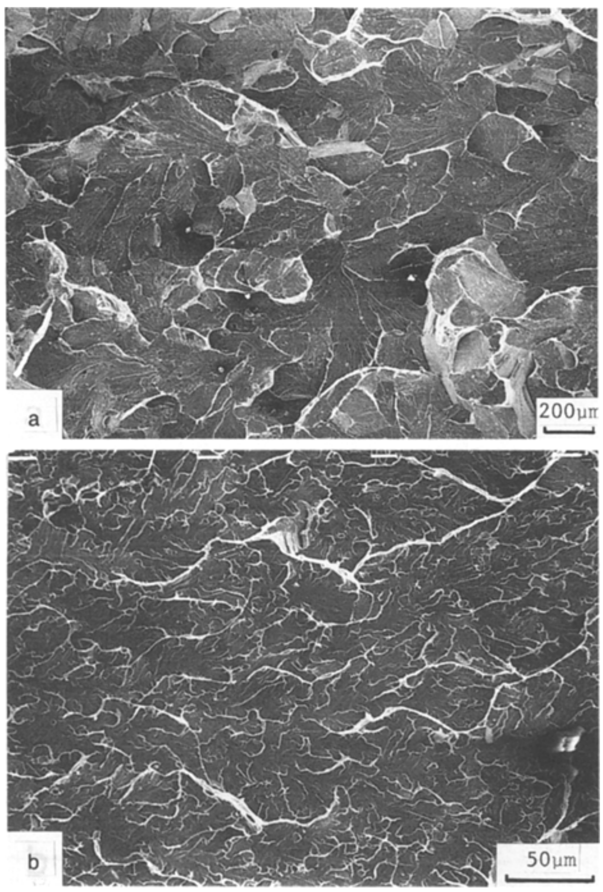


Figure 6 Scanning electron micrograph of the fracture surface. Ni = (a) 3% and (b) 4%.

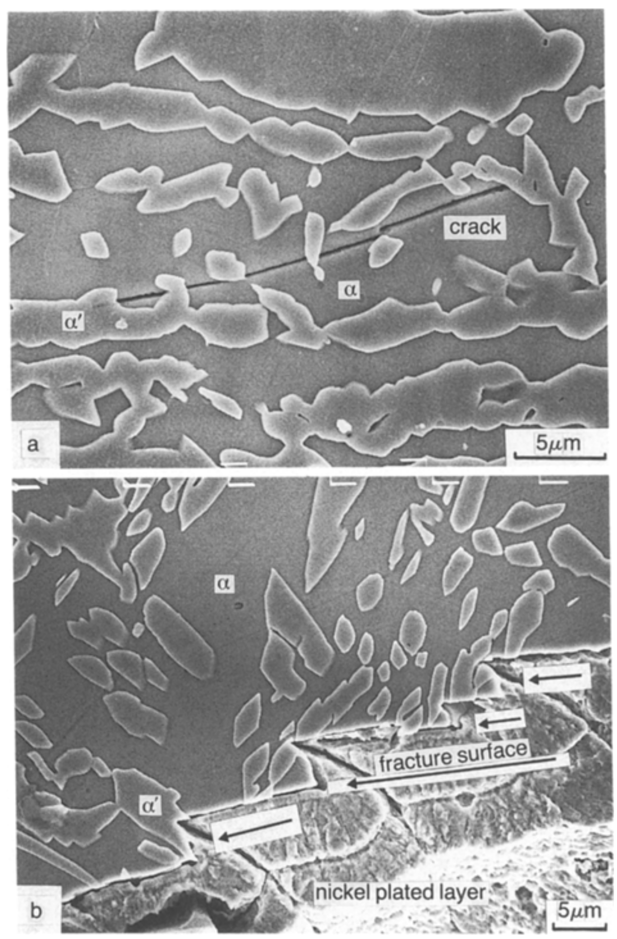


Figure 8 Scanning electron micrograph of the section of nickel plated fracture surface. Dispersed martensite particles have the effect of stopping the crack propagation. (a) Inside the fracture surface and (b) fracture surface.

boundary of martensite, would be suppressed sufficiently. However, in the tempered duplex structure of ferrite and martensite, the transition temperature decreased remarkably with martensite content. The value of $L_c^{-1/2}$, the precipitation of carbide and nitride and also the strength level of the specimen should be considered as factors affecting the transition temperature of the duplex structure. The change in transition temperature with tempering treatment seems to be due to the change in degree of these contributory factors.

4. Conclusions

The structural change with thermo-mechanical processing for 18Cr–1Mo–Ni alloys, in which nickel was added up to 4%, and its effect on toughness, were studied. The results obtained were as follows:

1. In the ferrite single phase, the transition temperature decreased remarkably due to water quenching from the solution treatment temperature by the addition of 1 to 2% nickel. The effect of solute nickel was seen clearly. However, in the case of air cooling, the effect of nickel addition was not seen at all.

2. When austenite formed at elevated temperatures, the homogeneous micro-duplex structure of ferrite and martensite was obtained by the proper thermo-mechanical processing, and its toughness was improved.

3. The toughness of tempered structures containing proper amounts of martensite was also improved.

References

1. M. HASEGAWA, K. TAKEDA and K. TAKESHITA, *Tetsu-to-Haganė* **60** (1974) 1353.
2. H. ABO, T. NAKAZAWA, S. TAKEMURA, M. ONOYAMA, H. OGAWA and H. OKADA, in "Stainless Steel '77", edited by R. Q. Barr (Climax Moly., 1977) p. 35.
3. A. PLUMTREE and R. GULLBERG, *Met. Trans.* **7A** (1976) 1451.
4. B. POLLARD, *Met. Tech.* **1** (1974) 31.
5. R. F. STEIGERWALD, H. J. DUNDAS, J. D. REDMOND and R. M. DAVISON, in "Stainless Steel '77", edited by R. Q. Barr (Climax Moly., 1977) p. 57.
6. H. W. HAYDEN and S. FLOREEN, *Met. Trans.* **1** (1970) 1955.
7. K. MONMA, H. SUTO and N. KIKUYAMA, *J. Jpn Inst. Met.* **31** (1967) 758.
8. K. NAKANO, M. KANAO and A. HOSHINO, *J. Iron Steel Inst. Jpn* **62** (1976) 1219.
9. W. JOLLEY, *Trans. Met. Soc. AIME* **242** (1968) 306.
10. C. Z. ZAPFFE, "Stainless Steel" (American Society for Metals, Cleveland, 1949) p. 136.

Received 28 November 1988
and accepted 8 May 1989