COMPATIBILITY OF MATERIALS FOR USE IN
LIQUID-METAL BLANKETS OF FUSION REACTORS* $CONF-830942 -- 72$

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DISCLAIMER

COMPATIBILITY OF MATERIALS FOR USE IN LIQUID-METAL BLANKETS OF FUSION REACTORS*

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A review of corrosion and environmental effects on the mechanical properties of austenitic and ferritic steels for use with liquid metals in fusion reactors is presented. The mechanisms and kinetics of the corrosion processes in liquid lithium and Pb-17Li systems are examined and their influence on degradation of structural material is discussed. Requirements for additional data are identified.

1. INTRODUCTION

The various blanket concepts that have been proposed for fusion reactors operating on a D-T fuel cycle require lithium or a lithium-bearing material for tritium production, a circulating coolant for heat recovery, and a structural material to contain and support the blanket system. Several blanket designs use liquid lithium or molten lithium alloys, e.g., Pb-17 at. % Li, as a breeder and possible coolant. The liquid-metal breeder/coolant concept provides an attractive option because of the simplicity of the system. Liquid lithium has long been considered an attractive blanket material because of its acceptable tritium-breeding and efficient heat-transfer characteristics. Liquid lithium/lead alloys can provide adequate tritium breeding with a reduced lithium inventory since the lead would serve as a neutron multiplier.

The use of liquid metals and alloys for such applications requires an assessment of their compatibility with the containment materials. The two major compatibility concerns arising from the use of liquid metals, such as lithium or Pb-17Li, are (a) corrosion and mass transfer, and (b) degradation of mechanical properties of the containment material. Corrosion in the form of uniform or selective dissolution, inter-granular penetration, and interstitial-element transfer can result in (a) significant wall-thinning/wastage and (b) deposition of corrosion products that may cause severe flow restrictions and excessive accumulation of radioactive material in unshielded regions. The former consequence results in a loss of mechanical integrity and the latter phenomenon would increase requirements for pumping power, decrease the energy conversion efficiency, and complicate system maintenance. The severity of the corrosion and mass transfer problem varies for different combinations of containment material and liquid metals and depends on many material and process variables.

Deterioration of the mechanical properties of structural materials can result from the influence of the environment itself and the effects of microstructural and compositional changes that occur in the material during long-term high-temperature exposure to the liquid metal environment. Liquid metals can influence the surface-active properties of the material through liquid metal embrittlement

(LME), oxidation, nitridation, or carburization-decarburization phenomena, all of which can alter the near-surface deformation behavior and thus affect mechanical properties such as fatigue crack propagation, creep ductility, etc.

2. CORROSION

The principal mechanisms of corrosion of materials in a liquid metal environment are dissolution, intergranular penetration, interstitial-element transfer, and mass transfer due to thermal and concentration gradients. The contribution of these mechanisms to corrosion varies with the composition and purity (i.e., concentration of N, O, C, H, etc. in liquid metal) of the specific liquid-metal system; composition and microstructure of the containment material; operating temperature and time; and system parameters such as flow velocity, ΔT (difference in temperature across the circulating system), surface area and temperature profile of the system, and material combination (i.e., bimetallic or monometallic system). A liquid metal system used solely for tritium breeding could be semistagnant with small ΔT, whereas the use of liquid metals as coolant requires high flow velocities and large ΔT. Thermal-gradient mass transfer occurs in nonisothermal systems because of the temperature dependence of the solubilities of the constitutive elements of the containment material.[1] Alloy elements dissolve in the hot regions of the liquid-metal loop because their solubility exceeds their bulk concentration in the liquid metal, and they are deposited in the cold regions of the loop where their concentration in the bulk stream is greater than the solubility limit. The overall rate of mass transfer is strongly influenced by the kinetics of the dissolution and deposition

reactions, which depend on system parameters such as temperature, ΔT, flow velocity, etc.

Several studies have been conducted to evaluate the influence of different material and system parameters on the corrosion behavior of austenitic and ferritic steels in liquid lithium and to a lesser extent in liquid Pb-17Li environment. Data have been obtained for static isothermal liquid-metal vessels or capsules as well as for circulating systems such as thermal-convection loops (TCLs) and forced circulation loops (FCLs). Corrosion tests in static lithium or Pb-17Li systems have been used to identify and characterize the various reactions between the environment and containment material and to study the influence of applied stress and material variables on these interactions. However, control of impurities is difficult in static liquid-metal systems. Data on circulating systems have been used to investigate the dissolution and mass transfer behavior of structural materials. TCLs are appropriate for mass transfer studies at low flow velocities (~0.03 m/s) which are typical of a semistagnant tritium-breeding blanket, while FCLs, with or without a cold-trap purification system, allow the flow velocity to be varied over a much wider range. FCLs are more representative of self-cooled blanket systems. For most studies, the corrosion behavior has been evaluated in terms of weight loss per unit surface area and/or as depth of internal corrosive penetration.

Corrosion tests conducted in circulating liquid metal at temperatures between 700 and 873 K (427 and 600°C) indicate that the dissolution rates of austenitic and ferritic steels in Pb-17Li are an order of magnitude greater than in lithium.[2,3] In both environments, the dissolution rates for austentic stainless steels reach steady-state

(time-independent) values after an initial period of 1500 to 3000 h which is characterized by higher, rapidly decreasing rates.[2-6] This type of dependence is shown in Fig. 1. The variation in dissolution rate with time is determined from weight change measurements for a single specimen exposed for different lengths of time. However, dissolution rates are often expressed as the average value for the entire time of exposure, i.e., rates A and B in Fig. 1. The average dissolution rates are strongly dependent on test duration; they are higher than the steady state rates and decrease with an increase in exposure time. The rapidly changing dissolution rates are indicative of a change in the rate-controlling step for dissolution. The initial stage of high dissolution rates corresponds to the formation of a ferrite layer which results from preferential leaching of nickel, and to a lesser extent, chromium from the surface of the austenitic steels. The steady-state regime is characteristic of the dissolution of an austenitic stainless steel having a ferritic surface layer depleted of nickel.

The corrosion data may be expressed by a power law relationship with time. For lithium TCL and FCL systems, the average value of the time exponent is 0.7 over extended exposure periods, which could represent an initial time dependence of $\sqrt{t}$ followed by the experimentally observed linear dependence with time. The weight loss data in circulating Pb-17Li for exposure times up to ~3500 h yield a time exponent of 0.5, shown in Fig. 2.[2] The weight loss in flowing Pb-17Li is significantly greater than in static capsules. Specimens exposed to static Pb-17Li show little or no weight loss after ~1000 h, whereas the specimens continue to lose weight in flowing Pb-17Li. Weight loss in a static

liquid-metal system is expected to reach a saturation value and stop because of the gradual buildup of the alloy elements in the liquid metal. Data for longer time periods are needed to accurately establish the dissolution rates in a Pb-17Li environment.
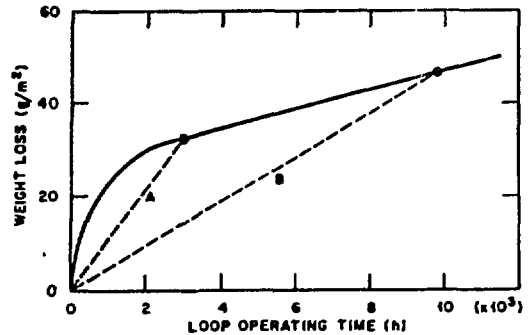

FIGURE 1
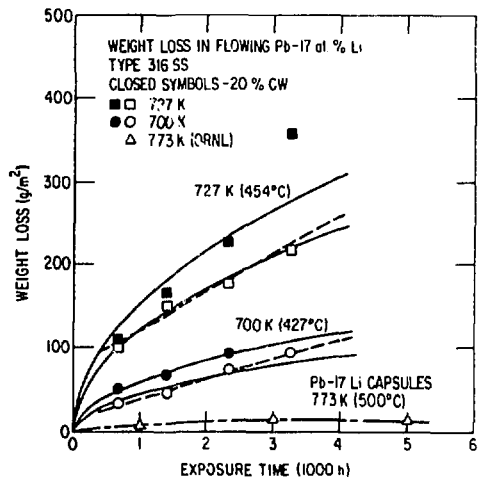Corrosion of austenitic stainless steel in flowing lithium or Pb-17 environments.

FIGURE 2
Weight loss for Type 316 stainless steel exposed to flowing Pb-17Li at 727 and 700 K. Each symbol represents weight loss for a single specimen after various exposure times (Ref. 2).

The weight loss for ferritic steels of medium chromium content, such as HT-9 alloy (Fe-12Cr-1MoVW) and Fe-9Cr-1Mo, in lithium or Pb-17Li TCL and FCL systems, follow a linear relationship with time.[2,7] The near-surface region of the steels is slightly depleted of chromium to a depth of 5 to 10 μm.

2.1. Influence of liquid metal purity

The influence of interstitial elements, such as nitrogen and hydrogen, on corrosion of materials in lithium has been investigated; however, little or no information is available on the effects of impurity content in a Pb-17Li environment. Several studies indicate that the concentration of nitrogen in lithium is an important factor in controlling corrosion of materials. Weight loss of austenitic and ferritic steels in circulating lithium systems contaminated with nitrogen is significantly greater than in systems with pure lithium.[7-13] Tests on static lithium systems have also shown that the corrosion of austenitic stainless steels is very sensitive to the nitrogen content of lithium[14-16] and of the steel.[17,18] The latter is particularly significant at temperatures >873 K. Increases in dissolution rates by factors of 1.5 to 3 have been observed to result from an increase in nitrogen content from ~100 to 300 wppm in a lithium FCL system operating at 633 to 755 K[7] and in a lithium TCL system at 866 to 1144 K.[8] In addition to nickel depletion, the specimens exposed to nitrogen-contaminated lithium showed preferential leaching of chromium from austenitic stainless steels.[12,13] A nitrogen-alloyed Cr-Mn steel exposed to static lithium at 873 K formed a Li-Cr-N compound at the grain boundaries.[18] A ternary nitride, Li_9CrN_5, was identified on the surface of Type 316 SS that was exposed at 748 K (475°C) to nitrogen-saturated lithium.[19,20] These results show the significance of chromium-nitrogen interactions on the corrosion of materials in lithium.

Addition of hydrogen to lithium appears to improve the corrosion resistance of Fe-Ni-Cr and Fe-Mn-Cr structural steels.[21-23] The depth of the internal corrosion layer and intergranular penetration in these steels is decreased when the lithium system contains hydrogen.

2.2. Influence of material composition and microstructure

The effects of alloy composition on the dissolution rates of ferrous alloys in flowing lithium have been established. In general, the corrosion rates increase with an increase in the nickel content of the material.[7,24-26] The mass transfer rates of Fe-Ni-Cr alloys in lithium TCL systems increase proportionally with increasing nickel concentration between 10 and 30 wt %.[25,26] Similar behavior is also observed in lithium FCL systems: weight loss of prime candidate alloy (PCA), containing ~16 wt % Ni, is a factor of ~1.5 greater than for Type 316 SS, with ~10 wt % Ni.[7] However, the dissolution rates of austenitic Type 316 SS in FCL systems[4-7] are an order of magnitude greater than in TCL systems.[26] The various FCL systems used for corrosion studies were equipped with a cold-trap (maintained at ~400 K) purification loop. The greater dissolution rates in FCLs may arise from differences in the mass transfer and deposition behavior, particularly of nickel. In lithium TCLs, the mass transfer deposits are initially pure chromium but gradually incorporate significant concentrations of iron and nickel.[27] Somewhat different behavior is observed in lithium FCL systems. Crystals of iron, nickel-manganese, and nickel-iron with a few chromium-rich deposits have been found in the electromagnetic pump ducts and magnetic

flowmeter tubing of FCLs.[11,28] Use of magnetic traps and/or the low-temperature cold traps in FCL systems may result in the different mass transfer/deposition behavior.

The dissolution rates of ferritic Fe-Cr-Mo steels are comparable in both TCL and FCL systems. However, the dissolution rates of ferritic steels, relative to the long-term steady-state rates of Type 316 SS, are an order of magnitude lower in FCL systems[6,7] and are similar in TCL systems.[29]

The influence of alloy composition on corrosion of ferrous alloys has also been observed in a Pb-17Li FCL system. The weight loss of ferritic HT-9 alloy and Fe-9Cr-1Mo steel is an order of magnitude lower than that for Type 316 SS.[2]

Micrographs of the ferrite layer on Type 316 SS exposed to lithium at 873 K and to Pb-17Li at 727 K are shown in Figs. 3 and 4, respectively. In both environments, the ferrite layer is very weak because of the high porosity, and it can easily spall under high liquid velocities, thermal transients, or cyclic straining. Consequently, the corrosion process will be accelerated by the re-exposure of fresh surface containing significant amounts of nickel. This behavior will result in relatively high rates of dissolution without attainment of the steady-state condition. The formation of a porous ferrite layer has a significant effect on the mechanical behavior of austenitic stainless steels and requires a better understanding of the kinetics and nature of the internal corrosive attack.

A detailed study of the porosity of Type 316 SS exposed over the temperature range of 793 to 873 K in a TCL operating between 723 and 873 K, showed that porosity decreased as weight loss and temperature decreased.[3,30] However, there was no correlation between the extent of nickel or chromium depletion and the changes in pore size and distribution. These results indicate that the porosity is not simply related to the nickel flux out of the steel but may depend on local reactions that form various corrosion products. Limited data on lithium TCL and FCL systems indicate that there may be a linear increase in thickness of the ferrite layer over time.[31] Figure 5 shows the thickness of the ferrite layer on Type 316 SS that was exposed to lithium at 755,

FIGURE 3
Type 316 stainless steel exposed for 8000 h in flowing lithium at 873 K.

FIGURE 4
Ferrite layer formed on annealed Type 316 stainless steel exposed for 1400 h in flowing Pb-17Li at 727 K (Ref. 2).
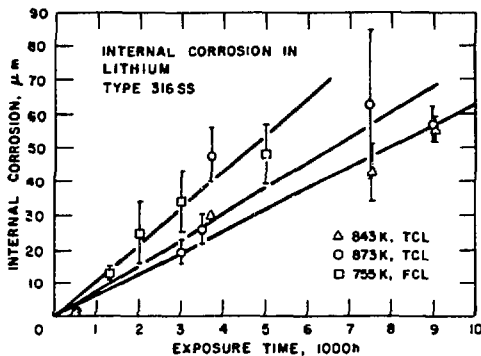
FIGURE 5
Depth of internal corrosion of Type 316 stainless steel exposed to flowing lithium.

843, and 873 K. Large differences are observed between TCL and FCL systems. The results yield a growth rate of ~100 μm/y at 755 K in an FCL and 55 μm/y at 843 K in a TCL.

The microstructure of the material also affects the corrosion behavior. The low alloy (Fe-2 1/4Cr-1Mo) ferritic steels containing relatively unstable carbides such as Fe_3C and ε-carbide, exhibit significant internal corrosion and decarburization in lithium,[32] whereas, steels with stable carbide phases, e.g., normalized and tempered Fe-2 1/4Cr-1Mo steel or HT-9 alloy, show little or no internal corrosion and carbon loss.[33] The weight loss for 20% cold-worked Type 316 SS in circulating lithium and Pb-17Li environments at temperatures between 700 and 755 K is a factor of ~2 greater than that for annealed steel.[2,7] However, the weight loss for 20% cold-worked PCA exposed to lithium at 843 and 873 in a TCL system was comparable or slightly lower than that of annealed material.[26]

2.3. Influence of temperature

The steady-state dissolution rates for Type 316 SS have been investigated in a lithium TCL at maximum temperatures of 773, 823, and 873 K.[34] An Arrhenius plot of the data yields an activation energy of 38 kcal/mole. The combined data from lithium FCL systems[5-7,35,36] for austenitic Types 304 and 316 SS show considerable scatter and yield values of activation energy between 11 and 30 kcal/mole. However, such analyses include data obtained for <100 h exposure time or a difference in exposure temperature of <50 K. For some studies, the dissolution rates are expressed as average values rather than as steady-state values.[35]

Limited data indicate that an increase in flow velocity or ΔT of the system increases the rates of dissolution of structural steels.[5,9] The effects of velocity on corrosion of austenitic stainless steel have been established in flowing sodium.[37] The influence of flow velocity on corrosion is dependent on the rate-controlling step for dissolution. For example, if the rate controlling mechanism is solid-state diffusion or a corrosion product reaction at the solid-liquid interface, then the dissolution rate will be independent of the liquid metal velocity. The opposite is true if dissolution is controlled by liquid-phase diffusion. Existing data are insufficient to precisely identify the rate-controlling mechanism for dissolution. For austenitic stainless steels, the initial parabolic relationship of dissolution rates with time suggests an initial dependence on flow velocity. Magnetohydrodynamic effects arising from the flow of a liquid metal across the magnetic field lines of a magnetically-confined fusion reactor may also influence the corrosion rate through changes in the fluid mechanics parameters.[38] Additional data are required to determine the nature and magnitude of such effects.

Because of the many variables, viz., lithium purity, temperature, ΔT, flow

velocity, surface area, etc., it is difficult to compile a definitive data set of measured corrosion rates in flowing lithium. An activation energy of 38 kcal/mole measured for TCL systems is comparable to the value of 36 kcal/mole determined for austenitic stainless steels in sodium.[37] Figure 6 shows the Arrhenius plot of dissolution rate data for various ferritic and austenitic steels in lithium TCL and FCL systems. The dashed lines represent the scatter band for the data in FCLs and the slope is assumed to be the same as that observed for the TCL data.
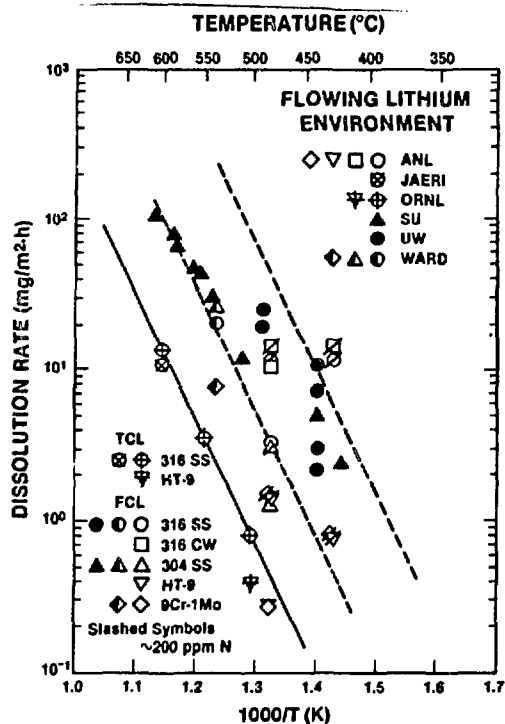

FIGURE 6
Arrhenius plot of dissolution rate data for austenitic and ferritic steels exposed to flowing lithium (Refs. 5-7, 34-36).

Relatively little data are available for studying corrosion and mass transfer behavior of structural materials in Pb-17Li. The Arrhenius plot of the dissolution rates for different austenitic and ferritic steels in Pb-17Li FCL[2] and TCL[3] systems is shown in Fig. 7. The temperature dependence is identical to that observed for lithium TCLs, i.e., an activation energy of 38 kcal/mole.

The corrosion data in Figs. 6 and 7 may be used to estimate the maximum operating-temperature limits for circulating liquid-metal systems. Assuming an acceptable corrosion and mass transfer rate of 5.5 mg/m^{2}h, i.e., equivalent to an uniform corrosion rate of ~5 μm/y, the temperature limit for austenitic stainless steels in lithium is ~723 K (~450°C) and ~653 K (380°C) in Pb-17Li systems. The limits for ferritic steels in lithium and Pb-17Li systems are ~823 and 713 K (550 and 440°C), respectively. A higher acceptable corrosion rate of 20 μm/y would increase these temperature limits by ~50 K. Data on the influence of the various systems parameters are required to accurately determine the limits of operating temperature for liquid-metal blankets.

2.4. Corrosion Inhibition

Addition of aluminium to lithium has been found to produce a stable aluminized surface layer that reduces corrosion of high-nickel Fe-Ni-Cr alloys in static lithium and of Type 316 SS in flowing lithium.[39] Adherent coatings of Mo, Mo-Ti, and Al, applied from a molten-sodium solution on nickel-bearing materials, showed improved resistance to corrosion in flowing lithium at temperatures up to 1173 K (900°C).[40] Techniques to inhibit corrosion in Pb-17Li have not been investigated.
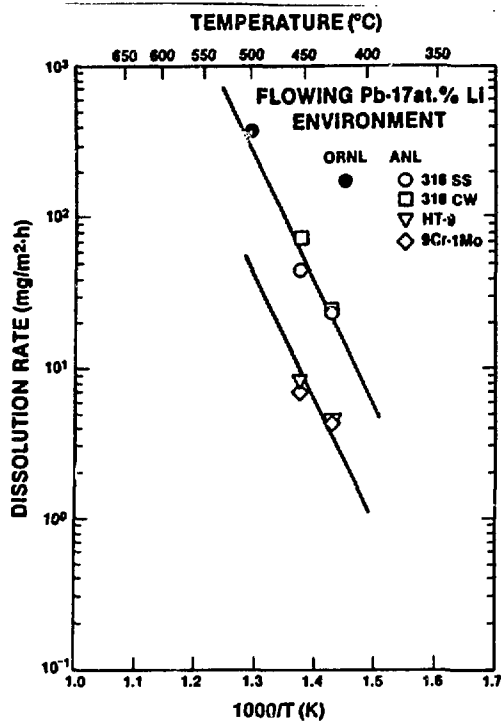
FIGURE 7
Arrhenius plot of dissolution rate data for austenitic and ferritic steels exposed to flowing Pb-17Li (Refs. 2 and 3).

3. ENVIRONMENTAL EFFECTS ON MECHANICAL PROPERTIES

The influence of the environment itself on mechanical behavior of materials is evaluated by conducting mechanical tests in the liquid-metal environment of controlled purity and comparing the results with the data obtained in other environments, such as air, vacuum, or inert gas. The long-term effects of environment, i.e., interstitial- or substitutional-element transfer and thermal aging, on the mechanical properties are determined by conducting tests on material that is either pre-exposed to the liquid-metal environment or thermally aged in an inert environment. Preexposure of the material is carried out in

liquid metals of known purity, namely, service conditions anticipated in real systems.

The tensile, creep, fatigue, and crack-growth properties of various austenitic and ferritic steels have been investigated in the liquid lithium environment. Most of the data have been obtained in static-lithium vessels or from tensile or creep tests on tube specimens filled with lithium. However, the purity of the lithium is generally not specified for these tests. Corrosion phenomena, such as intergranular penetration or preferential leaching of substitutional elements, are likely to have a profound effect on mechanical properties. Corrosion behavior of materials in liquid lithium indicates that the nature and kinetics of corrosive attack strongly depend on the concentration of nitrogen in lithium. Consequently, mechanical property data obtained in liquid metal of unknown purity may be biased by the effects of the corrosive interactions and not representative of the anticipated service conditions. Significant results from studies of the influence of lithium environments on mechanical properties of austenitic and ferritic steels are summarized below.

The tensile strength and, most particularly, ductility of Armco iron are considerably lower in static lithium than in vacuum.[41] The reduction in tensile properties occurs in the temperature range of 473 to 773 K (200 to 500°C) and is more pronounced at slow strain rates. At the critical temperatures the material exhibits low ductility (~5% total elongation) and fails by a brittle-fracture mode. The creep-rupture strength of Armco iron is also reduced in lithium at temperatures <773 K. The tensile and creep strength of austenitic stainless steels is not significantly affected in either static or flowing lithium.[42-44] The influence of

lithium on tensile ductility of austenitic steels depends on strain rate, e.g., elongation is slightly reduced at slow strain rate ($\sim$4 $\times$ 10^{-5} s^{-1}).

The influence of long-term exposure to lithium on tensile properties depends on the corrosion behavior. Factors that increase corrosion of materials also reduce tensile strength. For example, the tensile strength of Fe-18 Cr-9Ni stainless steel does not change significantly after exposure to static lithium but shows a sharp reduction when exposed to flowing lithium.[44] The changes in tensile properties of Types 304 and 316 and Cr-Mn stainless steels, exposed for up to 1500 h at 873 K to static lithium or lithium containing 40 ppm hydrogen, are attrributed to thermal aging, with minimal influence of preexposure.[45,46] Formation of a thin ferrite layer (10- to 30-μm thick) on the surface of the preexposed specimens has a negligible effect on mechanical properties. Preexposure of HT-9 alloy to static lithium at 773 K has no effect on tensile properties, whereas the tensile strength of Fe-? 1/4Cr-1Mo steel exposed to lithium at 773 and 873 K is reduced by decarburization.[33]

Nitrogen content in lithium plays an important role in controlling the fatigue properties of materials. Low-cycle fatigue life of austenitic Types 304 and 316 SS and HT-9 ferritic steel in low-nitrogen (e.g., <100 wppm) flowing lithium at 755 K (482°C) is superior to that in air.[43,47-49] Limited data also indicate that fatigue life of HT-9 alloy is independent of strain rate in low-nitrogen lithium (see Fig. 8).[49] However, fatigue life of the HT-9 alloy is significantly reduced in flowing lithium containing >1000 wppm nitrogen due to intergranular attack of the material. The reduction in life is greater at slow strain rates. Preexposure of HT-9 to

low-nitrogen lithium for 1100 h at 755 K has no effect on fatigue life,[49] whereas preexposure of Type 316 SS to lithium reduces life because of nickel depletion and formation of the ferrite layer.[43]

Fatigue crack-propagation rates (da/dN) for Type 304 SS and Fe-2 1/4Cr-1Mo ferritic steel at temperatures between 673 and 873 K (400 and 600°C) are greater in static lithium than in argon.[50-53] The growth rates depend on temperature and test frequency. For austenitic Type 304 SS, the growth rates increase with a decrease in test frequency. The crack growth rates for Fe-2 1/4Cr-1Mo steel in lithium at 673 and 773 K are minimum at intermediate test frequencies (see Fig. 9). A high density of cleavage-type fracture is observed in lithium. The increase in growth rates is attributed to corrosive interactions at low test frequencies and due to strain-rate-induced liquid-metal-enbrittlement at high frequencies.[53] The latter assumes that embrittlement occurs when the flow stress, increased by high strain rate and low temperatures, exceeds a critical value. However, the fatigue crack propagation tests were conducted in static lithium containing $\sim$2000 wppm nitrogen. The role of lithium purity on the embrittling effects of the environment are not established.

Barely any information is available on the influence of liquid Pb-17Li on the mechanical properties of materials. The influence of applied stress on the corrosion behavior of notched tensile specimens of Type 316 SS has been investigated in Pb-17Li at 673 K.[54] The results revealed several cracks filled with Pb. The current assessment of the environmental effects in Pb-17Li are based on mechanical property data obtained for liquid lead, bismuth, or lead-bismuth environments.[1]
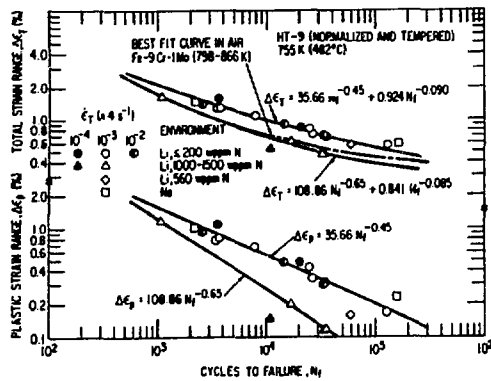
FIGURE 8
Total and plastic strain range vs cycles to failure for the HT-9 alloy tested in flowing lithium and sodium at 755 K (Ref. 49).
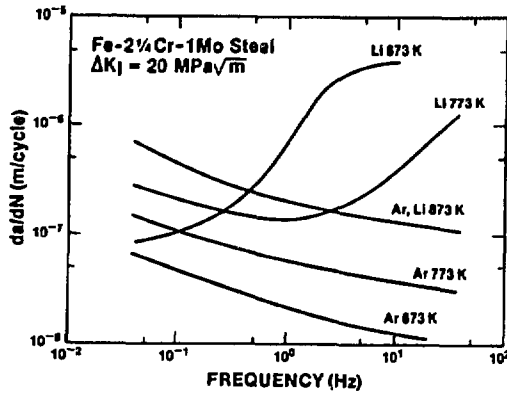

FIGURE 9
Variation of crack growth rate, da/dN, with loading frequency for Fe-2 1/4Cr-1Mo steel in lithium and argon at 673 K (Ref. 53).

4. SUMMARY

Corrosion and materials compatibility are major considerations in assessing the viability of the different liquid-metal blanket designs. Corrosion processes in circulating liquid lithium may include dissolution, intergranular penetration, and mass transfer. Increase in nitrogen content in lithium (≥ 100 wppm) increases corrosion and may lead to intergranular attack of austenitic and ferritic steels. Although additional data are required to define the corrosion processes in lithium, the use of austenitic stainless steels for service in lithium will probably be limited to 723 to 773 K (450 to 500°C) because of deleterious effects of mass transfer and deposition at higher temperatures. Corrosion-inhibition techniques can increase the resistance to corrosion in flowing lithium (with or without nitrogen impurities) and thus raise the operating temperature limits. The ferritic steels are more resistant to corrosion in lithium. Mass transfer and deposition effects are not likely to be the limiting factor at temperatures below 823 K (550°C). The corrosion and mass transfer behavior varies for the many possible combinations of containment material and liquid metals. Data on the effects of various system and material parameters on corrosion are needed to establish the maximum operating temperatues for different fusion reactor blanket systems.

Preliminary data indicate that the use of austenitic stainless steels with liquid Pb-17Li may be impractical: the high corrosion and mass transfer of austenitic stainless steels in Pb-17Li would probably limit the operating temperatures below 703 K (430°C). The corrosion behavior of ferritic steels in Pb-17Li at temperatures up to 763 K (490°C) is similar to that of the austenitic stainless steels in lithium. Additional data are required to evaluate the corrosion and mass transfer behavior of materials in a Pb-17Li environment.

A low-nitrogen lithium environment itself has no deleterious effects on the mechanical properties of austenitic stainless steels. In a lithium environment of controlled purity, the tensile and creep properties are similar

and the fatigue properties are superior to those in air. Long-term exposure of austenitic stainless steels to flowing lithium can reduce mechanical strength because of internal corrosion and the formation of the ferrite layer.

Embrittlement of ferritic steels, particularly low-alloy steels, may occur in lithium at temperatures between 473 and 673 K (200 and 400°C). Data in flowing lithium of controlled purity are required for a reliable assessment of the environmental effects on the mechanical properties of ferritic steels.

No data are available on the influence of a Pb-17Li environment on mechanical properties. The current assessment of environmental effects is based on data for Pb or Bi. A Pb-17Li environment may significantly affect mechanical properties of materials.

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