

## Effect of alloying elements on the Mechanical property of Austenitic Stainless Steel

Dong-yuk Kim<sup>a</sup>, Sung-sik Kang<sup>a</sup>, Hyun-je Sung<sup>b</sup>, Sung-joon Kim<sup>b\*</sup>

<sup>a</sup>Korea Institute of Nuclear Safety, 62, Gwahak-ro, Yuseong-gu, Daejeon, 34142..

<sup>b</sup>POSTECH, Graduate Institute of Ferrous Technology, 77, Cheongam-RO, Nam-Gu, Pohang, Gyeongbuk, 37673

\*Corresponding author: sjkim1@postech.ac.kr

### 1. Introduction

Austenitic stainless steel must be one of essential structural materials in a nuclear power plant due to high strength, ductility, toughness, and corrosion resistance. In particular, the integrity of austenitic stainless steel pipe is so important in the nuclear energy system that it should be processed for a specific operational purpose. Most of the austenitic stainless steel could be developed by the addition of alloying elements contributing to mechanical properties significantly. However, there are still a lot of the unidentified mechanism about alloying elements leading to change of mechanical properties of austenitic stainless steel. Therefore, its research results which have recently been issues with regard to austenitic stainless steel were confirmed in this report.

### 2. Methods and Results

Among the austenitic stainless steel, the Type 304, 316 and 347 are commonly used in the nuclear power plant system. Type 304 typically could be applied for in coolant pipe, internals, pump, and valve. However, stress corrosion cracking (SCC) is a common mode of failure encountered in Type 304 so that a new type of austenitic stainless steels called Super 304H containing 3% Cu is being adopted. Moreover, Type 316L was used widely in order to avoid the sensitization ( $Cr_{23}C_6$ ) by carbon and increase the pitting resistance with Mo. Therefore, the mechanism about effect of alloying elements with above-mentioned austenitic stainless steels were investigated through creep test, rupture test and stress test with TEM analysis.

#### 2.1 Effect of N and Cu on the precipitation and the creep life of 316L at 650°C

It is obvious that the YS at room temperature increases with N concentration (Fig. 1a). However, the effect of Cu is trivial; it decreases slightly with 2wt% Cu. UTS at room temperature maintains the same trend as the YS for the effects of both N and Cu. Moreover, similar behavior is observed at higher temperatures (Fig. 1b).

Since the microstructure is fully austenitic, the effect of N is attributed to solid solution strengthening. Since Cu is incorporated as a substitutional alloying element in steels, its effect is smaller than N which is an interstitial alloying element [1,2].

The addition of N and Cu shows a similar trend in the creep life. N effectively increases the creep life under all

stresses while the addition of Cu starts to improve the life when the stress becomes lower than 125 MPa.

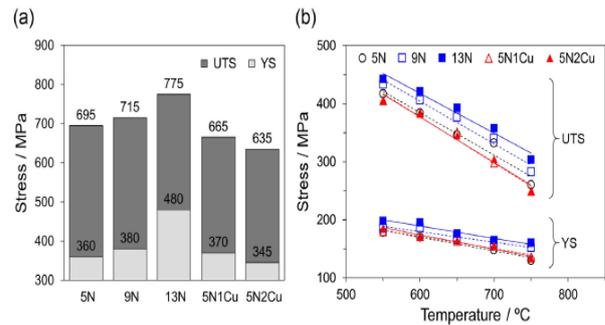


Fig. 1. Yield strength (TS) and ultimate tensile strength (UTS) at (a) room temperature and (b) high temperatures.

As the concentration of N increased, the size of  $M_{23}C_6$  significantly reduces as shown in Fig. 2. After 300h under 125MPa at 650°C, 5N shows  $\sim 1\mu\text{m}$ -sized  $M_{23}C_6$  particles while they remain smaller than  $\sim 200\text{nm}$  in 13N.

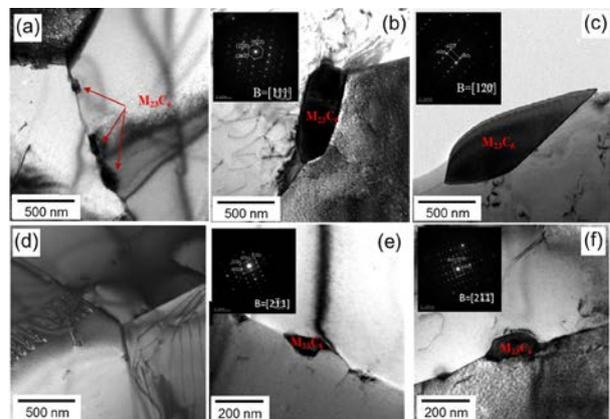


Fig. 2. Representative  $M_{23}C_6$  particles of (a,b,c) 5N and (d,e,f) 13N observed after the interrupted creep tests. (a,d) after 10h, (b,e) after 100h, and (c,f) after 300h under 125MPa at 650°C. Note that the magnification for (e,f) is higher than others due to the small size of the particles.

The addition of N in 316 stainless steel suppresses the coarsening of  $M_{23}C_6$  either by reducing  $M_{23}C_6$ /austenite interfacial energy [3] or by decreasing the diffusivity of either Cr or C [4]. Since large precipitates at the grain boundaries can act as nucleation sites of a crack, it is believed that the creep life of N-containing alloys are enhanced by the delay in the coarsening of  $M_{23}C_6$ .

More particles are observed in 5N2Cu than in 5N1Cu as shown in Fig. 3, which would be due to higher Cu concentration and longer holding time before the rupture of 5N2Cu. The particles are fine and located within austenitic grains, which is consistent with the Cu-rich phase observed in Super 304H austenitic stainless steel [5-7].

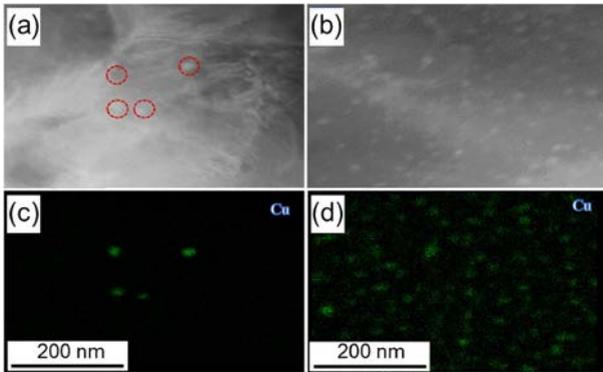


Fig. 3. Scanning transmission micrographs of the Cu-rich phase precipitates observed in (a) 5N1Cu and (b) 5N2Cu, which ruptured under 125MPa at 650°C (c,d) corresponding energy dispersive X-ray spectroscopy Cu-maps of the particles in (a,b), respectively.

Owing to its distribution and small size, we believe that the Cu-rich precipitate strengthens the matrix, resulting in creep life improvement. Under high stresses, the Cu-rich phase has hardly precipitated in both 5N1Cu and 5N2Cu, since they all failed too early.

### 2.2 Coherent to incoherent transition of precipitates during rupture test in Super304H

MnS precipitates form at the early stage of the rupture test due to the fast diffusion rate of sulfur and supply a heterogeneous nucleation site for the other precipitates [8-10]. The premature IG cracking occurring in the HAZ (Heat Affected Zone) of heat-resistant steels has been clarified in the light of impurities (S, P) segregation to GBs [11,12]. In particular, the austenitic stainless steel is highly sensitive to the sulfur segregation.

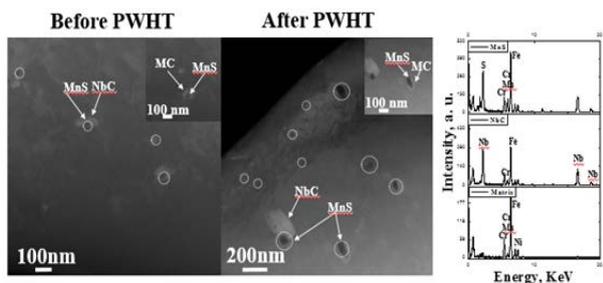


Fig. 4. HAADF-STEM images obtained from fractured specimen: MnS distribution increase in the Super 304H after PWHT

However, the intergranular corrosion by sulfur segregation can be avoided because of the formation of MnS precipitates with post weld heat treatment as shown in Fig. 4. The MnS precipitation reaction retards the sulfur segregation to the grain boundary of austenite stainless steel due to the role of the MnS/  $\gamma$  matrix interface as a sink of free sulfur [13].

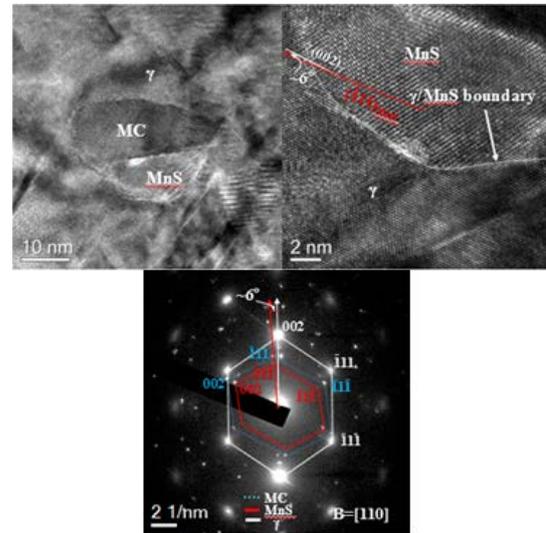


Fig. 5. Coherency loss in the interface between the MC+MnS complex and the  $\gamma$  matrix due to the growth of MC.

In addition, as the MC formed on the surface of the pre-formed MnS particle grows, the coherency between the MnS and the  $\gamma$  matrix is lost. The interfaces of the MnS and the MC with  $\gamma$  matrix are a {111}-type of planes. The SADP supports that the MnS and the MC show still the cube-cube OR. However, the (111) planes of the MnS and the MC are 6° deviated from the (002) $_{\gamma}$  plane in Fig. 5, which deviation angle increases as MC grows. It is attributed to the longer time to intergranular failure. This is due to the higher precipitation of MnS and the larger area fraction of the incoherent MnS+MC complex/  $\gamma$  interface which can act as a stringer sink of free sulfur.

### 3. Conclusions

N in solid solution effectively strengthens the matrix, which improves yield strength and tensile strength while the effect of Cu on tensile strength is minimal. The addition of Cu remarkably improves the creep life under low stress due to the fine precipitation of the Cu-rich phase. However, the creep life under high stress is not affected since the alloy ruptures before the formation of the Cu-rich phase.

Since the higher precipitation of MnS and the incoherent interface between the MC+MnS complex/ the  $\gamma$  matrix which acts as an active sink of free sulfur segregating to grain boundaries, the formation of MnS could be attributed to the longer time to intergranular

failure with sulfur segregation.

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