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# Development of new generation reduced activation ferritic-martensitic steels for advanced fusion reactors[☆]

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## ABSTRACT

International development of reduced activation ferritic-martensitic (RAFM) steels has focused on 9 wt percentage Cr, which primarily contain $M_{23}C_6$ (M = Cr-rich) and small amounts of MX (M = Ta/V, X = C/N) precipitates, not adequate to maintain strength and creep resistance above ~500 °C. To enable applications at higher temperatures for better thermal efficiency of fusion reactors, computational alloy thermodynamics coupled with strength modeling have been employed to explore a new generation RAFM steels. The new alloys are designed to significantly increase the amount of MX nanoprecipitates, which are manufacturable through standard and scalable industrial steelmaking methods. Preliminary experimental results of the developed new alloys demonstrated noticeably increased amount of MX, favoring significantly improved strength, creep resistance, and Charpy impact toughness as compared to current RAFM steels. The strength and creep resistance were comparable or approaching to the lower bound of, but impact toughness was noticeably superior to 9–20Cr oxide dispersion-strengthened ferritic alloys.

## 1. Introduction

Reduced activation ferritic-martensitic (RAFM) steels for fusion reactor structural applications have been under development for about three decades. These steels have converged to a nominal 9 wt percentage (wt%) Cr content, largely because this minimizes the radiation-induced ductile-brittle transition temperature (DBTT) shift compared to steels with either lower or higher Cr content [1]. The goals implied for a *reduced activation material*, complying with waste disposal and materials recycle limits, require alloying elements Mo and Nb generally used in conventional FM steels such as Grade 91 and Grade 92 [2] be replaced by W and Ta. There are also restrictions on the content of other alloying elements (e.g., Ni, Co, etc.) and impurities (e.g., Ag, Al, etc.) in RAFM steels [3].

Current versions of RAFM steels such as F82H and Eurofer97 exhibit comparable tensile properties, though appreciably lower creep resistance than similar engineering alloys such as Grade 91. Recovery of lath boundaries and cell/packet structures occurred during creep tests, resulting in the softening of the materials during creep [4]. Unfortunately, the presence of the relatively large amount of $M_{23}C_6$ (M = Cr-rich) precipitates, up to ~1.9 vol percentage (vol%, calculated) in F82H, primarily at prior-austenite grain and cell/packet boundaries cannot effectively pin the grain boundaries during creep due to their easy coarsening to $> \sim 200$ μm from initial $\leq 100$ nm. In contrast, MX-type (M = Nb/Ta/V, X = C/N) nanoprecipitates, usually with a size of ~50 nm, have been shown to provide excellent stability at similar temperatures [5,6], but only limited amounts of MX nanoprecipitates are present in current RAFM steels, e.g., calculated to be ~0.12 vol% in F82H, well below the ~0.35 vol% (calculated) in Grade 91. The low density of MX precipitates provide little grain boundary pinning, hence the lower creep resistance of RAFM as compared to Grade 91.

In addition to RAFM steels, oxide dispersion-strengthened (ODS) alloys, also named as nanostructured ferritic alloys (NFA) recently, with Cr content generally in the range of 9–20 wt% are also being developed for fusion and other applications. Unlike

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RAFM steels manufactured using traditional industrial scale vacuum induction melting (VIM) and remelting techniques such as electro-slag remelting (ESR) and vacuum arc remelting (VAR), the NFA/ODS alloys are fabricated by complex mechanical alloying (MA) steps, then hot consolidation, as schematically shown in Fig. 1 [7]. MA enables the incorporation of large amount of oxide nanoclusters. The predominant small oxide nanoclusters (<10 nm in size) in NFA/ODS alloys have been reported in the range from ~0.7 vol% in ODS-Eurofer (9Cr: nanoclusters of ~5 nm and $1.2 \times 10^{23}$ $m^{-3}$ [8]) to ~1 vol% in PM2000 (20Cr: ~20 nm and $2.5 \times 10^{21}$ $m^{-3}$ [9]) and ~3 vol% in 14YWT (14Cr: ~3 nm and $2 \times 10^{24}$ $m^{-3}$ [10]).

The extremely large amount of oxide nanoclusters in NFA/ODS alloys effectively pin the sub-micrometer ferritic grains, leading to superior creep resistance of these alloys. The oxide nanoclusters also potentially help radiation resistance and helium management, by acting as sink/trapping sites for irradiation produced defects and helium atoms. As shown in Fig. 2 (adapted from Ref. [11]), the increase of sink strength, primarily contributed by the increased amount of nanoparticles in going from Eurofer97 to ODS-Eurofer and 14YWT, noticeably decreases radiation hardening at 250 and 300 °C. Despite the high strength and radiation resistance advantages of NFA/ODS alloy, the MA production of NFA/ODS alloys results in high fabrication cost, small volume of products, limitations on the complexity of the final product, and a high likelihood of mechanical property anisotropy and low toughness. RAFM steels have approximately the opposite features compared to NFA/ODS alloys, as summarized in Fig. 2. As suggested in Fig. 2, a class of advanced RAFM steels, called castable nanostructured alloys (CNAs), is between RAFM and NFA/ODS alloys in capturing the positive engineering attributes of each alloy system for nuclear applications. The CNAs are manufactured using the traditional industrial scale steelmaking methods used for RAFM steels, favoring flexible product forms such as forgings, plates, pipes, tubes, as well as castings as the ASTM standardized Grade 91. The CNAs contain a significantly increased amount of MX nanoprecipitates, leading to superior properties embracing the merits of current RAFM steels and NFA/ODS alloys.

## 2. Alloy microstructural design

Two routes were explored for increasing the amount of MX nanoprecipitates in CNAs. One is the generally pursued route of increasing V, Ta, and N content to form (V,Ta)N. The other is to include Ti and restrict N, which favors carbide formation, e.g., (Ti,Ta)C. The amount of $M_{23}C_6$ is also reduced, which is expected to improve yield strength and creep resistance [12,13]. As the austenite stabilizers are limited to C, N, and Mn in RAFM/CNA steels, the content of other alloy elements such as Cr, W, Ta, V, Si, and Ti (all ferrite stabilizers) was balanced using computational alloy thermodynamics. The initial composition of CNAs in wt% is targeted to be in the range of 8.3–9.0 Cr, 1.0–1.5 W, <1.0 Mn, 0.1–0.3 V, 0.05–0.15 Ta, <0.2 Ti, <0.2 Si, 0.08–0.15 C, <0.06 N with the balance Fe. The content of each alloying element has been guided by the results of computational alloy thermodynamics, together with literature findings. For instance, increasing W had a more beneficial effect on fatigue life compared with increasing Ta [14]. Increasing W content to 0.7 wt% was found to reduce minimum creep rate [15], going to 2 wt% increased DBTT [16,17], and further increases to 3 wt% and beyond would have the undesirable effect of stabilizing δ-ferrite and favoring more $Fe_2W$ Laves phase formation [18,19]. Increasing Ta in the range of 0.06–0.15 wt% reduced the prior-austenite grain size and favored small spherical (Ta,V) (C,N) precipitates [20–22]. However, the higher Ta content deteriorated creep-fatigue resistance with increasing W [23]. Increasing Si generally decreased prior-austenite grain size and favored Laves phase formation [24,25]. Impact toughness was improved by the refined grains, but decreased significantly when Si content exceeded 0.22 wt% due to pronounced Laves phase formation [24].

Fig. 3 shows an example of alloy design using computational alloy thermodynamics coupled with strength modeling. Computational alloy thermodynamics calculations were conducted with Pandat software [26] with the in-house developed OCTANT database [27]. The calculated temperature-dependent phase mole fractions in Fig. 3a show the precipitate phases in two CNAs and are compared to Eurofer97. To have better readability, the precipitate phases in the CNA3 (similar to those in the CNA2) and the predominant phases (i.e., ferrite [α and δ], austenite [γ] and liquid from low to high temperatures) in the alloys are not included in the plot. The CNA1, with a chemical composition listed in Table 1, is primarily strengthened by MN, e.g., (V,Ta)N, which was developed following the outline of Klueh [28] and related studies [22,29]. In contrast, the CNA2 and CNA3 (Table 1) are primarily strengthened by MC, e.g., (Ti,Ta)C. These new alloy compositions favor the formation of larger amounts of MX in the CNAs than found in Eurofer97. Unlike the stable MC in the CNA2 and CNA3, MN in the CNA1 and Eurofer97 undergoes a sluggish phase transformation into Z-phase, primarily in the form of CrTaN, at temperatures below ~750 °C. The coarse Z-phase consumes fine MN during long-term tests and services, so is expected to impair the strength of the

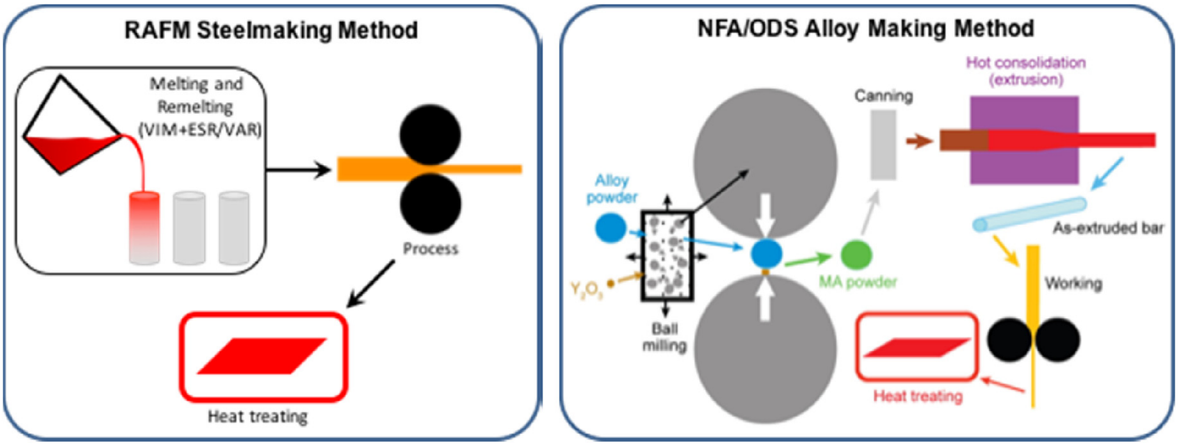

**Fig. 1.** Schematic manufacture routes of RAFM steels and NFA/ODS alloys [7].