Cladding Development Activities for SFR Fuel

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1. Introduction

Ferritic/martensitic steel has been used and/or considered as a primary candidate material in the sodium-cooled fast reactor (SFR) fuel because of its low radiation-induced swelling and high resistance to irradiation hardening and embrittlement [1,2]. The mechanical capability, including tensile, creep and hardness, of the steel (e.g. HT9) has been verified up to very high dose (≥ 200 dpa) in a fast fission reactor facility. The steel for the fuel cladding are commonly used in a 'normalized and tempered' condition. This heat treatment involves a solutionized treatment (austenitizing) that produces austenite and dissolves the $M_{23}C_6$ carbides and MX carbonitrides, followed by an air cooling that transforms the austenite to martensite. Precipitation sequence during a long-term creep exposure is strongly influenced by the distribution of those in the heat treated condition of the steels. Their creep strength has been improved by their martensitic lath structure, the precipitation strengthening effects of $M_{23}C_6$ carbides and MX carbonitrides and the solid solution strengthening effects of Mo and W in the matrix. Especially, the precipitation strengthening effect of MX is important because its coarsening rate is low and a fine particle size is maintained for a long-term creep exposure. Z-phase formation from MX-type precipitates has been proposed as a degradation mechanism for a long-term creep regime.

The ferritic/martensitic steels should be improved their performance in order to utilize in the high burn-up fuel cladding. For this purpose, KAERI has developed advanced ferritic/martensitic steels since 2007. This paper includes two categories; the first explains some results of the performance evaluation regarding mechanical and microstructural properties of the advanced steels. The second one is comprised of the works related to cladding tube fabrication in Korea.

2. Development of Advanced Cladding Steel

2.1. Effects of Ta addition [3]

Large precipitates (200~300 nm) were displayed at the prior austenite grain boundaries, which were identified as $M_{23}C_6$-type chromic carbides. Large NbC precipitates were also observed intermittently on the prior austenite grain boundaries. In addition, the $M_{23}C_6$-type precipitates were located on the lath boundaries of the martensitic structure. Nitride, carbide and carbonitride such as NbC, NbN, VN, Nb(CN) and (NbV)C were observed both at the lath grain boundaries and within the lath regime. These precipitates were MX-type precipitates and their sizes ranged from 10 to 40 nm. In the case of the Ta-containing steels, Ta was mainly detected within the MX-type precipitates including Ta of 5~20 at.%. And the Ta concentration within the MX-type precipitates increased with an increase of the Ta content as an alloying element.

Vickers hardness increased with the concentration ratio of Ta to (Nb+V+Ta). The degree of hardness increase was also enhanced with an increase of the Ta concentration in the steels. YS and UTS increased slightly with an increase of the concentration ratio of Ta to (Nb+V+Ta). But the elongation was the lowest value at 0.2 of the ratio.

2.2. Effects of B addition [4]

The boron in the experimental steel would be distributed at the interface between the $M_{23}C_6$ carbides and the matrix, and within the carbides. The boron mapping by SIMS also indicates that the boron is distributed along the prior austenite boundaries. The boron addition could suppress the coarsening of the $M_{23}C_6$ carbides and the nucleation of the Laves phases that precipitate near the prior-austenite boundaries during a long-term creep exposure. This would improve the creep resistance of the boron-added steel when compared with that of the plain steel.

The presence of boron in the carbides could retard the coarsening rate of these carbides and the recovery of a dislocation substructure during a long-term creep exposure. The beneficial effect of a boron addition to the boron-added steel was observed in the creep environments at 650°C.

2.3. Effects of lower Nb/Higher B addition [5]

The $M_{23}C_6$ carbides were formed as major precipitates around the prior-austenite and lath boundaries. A small content of MX-type carbonitride precipitates (Nb-rich

MX and V-rich MX) was observed on both boundaries and in their interiors, as shown in Fig.4. The Ta below 10 at.% was incorporated into the Nb-rich MX precipitates.

The higher Nb amount and additional Ta in the steels could enhance the creep rupture stress at 650oC since the Nb-rich MX precipitates would be stabilized by the incorporation within the precipitates. And the combination of the lower N and higher B could improve its creep properties due to an inhibition of BN precipitation on both of the martensitic lath and the prior austenite grain boundaries.

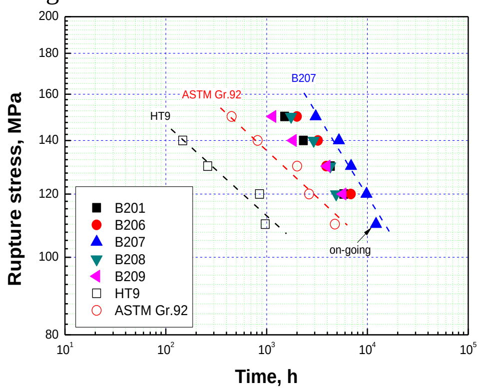

Fig. 1 Creep properties of advanced cladding steel (B207)

3. Fabrication of HT9 Cladding Tube

3.1. Mother tube fabrication

In cooperation with a domestic steel maker, mother tubes for the cladding fabrication were prepared in 2010. HT9 ingot was prepared by using a VIM (vacuum induction melting) method on a 1,000 kg scale. The ingot was hot-forged at a γ-phase temperature of 1,170oC to shape as round bar (150 mm in diameter). The round bars were heat-treated at 800^{o}C for 4h and then air-cooled. Hollow billets drilled from the round bars were heated to 1,170^{o}C by an induction heater and then extruded to 46-mm tubes by a 2,000 kg press. The extruded tubes were also heated at 800^{o}C for 2h and picked in an acid solution. The heat-treated extruded tubes were fabricated to 32-mm tubes via 3-time cold drawings (i.e. 46 mm 40 mm 36 mm 32 mm in diameter reduction). After each drawing was carried out, the intermediate annealing and picking were performed at 800^{o}C for 2h and picked in an acid solution, respectively. To fabricate mother tube (19.05 mm in diameter), the 32-mm tubes were pilgered to mother tubes, heat-treated at 800^{o}C for 2h, and then pickled in an acid solution. All mother tubes were inspected to check internal defects and dimensional variations in viewpoint of the product quality control.

3.2. Cladding tube fabrication

The qualified mother tubes were fabricated to final cladding tubes by using the drawings, which were repeated by five times as following schedule; 19.05 mm → 15.8 mm → 12.7 mm → 9.5 mm → 8.5 mm → 7.4 mm in diameter reduction. Cleaning and intermediate annealing were introduced to remove the foreign matters and oxide scales at each reduction step. The oxide scales were removed by a heat treatment at 1,050^{o}C under the hydrogen (H_2) atmosphere and then the stress-relived annealing were followed at 800^{o}C. After each intermediate process, the straightening calibration was performed to obtain the straightness of the products. Final cladding of 7.4-mm tubes were normalized at 1,038^{o}C and followed by tempering treatment at 760^{o}C for 2h. The final cladding tubes were qualified by the final straightening calibration, surface polishing and internal defect inspection. The quality of the final cladding, including dimensional variation, straightness, surface state and internal defect, was satisfied within the specification limits for SFR fuel cladding.

4. Summary

Based on the coupon test, the advanced cladding steels were showed superior creep resistance to references (ASTM Gr.92 and HT9). Sample cladding tubes of the HT9 and ASTM Gr.92 steels were prepared in 2011. The sample cladding tubes of the advanced steel will be prepared in coming years. Three kinds of cladding tubes (HT9, ASTM Gr.92 and Advanced steel) will be evaluated for out-of-pile performance by 2016. KAERI also plans to evaluate the in-pile performances of the tubes in a fast research reactor in foreign countries.

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