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Crack Growth Mechanism in Defective Fuel Cladding Tubes in Reactor

Young Suk Kim [a]
[a]*Hydrogen Cracking Team, Korea Atomic Energy Research Institute, 150, Dukjin-dong, Yuseong, Daejeon 305-353, Republic of Korea*
[*]*Corresponding author: yskim1@kaeri.re.kr*

1. Introduction

Despite DHC occurring by nucleation, growth and cracking of hydrides, all the classical DHC models termed the old models [1-2] assumed that the DHC rate or the crack growth rate (CGR) was governed solely by the hydride growth rate, ignoring the rates of nucleation and cracking of hydrides. This simple assumption is the cause of many unresolved DHC issues[3]. In contrast, a new DHC model termed Kim's model demonstrates that the CGR is governed by the rate of the slowest process among the three processes involved in DHC[4-5], which can explain all the unresolved DHC issues.

Since the late 1980s, a number of fuel rods with long axial splits have been seen in boiling water reactors (BWRs) [6-8], which remains unresolved to date despite two different mechanisms such as DHC [8] or corrosion hydrogen cracking (CHC)[6] being proposed. Considering that the axial splits occurred in irradiated Zircaloy-2 cladding with as low a hydrogen concentration as 150 ppm even in isothermal conditions at 300°C without an overtemperature, the former was rejected. However, the latter that was proposed based on the model experiment termed the split propagation laboratory investigation test (SPLIT) needs more detailed explanations about the enhanced CGR due to corrosion. The aim of this work is to understand the mechanism of long axial splits in defective BWR fuel rods that have grown during their continued operation, using Kim's model.

2. Crack growth in defected fuel rods

Considering that zirconium alloys are one of the hydride forming metals, most failures of zirconium alloy components used in reactors must occur due to the cracking of hydrides precipitated in the stressed region. Likewise, the long axial splits of the defective fuel rods in BWRs should occur by the cracking of hydrides, not by hydrogen cracking. Evidence is provided by the presence of reoriented hydrides nucleated at the end of the axial splits and the brittle fracture surfaces [7]. Furthermore, considering that crack advance in defective BWR fuel rods occurs discontinuously along the hydrides precipitated at the crack tip[8], the failure mechanism to cause the long axial splits must be DHC although all the classical DHC experiences and models where reoriented hydrides grow through the bulk diffusion in a closed system cannot explain the SPLIT results, demonstrating crack growth at 180°C without a thermal cycle producing an overtemperature. A non-penetrating crack was seen in a PWR cladding despite a rapid increase of the rod power in a short time and holding the high power for a certain period as shown in Fig. 1 [9]. The result of Fig. 1 provides evidence for the validity of Kim's model demonstrating that the slowest rate of hydride cracking at higher temperature is the cause of crack arrest above 300°C and the rate of hydride nucleation has nothing to do with the crack arrest. Note that a through-wall crack occurred after a step-wise power ramp in a BWR cladding whose inside temperature was quite lower when compared to that of the PWR cladding [9].

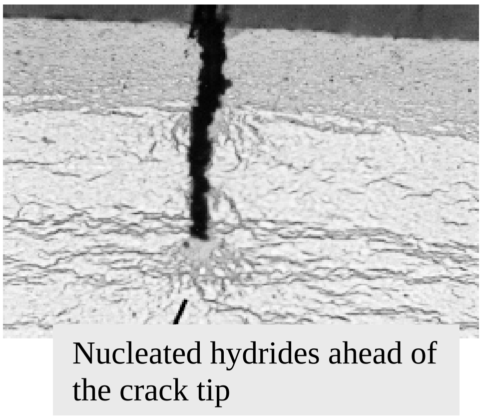

Fig. 1. Arrest of a crack in a PWR fuel rod during a power ramp and holding at the peak power for a certain period despite the reoriented hydrides nucleated at the crack tip (taken from Ref. 9).

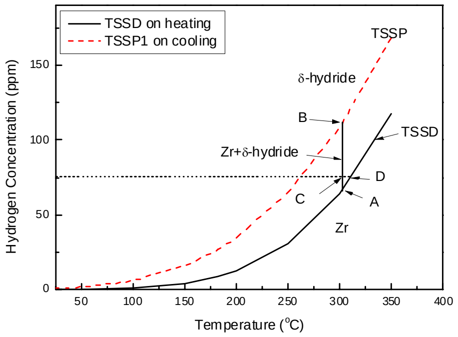

Fig. 2. Phase diagram of Zr-H system indicating the co-existence of zirconium and hydrides at the hydrogen concentrations ranging from A to B at 300°C.

A sufficient condition for hydride nucleation is a hydrogen supersaturation above TSSD as shown in Fig. 2: at the hydrogen concentration above point A, hydride can coexist with hydrogen dissolved in zirconium matrix but above point B, all the hydrogen transforms to the hydride phase of δ. The absorbed hydrogen by the cladding comes from the two sources: one is oxidation of the inside cladding and the UO_2 fuel in steam or water after the ingress of the coolant inside the cladding, and the other is oxidation of the outside cladding. In case of defective BWR fuel rods with a primary defect, most of hydrogen is produced primarily from oxidation of UO_2 fuel and cladding or the pure zirconium liner in case of a zirconium-liner cladding tube, and then diffuses from the inner surface to the outer surface along a temperature gradient. The thick oxide layer formed on the zirconium liner would contribute to the closure of the fuel-cladding gap, providing the primary site for hydride nucleation due to the stress. Hydrogen moving from the outer surface to the inner surface due to the temperature gradient will go primarily to the most stressed region, overridingly producing a hydrogen supersaturation there and causing the reoriented hydride to be nucleated. Note that since all the zirconium alloys exposed to water in reactors absorb hydrogen from the environment so that they behave like an open system, hydrogen picked up by the cladding can move primarily to the stressed region, not to the unstressed region. This circumstance is different from out-of-pile DHC tests using zirconium alloys containing hydrogen that is a closed system because no hydrogen is entered into or coming out of the specimen. Evidence for the role of the temperature gradient is provided by the simulating experiments demonstrating that a crack grows from the outside cladding to the inside cladding only under a temperature gradient but no crack grows without it despite the same stress being applied. According to Kim's model[3-5], the crack growth rate of zirconium alloys below 300^{o}C is governed primarily by the hydride growth rate and to a less extent by the hydride cracking rate only if the hydrogen concentration in the stressed region is large enough to exceed the difference between the terminal solid solubility of hydride precipitation termed TSSP1 and TSSD. In short, hydrogen diffusion governs the hydride growth rate and hence the CGR below 300^{o}C because it is the slowest process as compared to the other processes. Since the hydrogen is supplied continuously from the inner cladding surface through the temperature gradient, the initial hydrogen concentration of the defective cladding does not affect the CGR at all. Thus, considering that the temperature gradient enhances diffusion of hydrogen toward the outer surface of the cladding from the inner surface in addition to the concentration gradient, the CGR of the defected fuel rods will be much higher when compared to that of the intact fuel rod where hydrogen moves to the stress region from the bulk of the cladding only by the concentration gradient without the temperature gradient. Hence, it is recommended that the DHC tests to simulate crack growth of the zirconium alloys in reactors shall be conducted with hydrogen picked up continuously from the environment.

4. Conclusions

Given the fact that crack advance in defective BWR rods occurs discontinuously along the hydrides precipitated at the crack tip, it is evident that the failure mechanism of the long axial splits of the defective fuel rods in BWR is DHC. Since the source of hoop stress results from oxidation of UO_2 fuel and cladding, the preferential site for crack initiation corresponds to the fuel-cladding gap closure region due to fuel expansion or the heavy oxide layer on the inside cladding surface due to faster oxidation of the zirconium spongy liner in case of a BWR zirconium liner fuel. As hydrogen produced by corrosion of cladding and or UO_2 fuel in water or steam can move to the crack tip on the outer cladding surface exposed to the lower temperature, the hydrogen supersaturation that is a necessary condition to nucleate hydrides according to Kim's model can be accomplished even in isothermal conditions. In addition, the temperature gradient formed across the thickness of the fuel rod also makes it faster for hydrogen to diffuse from the bulk to the crack tip in the outer surface, thus, enhancing the DHC rate. Consequently, it is demonstrated that Kim's DHC model explains the axial splits in defective BWR fuel rods.

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